TRANSACTIONS

of the

American Foundrymen's Society



Proceedings of the Sixty-Third

Annual Meeting, April 13-17, 1959

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Merle C. Hansen, Editor

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OFFICERS AND DIRECTORS

of the

AMERICAN FOUNDRYMEN'S SOCIETY

(INCORPORATED)

(Year ending May 13, 1960)

President, CHARLES E. NELSON, The Dow Chemical Co., Midland, Mich.
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*H. M. PATTON, American Hoist & Derrick Co., St. Paul, Minn.
C. A. SANDERS, American Colloid Co., Skokie, Ill.

Terms expire 1962

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D. L. COLWELL, Apex Smelting Co., Cleveland, Ohio
R. E. MITTLESTEAD, Akron Standard Mold Co., Barberton, Ohio
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PROCEEDINGS SUMMARY OF 63d ANNUAL MEETING

The 63d Castings and 2d Engineered Castings Show of the American Foundrymen's Society was held in Chicago, Illinois, April 13 to 17, 1959. The company exhibitors at the 2d Engineered Castings Show included foundries, pattern shops, metals and alloy producers and testing equipment manufacturers.

Registered attendance of 4817 was less than for the previous alternate-year AFS Casting Congress, and included design engineers, castings buyers and foundrymen. About 100 papers were presented at 48 Technical Sessions, which included 6 popular shop courses and seminars.

The Charles Edgar Hoyt Memorial Lecture was delivered by Harry M. St. John, Fort Pierce, Fla., on the subject "The Control of Quality in the Brass Foundry," and is published starting on page 477 of this volume.

A summary of the sessions held follows:

MONDAY, APRIL 13

8:00 am — Author-Chairman Breakfast

9:30 am — Light Metals Session

Presiding — W. E. SICHA, Aluminum Co. of America, Cleveland.
S. LIPSON, Frankford Arsenal, Philadelphia.

Secretary — G. P. MESSENGER, Ordnance Tank-Automotive Command, Centerline, Mich.

Starch Content Effect on a Rammed Graphitic Mold Material for Casting Titanium, H. W. ANTES and R. E. EDELMAN, Frankford Arsenal, Philadelphia.

Some Principles for Producing Sound Al-7 Magnesium Alloy Castings, W. H. JOHNSON and J. G. KURA, Battelle Memorial Institute, Columbus, Ohio.

9:30 am — Malleable Session

Presiding — W. D. MCMILLAN, International Harvester Co., Chicago.
R. V. OSBORNE, Lakeside Malleable Castings Co., Racine, Wis.

Secretary — J. T. BRYCE, Albion Malleable Iron Co., Albion, Mich.

Malleable Iron A Magnetic Alloy, W. K. BOCK, National Malleable & Steel Castings Co., Cleveland.

Chemical Analysis vs. Heat Treatment Effects on Tensile Strength of Malleable Iron, Controlled Annealing Committee (6-D) Report, L. R. JENKINS, Wagner Castings Co., Decatur, Ill.

Mottling in Heavy Section White Iron Castings, Research Progress Report, C. R. LOPER, JR. and R. W. HEINE, University of Wisconsin, Madison.

9:30 am — Pattern Session

Presiding — J. M. KREINER, National Malleable & Steel Castings Co., Cleveland.
J. F. ROTH, Cleveland Standard Pattern Works, Cleveland.

Secretary — R. R. LEMASTER, R. A. Nelson Pattern Co., Milwaukee.

Core Boxes for Shell Cores, J. E. STOCK, John Deere Waterloo Tractor Works, Waterloo, Iowa.

Plastic Tooling and Patternmaking with Epoxical Resins, motion picture.

12:00 Noon — Light Metals Round Table Luncheon

Presiding — D. L. LAVELLE, American Smelting & Refining Co., South Plainfield, N. J.

What Does the Test Bar Mean to the Producer and User of Castings.

Panel — R. L. HEATH, Rocketdyne Div., North American Aviation, Inc., Canoga Park, Calif.; H. ROSENTHAL, Frankford Arsenal, Philadelphia; M. E. BROOKS, The Dow Chemical Co., Bay City, Mich.

12:00 Noon — Malleable Round Table Luncheon

Presiding — F. B. ROTE, Albion Malleable Iron Co., Albion, Mich.
J. H. LANSING, Consultant, Shaker Heights, Ohio.
E. WELANDER, John Deere Malleable Works, East Moline, Ill.

Selection of Materials for Ordnance Development Programs, V. LINDNER, Feltman Rsch. & Engrg. Laboratories, Picatinny Arsenal, Dover, N. J.

Toward More Effective Shell, motion picture.

2:00 pm — Brass and Bronze Session

Presiding — W. H. BAER, Fort Belvoir, Va.
J. E. GOTHERIDGE, Foundry Services, Inc., Columbus, Ohio.

Secretary — W. M. SPEAR, Worthington Corp., Harrison, N. J.

Electrical Conductivity of Sand Cast Copper-Base Alloys, D. G. SCHMIDT and F. L. RIDDELL, H. Kramer & Co., Chicago.

Some Foundry Problems in the Development of a New Marine Propeller Alloy, A. J. SMITH, Bethlehem Steel Co., Shipbuilding Div., Staten Island, N. Y.

2:00 pm — Pattern Session

Presiding — W. E. MASON, Westinghouse Air Brake Co., Wilmerding, Pa.
N. C. JONES, New York Air Brake Co., Watertown, N. Y.

Secretary — J. W. TIERNEY, Houghton Laboratories, Inc., Olean, N. Y.

Pattern Engineering Materials, new horizons, J. E. OLSON, Dike-O-Seal, Inc., Chicago.

Construction of and Materials for CO_2 Pattern & Core Box Equipment, R. J. CARVER, Carver Foundry Products Co., Muscatine, Iowa.

2:00 pm — Sand Session

Presiding — R. H. JACOBY, St. Louis Coke & Foundry Supply Co., St. Louis.
W. A. WRIGHT, Woodruff & Edwards, Inc., Elgin, Ill.

Secretary — W. O. MCFATRIDGE, International Harvester Co., Chicago.

Experimental Determination of Specific Surface and Grain Shape of Foundry Sands, FRANZ HOFMANN, George Fischer, Ltd., Schaffhausen, Switzerland; *presented by* V. ROWELL, Harry W. Dietert Co., Detroit.

Optimum CO_2 Molding, critical formulations of sodium silicate and sand, R. J. COWLES, Walworth Co., Braintree, Mass.

New Foundry Resins and Application Techniques for Shell Molds and Shell Cores, W. C. CAPEHART, Monsanto Chemical Co., Springfield, Mass.

4:00 pm — Brass and Bronze Seminar

Presiding — R. A. COLTON, Federated Metals Div., American Smelting & Refining Co., Houston, Texas.
C. E. MULLER, Ingersoll-Rand Co., Phillipsburg, N. J.

The Metallurgical and Foundry Aspects of Quality Control.

Nondestructive Testing, S. GOLDSPIEL, New York Naval Shipyard, Brooklyn, N. Y.
Mechanical Aspects of Testing, A. J. SMITH, Bethlehem Steel Co., Shipbuilding Div., Staten Island, N. Y.
Fracture Testing in a Production Foundry, C. W. WARD, American Radiator & Standard Sanitary Corp., Cincinnati.

4:00 pm — Light Metals Session

Presiding — C. A. KOERNER, Central Foundry Div., General Motors Corp., Saginaw, Mich.
D. L. LAVELLE, American Smelting & Refining Co., South Plainfield, N. J.
Secretary — J. G. KURA, Battelle Memorial Institute, Columbus, Ohio.
Radiography, Microstructure and Mechanical Properties of Cast Magnesium-Thorium-Zirconium Alloy HK31A, a correlation, T. R. BERGSTROM and R. G. BASSETT, Boeing Airplane Co., Seattle, Wash.
Magnesium Casting Alloy EK31XA, K. E. NELSON, The Dow Chemical Co., Midland, Mich.

4:00 pm — Malleable Shop Course

Presiding — K. MOLTRECHT, University of Michigan, Ann Arbor.
Machining of Standard and Pearlitic Malleable Iron.
The Technical Aspects of Machinability Testing, L. C. MARSHALL, Link-Belt Research Laboratory, Indianapolis, Ind.
Availability of Machinability Data, W. C. TRUCKENMILLER, Albion Malleable Iron Co., Albion, Mich.
Solving Machining Problems, G. KRAMER, Central Foundry Div., General Motors Corp., Saginaw, Mich.

8:00 pm — Sand Shop Course

Presiding — B. H. BOOTH, Carpenter Bros., Inc., Milwaukee.
Air-set and CO_2 Binders.
CO_2, C. W. SUNDBERG, Minneapolis Electric Steel Castings, Minneapolis; F. P. ILENDA, Diamond Alkali Co., Painesville, Ohio.
Air-set, R. M. OVERSTRUD, Reichhold Chemicals, Inc., Eastern Resin Div., Elizabeth, N. J.

TUESDAY, APRIL 14

8:00 am — Author-Chairman Breakfast

9:30 am — Brass and Bronze Session

Presiding — H. F. BISHOP, Exomet, Inc., Conneaut, Ohio.
R. W. RUDDLE, Foundry Services, Inc., Columbus, Ohio.
Secretary — R. F. SCHMIDT, Ajax Metal Div., H. Kramer & Co., Chicago.
Pressure Tightness of 85-5-5-5 Bronze, relationship of thermal gradients during solidification, Research Progress Report, R. A. FLINN and C. R. MIELKE, University of Michigan, Ann Arbor.
Research in Progress in Industry:
Getting Castings Back into the Foundry, R. R. ASHLEY, American Radiator and Standard Sanitary Corp., Cincinnati.
Research and Development on Design of Bronze Sleeve Bearings, G. E. LANGFORD, Superior Kendrick Bearings, Inc., Detroit.
Properties and Composition of Cast Copper-Nickel Alloys, S. GOLDSPIEL, New York Naval Shipyard, Brooklyn, N. Y.
Research in 90-10 Cupro-Nickel Castings, E. F. TIBBETTS, Wollaston Brass & Aluminum Foundry, Inc., North Quincy, Mass.

9:30 am — Malleable Session

Presiding — W. M. ALBRECHT, Chain Belt Co., Milwaukee.
C. R. SORENSEN, National Malleable & Steel Castings Co., Cicero, Ill.
Secretary — P. W. GREEN, General Electric Co., Erie, Pa.
Specialized Foundry Statistical Controls Improve Customer Satisfaction, B. M. APPLEMAN, Texas Foundries, Lufkin, Texas.
Proper Gating through Use of Cobalt 60, A. J. KARAM, Central Foundry Div., General Motors Corp., Saginaw, Mich.
Fluorescent Magnetic Particle Method for Casting Inspection, proper interpretation and technique, A. R. LINDGREN, Magnaflux Corp., Chicago.

9:30 am — Pattern Session

Presiding — R. R. LEMASTER, R. A. Nelson Pattern Co., Milwaukee.
D. T. KINDT, The Kindt-Collins Co., Cleveland.
Secretary — R. L. SWANSON, Arrow Pattern & Foundry Co., Chicago.
Core Box Rigging for High Production, W. H. MILLER, Ford Motor Co., Cleveland Foundry Div., Berea, Ohio.
Consulting Service on Pattern and Core Box Equipment Problems.
Panel — M. K. YOUNG, U. S. Gypsum Co., Chicago; R. L. OLSON, Dike-O-Seal, Inc., Chicago; J. F. ROTH, Cleveland Standard Pattern Works, Cleveland; R. LEMASTER, R. A. Nelson Pattern Co., Milwaukee: W. E. MASON, Westinghouse Air Brake Co., Wilmerding, Pa.

9:30 am — Safety, Hygiene & Air Pollution Session

Presiding — K. M. SMITH, Caterpillar Tractor Co., Peoria, Ill.
R. L. MCILVAINE, National Engineering Co., Chicago.
Secretary — H. J. WEBER, American Foundrymen's Society.
Practical Planning for Non-destructive Testing by use of Radiation, F. S. SUTHERLAND, Continental Foundry & Machine Div., Blaw-Knox Co., East Chicago, Ind.
Audiometric Testing Programs in Foundry Noise Control, H. T. WALWORTH, Lumbermen's Mutual Casualty Co., Chicago.

12:00 Noon — Brass and Bronze Round Table Luncheon

Presiding — H. L. SMITH, Federated Metals Div., American Smelting & Refining Co., Whiting, Ind.
E. F. TIBBETTS, Wollaston Brass & Aluminum Foundry, Inc., North Quincy, Mass.
Why use Brass and Bronze Castings:
The Physical and Mechanical Properties of Cast Bronze Alloys, J. KURA, Battelle Memorial Institute, Columbus, Ohio.
Where one Automobile company would like to use Copper-Base Alloy Castings, H. C. TYREE, Electro-motive Div., General Motors Corp., LaGrange, Ill.
Cast Copper-Base Alloys for Corrosive Environments, C. L. BULOW, Bridgeport Brass Co., Bridgeport, Conn.

12:00 Noon — Pattern Round Table Luncheon

Presiding — J. M. KREINER, National Malleable & Steel Castings Co., Cleveland.
R. L. OLSON, Dike-O-Seal, Inc., Chicago.
Atomic Blast, a pattern for survival, F. B. PORZEL, Armour Research Foundation, Chicago.

12:00 Noon — AFS Board of Directors Luncheon and Business Meeting

Presiding — AFS President, LEWIS H. DURDIN

2:00 pm — Education Session

Presiding — B. L. BEVIS, East St. Louis Castings Co., East St. Louis, Ill.
R. W. SCHROEDER, University of Illinois, Chicago.
Secretary — J. TOTH, Harry W. Dietert Co., Detroit.
Shell Molding Developments in the Automotive Industry, H. C. GRANT, Ford Motor Co., Dearborn, Mich.

2:00 pm — Industrial Engineering & Cost Session

Presiding — M. T. SELL, Sterling Foundry Co., Wellington, Ohio.
E. C. REID, Ford Motor Co., Ltd., Windsor, Ont., Can.
The Industrial Engineer as a Potential Executive, M. E. MUNDEL, M. E. Mundel & Associates, Milwaukee.
Industrial Engineering for the Smaller Foundry, a practical approach, F. E. NOGGLE, Westover Corp., Milwaukee.

2:00 pm — Light Metals Session

Presiding — H. E. ELLIOTT, The Dow Chemical Co., Bay City, Mich.
K. B. BLY, Fabricast Div., General Motors Corp., Bedford, Ind.
Secretary — R. W. BRUNER, JR., American Smelting & Refining Co., Alton, Ill.
Hot Cracking Test for Light Metal Casting Alloys, E. J. GAMBER, Aluminum Co. of America, Cleveland.

Sample for Rapid Measurement of Gas in Aluminum, H. V. SULINSKI and S. LIPSON, Frankford Arsenal, Philadelphia.
Ultrasonic Attentuation in Cast Aluminum, H. SMOLEN and H. ROSENTHAL, Frankford Arsenal, Philadelphia.

2:00 pm — Malleable Session

Presiding — C. F. JOSEPH, Central Foundry Div., General Motors Corp., Saginaw, Mich.
L. E. EMERY, Marion Malleable Iron Works, Marion, Ind.
Secretary — D. L. LAMARCHE, JR., American Malleable Castings Co., Marion, Ohio.
Pinhole Occurrence in Malleable Castings, bonding clays as cause, D. R. JONES, Illinois Clay Products Co., Chicago and R. E. GRIM, University of Illinois, Urbana, Ill.
Porosity, Inclusions and Pinholes in Malleable Castings, C. A. SANDERS, American Colloid Co., Skokie, Ill.

4:00 pm — Gray Iron Shop Course

Presiding — H. H. WILDER, Vanadium Corp. of America, Chicago.
E. J. BURKE, Hanna Furnace Corp., Buffalo, N. Y.
W. W. LEVI, Lynchburg Foundry Co., Radford, Va.
How to Melt Several Irons in One Heat.
Panel — R. L. JOHNSON, Bucyrus-Erie Co., Erie, Pa.; M. D. NEPTUNE, James B. Clow & Sons, Inc., National Works, Birmingham, Ala.; K. G. PRESSER, Buckeye Foundry Co., Cincinnati.

4:00 pm — Malleable Shop Course

Presiding — C. F. JOSEPH, Central Foundry Div., General Motors Corp., Saginaw, Mich.
Effect of Melting Variables on the Properties of Malleable Irons, a Symposium, R. W. HEINE, University of Wisconsin, Madison.
Panel — W. D. MCMILLAN, International Harvester Co., Chicago; M. TILLEY, National Malleable & Steel Castings Co., Cleveland; E. WELANDER, John Deere Malleable Works, East Moline, Ill.

4:00 pm — Light Metals Session

Presiding — F. C. BENNETT, The Dow Chemical Co., Midland, Mich.
D. S. MILLS, Process Development Staff, General Motors Corp., Warren, Mich.
Secretary — J. G. MEZOFF, The Dow Chemical Co., Midland, Mich.
How Aircraft Designers Look at Light Metal Castings, A. R. MEAD, Grumman Aircraft Corp., Bethpage, Long Island, N. Y.
Bending Techniques for Evaluation of Cast Materials and Structures, a preliminary appraisal, J. C. GRADDY, Douglas Aircraft Co., Santa Monica, Calif.
Design Engineering as related to Magnesium Castings, G. H. FOUND, Arthur D. Little, Inc., Cambridge, Mass.

4:00 pm — Sand Session

Presiding — T. W. SEATON, American Silica Sand Co., Ottawa, Ill.
R. H. OLMSTED, Whitehead Bros., New York.
Secretary — J. G. SMILLIE, John Deere & Co., Moline, Ill.
How to Avoid Sand Segregation, report of Committee 8-F, W. D. CHADWICK, Manley Sand Co., Rockton, Ill.
Segregation of Sand During Handling, motion picture.
Ramming, Superheat and Alloys (Type of Metal) Effects on Metal Penetration, report of Committee 8-H, G. J. VINGAS, Magnet Cove Barium Corp., Des Plaines, Ill.

4:00 pm — Steel Session

Presiding — A. J. KIESLER, General Electric Co., Schenectady, N. Y.
W. A. KOPPI, International Nickel Co., New York.
Secretary — R. L. LORD, Wehr Steel Co., Milwaukee.
Interrelation between Stress Concentration and Castability, J. B. CAINE, Consultant, Cincinnati.
Aid for the Design Engineer, foundrymen! "accentuate the positive," J. J. HENRY, Missouri Steel Castings Co., Joplin, Mo.
A Method of Obtaining Castability and Maximum Field Service from Cast Products, J. W. BECKHAM, Texas Foundries, Inc.. Lufkin, Texas.

6:30 pm — Sand Dinner

Presiding — E. C. ZIRZOW, Werner G. Smith, Inc., Cleveland.
Let's Start Over, J. B. CAINE, Consultant, Cincinnati.

7:00 pm — Canadian Dinner

Presiding — J. H. KING, Toronto, Ont., Can.

WEDNESDAY, APRIL 15

8:00 am — Author-Chairman Breakfast

9:30 am — AFS Annual Business Meeting

Presiding — AFS President LEWIS H. DURDIN

President Durdin called upon General Manager Maloney who reported on the nomination of Officers and Directors for the coming year, and stated that no additional nominees had been received in accordance with the procedure prescribed in Art. XI of the Society By-Laws. He therefore cast a unanimous ballot of the membership of AFS for the election of the following:

President (to serve one year):
CHARLES E. NELSON, The Dow Chemical Co., Midland, Mich.
Vice-President (to serve one year):
N. J. DUNBECK, International Minerals & Chemical Corp., Skokie, Ill.
Directors (to serve three years):
N. N. AMRHEIN, Federal Malleable Co., West Allis, Wis.
R. E. MITTLESTEAD, Lectromelt Castings Div., Akron Standard Mold Co., Barberton, Ohio.
A. J. MOORE, Canadian Bronze Co., Ltd., Montreal, Que., Canada.
W. H. OLIVER, American Radiator & Standard Sanitary Corp., Buffalo, N. Y.
L. J. PEDICINI, Congress Die Casting Div., Tann Corp., Detroit.
J. N. WESSEL, Puget Sound Naval Shipyard, Bremerton, Wash.
D. L. COLWELL, Apex Smelting Co., Cleveland, was elected earlier to the AFS Board of Directors by the members of the board itself, in accordance with the bylaws.
Director (to serve one year):
L. H. DURDIN, Dixie Bronze Co., Birmingham, Ala.

The Society arranged to have the five first and second prize winners in the 1959 Robert F. Kennedy Apprentice Contest present at the Convention to receive their awards in person.

Wood Patternmaking

1st — HORST N. PROHASKA, Caterpillar Tractor Co., Peoria, Ill.
2nd — DARRELL A. GILES, Beaver Pattern Co., Detroit.
3rd — RAY BUNK, Rupp Pattern Co., Rockford, Ill.

Metal Patternmaking

1st — LOUIS S. UVEGES, Ford Motor Co., Berea, Ohio.
2nd — ZANE HOLOWACHUK, International Harvester Co., Hamilton, Ont., Can.
3rd — S. GREGORY MONOC, Ford Motor Co., Berea, Ohio.

Iron Molding

1st — FORREST R. WILLS, Link-Belt Co., Philadelphia.
2nd — ALFRED A. DIRAIMO, Brown & Sharpe, Providence, R. I.
3rd — WILLIAM GREGORY, W. O. Larson Foundry Co., Grafton, Ohio.

Steel Molding

1st — ROBERT LORD, Dominion Engineering Works, Montreal, Que., Can.
2nd — JAMES FAZIO, Atlas Foundry & Machine Co., Tacoma, Wash.
3rd — JAMES ROGERS, Oregon Steel Foundry Co., Portland, Ore.

Non-Ferrous Molding

1st — STANLEY GRAFF, Eck Foundries, Inc., Manitowoc, Wis.
2nd — RONALD MCCONVILLE, Nordberg Mfg. Co., Milwaukee.
3rd — RALPH LELAND, Alloy Aluminum Foundry Co., Los Angeles.

President Durdin then called upon Past-President Dietert to present the AFS Service Citations and Awards of Scientific Merit.

The AFS Service Citation

JAMES R. ALLAN, Allan Industries, Melbourne, Fla.
BERNARD D. CLAFFEY, Dayton Malleable Iron Co., Dayton, Ohio.
ROY W. SCHROEDER, University of Illinois, Chicago.

The Award of Scientific Merit

ROBERT H. MOONEY, Central Foundry Div., General Motors Corp., Saginaw, Mich.
HOWARD H. WILDER, Vanadium Corp. of America, Chicago.
ELMER C. ZIRZOW, Werner G. Smith, Inc., Cleveland.

10:30 am — Charles Edgar Hoyt Memorial Lecture

The Control of Quality in the Brass Foundry, HARRY M. ST. JOHN, Foundry Consultant, Fort Pierce, Fla.

12:00 Noon — Management Luncheon and Session

Presiding — C. E. WESTOVER, Westover Corp., Milwaukee.
So You Want to be a President.
Developing Technical Personnel, J. H. CULLING, Carondelet Foundry Co., St. Louis.
Developing Operating Supervisory Personnel, L. J. WOELKE, Grede Foundries, Inc., Milwaukee.
Developing Sales Personnel, H. A. FORSBERG, Blaw-Knox Co., East Chicago, Ind.
Developing Administrative Management, B. L. SIMPSON, National Engineering Co., Chicago.
Moderator — R. B. PARKER, American Brake Shoe Co., New York.
Panel Discussion.

12:00 Noon — Die Casting and Permanent Mold Luncheon

Presiding — D. L. COLWELL, Apex Smelting Co., Cleveland.
F. C. BENNETT, The Dow Chemical Co., Midland, Mich.
Vacuum Die Casting, G. R. MORTON, Morton Mfg. Co., Omaha, Nebr.

2:00 pm — Gray Iron Session

Presiding — S. G. JOHNSON, International Harvester Co., Indianapolis, Ind.
K. G. PRESSER, Buckeye Foundry Co., Moline, Ill.
Secretary — M. H. HORTON, John Deere & Co., Moline, Ill.
Continuous Carbon Injection, J. E. WILSON, and R. C. SHNAY, Canada Iron Foundries, Ltd., Montreal, Que., Can.
Gray Cast Iron Machinability, quantitative measurements of pearlite and graphite effects, W. W. MOORE, Battelle Memorial Institute, Columbus, Ohio and J. O. LORD, Ohio State University, Columbus, Ohio.
Significance of Reported Chemical Analysis of Cast Iron, D. E. KRAUSE, Gray Iron Research Institute, Columbus, Ohio.

2:00 pm — Die Casting and Permanent Mold Session

Presiding — H. E. ERIKSEN, Chrysler Casting Plant, Kokomo, Ind.
R. C. CORNELL, Litemetal Diecast, Inc., Jackson, Mich.
Secretary — A. SUGAR, Castmaster, Inc., Bedford, Ohio.
Aluminum Die Casting Metal Cost, its practical and analytical evaluation, G. H. FOUND, Arthur D. Little, Inc., Cambridge, Mass. and J. LAPIN, Central Foundry Div., General Motors Corp., Saginaw, Mich.
Cyclic Permanent Mold Operation, some thermal aspects, C. L. GOODWIN and H. Y. HUNSICKER, Aluminum Co. of America, Cleveland.
Gray Iron Permanent Molding, H. U. MCCLELLAND, Eaton Mfg. Co., Vassar, Mich.

2:00 pm — Steel Session

Presiding — D. N. ROSENBLATT, American Foundry and Machine Co., Salt Lake City, Utah.
S. L. GERTSMAN, Dept. of Mines and Technical Surveys, Ottawa, Ont., Can.
Secretary — E. W. O'BRIEN, Oklahoma Steel Castings Div., American Steel & Pump Corp., Tulsa, Okla.
Cast and Wrought Alloy Steels, elevated temperature properties, R. K. BUHR and W. A. MORGAN, Dept. of Mines and Technical Surveys, Ottawa, Ont., Canada.
Cast Low Alloy Steels Ductility and Toughness, reduced phosphorus and sulfur content significance, J. ZOTOS, Watertown Arsenal, Rodman Laboratory, Watertown, Mass.
Mn-V-Mo Age-Hardening Steel, foundry characteristics, N. C. HOWELLS and E. A. LANGE, U. S. Naval Research Laboratory, Washington, D. C.

4:00 pm — Die Casting and Permanent Mold Session

Presiding — W. N. BRAMMER, Pressure Castings, Inc., Cleveland.
C. B. CURTIS, The Maytag Co., Newton, Iowa.
Secretary — R. P. DUNN, Lindberg Engineering Co., Chicago.
Cast Surfaces Evaluation for Roughness Standards, E. SWING, Boeing Airplane Co., Wichita, Kans.
A Hypereutectic Aluminum-Silicon Alloy, R. KISSLING and O. TICHY, Apex Smelting Co., Cleveland.
New Aluminum Die Casting Alloy, J. H. MOORMAN and E. V. BLACKMUN, Aluminum Co. of America, Pittsburgh, Pa.

4:00 pm — Gray Iron Shop Course

Presiding — W. W. HOLDEN, Dostal Foundry & Machine Co., Pontiac, Mich.
D. E. MATTHIEU, Wysong & Miles Foundry, Inc., Greensboro, N. C.
K. H. PRIESTLEY, Vassar Electroloy Products, Vassar, Mich.
Correction of Cupola Irregularities.
Panel — W. W. LEVI, Lynchburg Foundry Co., Radford, Va.; M. H. HORTON, Deere & Co., Moline, Ill.; D. E. KRAUSE, Gray Iron Research Institute, Columbus, Ohio.

4:00 pm — Industrial Engineering and Cost Session

Presiding — M. O. JOHNSON, Gunite Foundries Corp., Rockford, Ill.
L. W. LEHMANN, John Deere Horicon Works, Horicon, Wis.
Profit Management through Cost Control, R. HILL, Dominion Bridge Co., Montreal, Que., Canada.
Time Study and Methods Training for Supervisors, J. TAYLOR, Norris & Elliott, Inc., Columbus, Ohio.

4:00 pm — Sand Session

Presiding — B. H. BOOTH, Carpenter Bros., Inc., Milwaukee.
W. R. MOGGRIDGE, Ford Motor Co., Windsor, Ont., Canada.
Secretary — G. DI SYLVESTRO, American Colloid Co., Skokie, Ill.
Green Tensile and Shear Strengths of Molding Sands, R. W. HEINE, University of Wisconsin, Madison; E. H. KING and J. S. SCHUMACHER, The Hill & Griffith Co., Cincinnati, Ohio.
Stickiness in Core Sand Mixtures, final report of AFS Committee 8-K, J. R. YOUNG, Cadillac Motor Car Div., General Motors Corp., Detroit.
Introduction of Tentative Hot Shell Deformation Test, report of AFS Committee 8-N, R. J. COWLES, Walworth Co., Braintree, Mass.

7:00 pm — AFS Annual Banquet

Presiding — AFS President LEWIS H. DURDIN

The AFS Annual Banquet was called to order by President Durdin. Presentation of AFS Gold Medal Awards was made as follows:

The John H. Whiting Gold Medal

Awarded to HAROLD W. LOWNIE, JR., Battelle Memorial Institute, Columbus, Ohio . . . *"For outstanding contributions to the Society and to metallurgical progress, especially in the field of Gray Cast Iron."*

The Peter L. Simpson Gold Medal

Awarded to JOHN A. RASSENFOSS, American Steel Foundries, East Chicago, Ind. . . . *"For exceptional contributions to the Society and the Steel Castings Industry, by elevating technical endeavor in steel foundry research."*

The John A. Penton Gold Medal

Awarded to FRED J. WALLS, retired from International Nickel Co., Detroit . . . *"For outstanding contributions to the Society and the Castings Industry, especially in the field of Gray Iron metallurgy and foundry practice, and for lifetime devotion to the industry's potentialities."*

Following the presentation of awards, President Durdin introduced the guest speaker of the evening WARREN WHITNEY, Vice Pres., James B. Clow & Sons, Inc., and General Manager of the

National Works, Birmingham, Ala. Mr. Whitney's subject was *"Solutions to Problems that Don't Exist."*

THURSDAY, APRIL 16

8:00 am — Author-Chairman Breakfast

9:30 am — Ductile Iron Session

Presiding — C. K. DONOHO, American Cast Iron Pipe Co., Birmingham.
A. W. ANDERSON, International Harvester Co., Chicago.

Secretary — L. J. VENNE, Electo Metallurgical Co., Div. of Union Carbide Corp., Cleveland.

Ductile Iron As-Cast and Annealed Tensile Properties, A. H. RAUCH, J. B. PECK and E. M. MCCULLOUGH, Deere & Co., Moline, Ill.

Ductile Iron Castings vs. Carbon Steel Forgings and Weldments, some comments and recent conversions, J. B. SALBAING, International Nickel Co., New York.

9:30 am — Fundamental Papers Session

Presiding — J. F. WALLACE, Case Institute of Technology, Cleveland.
C. F. WALTON, Gray Iron Founders' Society, Cleveland.

Secretary — G. K. TURNBULL, Case Institute of Technology, Cleveland.

Solidification Mechanisms of Eutectic and Gray Irons, A. DESY, University of Ghent, Belgium.

Grain Refinement of Solidifying Metals by Vibration, R. G. GARLICK and J. F. WALLACE, Case Institute of Technology, Cleveland.

Mass Effect on Castings Tensile Properties, a rationalization, P. J. AHEARN, Watertown Arsenal, Watertown, Mass., G. W. FORM and J. F. WALLACE, Case Institute of Technology, Cleveland.

9:30 am — Steel Session

Presiding — R. L. LORD, Wehr Steel Co., Milwaukee.
C. A. FAIST, Burnside Steel Foundry Co., Chicago.

Secretary — C. K. DONOHO, American Cast Iron Pipe Co., Birmingham, Ala.

Application of Press Forged Castings, P. R. GOUWENS, T. WATMOUGH and J. BERRY, Armour Research Foundation, Chicago.

Reclaim Sodium Silicate Bonded Sand, G. C. WARNEKE, National Malleable & Steel Castings Co., Melrose Park, Ill.

Rapid Hydrogen Determination for Steel Foundry Control, C. C. CARSON and B. J. ALPERIN, General Electric Co., Schenectady, N. Y.

10:00 am — National Casting Council Meeting and Luncheon

Presiding — G. L. SEVOY, Pres., Foundry Equipment Mfgrs. Assn.

12:00 Noon — AFS Past Presidents' Luncheon

Presiding — F. W. SHIPLEY, Caterpillar Tractor Co., Peoria.

12:00 Noon — Ductile and Gray Iron Round Table Luncheon

Presiding — H. H. WILDER, Vanadium Corp. of America, Chicago.
D. MATTER, Ohio Ferro-Alloys Corp., Canton, Ohio.

What the Casting Designer and Buyer Expect of the Foundryman, T. O. KUIVINEN, The Cooper-Bessemer Corp., Mt. Vernon, Ohio.

12:00 Noon — Steel Round Table Luncheon

Presiding — A. J. KIESLER, General Electric Co., Schenectady, N. Y.
E. A. ZEEB, Dodge Steel Co., Philadelphia.

Vacuum Arc Melting, S. J. NOESEN, General Electric Co., Schenectady, N. Y.

Deoxidation Defects in Steel Castings, final report for the AFS Steel Division research committee, R. A. FLINN and L. H. VANVLACK, University of Michigan, Ann Arbor.

2:00 pm — Ductile Iron Session

Presiding — E. WELANDER, John Deere Malleable Works, Moline, Ill.
S. S. PHILLIPS, Sibley Machine & Foundry Corp., South Bend, Ind.

Secretary — A. F. SPENGLER, JR., International Harvester Co., Chicago.

Basic Cupola Melting of Ductile Iron, J. T. WILLIAMS, Grede Foundries, Inc., Reedsburg, Wis.

Indirect Arc Electric Furnace Production of Ductile Iron, L. MILLER, Westinghouse Air Brake Co., Wilmerding, Pa.

Acid Cupola Melting for Ductile Iron, H. E. HENDERSON, Lynchburg Foundries, Lynchburg, Va.

Ductile Iron Production in Basic Direct-Arc Furnace, C. R. ISLEIB, International Nickel Co., New York.

2:00 pm — Fundamental Papers Session

Presiding — S. LIPSON, Frankford Arsenal, Philadelphia.
H. R. LARSON, American Brake Shoe Co., Mahwah, N. J.

Secretary — G. A. COLLIGAN, University of Michigan, Ann Arbor.

Application of Theory in Understanding Fluidity of Metals, J. E. NIESSE, M. C. FLEMINGS and H. F. TAYLOR, Massachusetts Institute of Technology, Cambridge.

Aluminum Alloys Fluidity Test, fluidity tripled with new mold coating, M. C. FLEMINGS, H. F. CONRAD and H. F. TAYLOR, Massachusetts Institute of Technology, Cambridge.

Linear Programming Applied to Economic Selection of Materials for Furnace Charges, G. I. GARTNER, Watertown Arsenal, Watertown, Mass.

Dimensioning of Sand Casting Risers, H. D. MERCHANT, Case Institute of Technology, Cleveland.

2:00 pm — Heat Transfer Session

Presiding — J. G. HENZEL, JR., General Electric Co., Schenectady, N. Y.
R. W. RUDDLE, Foundry Services, Inc., Columbus, Ohio.

Design of Castings.

Panel — D. J. ALBANESE, National Malleable & Steel Castings Co., Cleveland; A. BAIN, Canada Iron Foundries, Ltd., Trois Rivieres, Que., Can.; J. L. FLITZ, Central Foundry Div., General Motors Corp., Saginaw, Mich.; E. S. FRENS, General Electric Co., Schenectady, N. Y.; J. F. WALLACE, Case Institute of Technology, Cleveland.

2:00 pm — Sand Session

Presiding — R. L. DOELMAN, Miller & Co., Chicago.
J. A. GITZEN, Delta Oil Products Co., Milwaukee.

Secretary — L. D. MARINELLI, Kensington Steel Co., Chicago.

Tempering Molding Sand, H. W. DIETERT, V. ROWELL and A. L. GRAHAM, Harry W. Dietert Co., Detroit.

Crust Separation Test for Investigating Sand Expansion Defects, P. W. GOAD, The Institute of British Foundrymen, Australian Branch, Melbourne. *Presented by* L. E. TAYLOR, Ottawa Silica Co., Ottawa, Ill.

4:00 pm — Gray Iron Session

Presiding — J. S. VANICK, International Nickel Co., New York.
B. F. SHEPHERD, Ingersoll-Rand Co., Phillipsburg, N. J.

Secretary — D. KUHN, Valve & Mfg. Co., Indian Orchard, Mass.

Foundries can Produce their own Cast Iron directly from Ore, H. W. LOWNIE, JR. and A. J. STONE, Battelle Memorial Institute, Columbus, Ohio.

Radioisotopes Utilization in the Foundry Industry, M. POBERESKIN and D. N. SUNDERMAN, Battelle Memorial Institute, Columbus, Ohio.

Report of Joint AWS-AFS Committee on Welding Iron Castings:

Gray Iron, S. LOW, Chapman Valve Mfg. Co., Indian Orchard, Mass.

Ductile Iron, J. E. FITZWATER. International Harvester Co., Chicago.

Malleable Iron, S. T. WALTER, Air Reduction Sales Co., New York.

4:00 pm — Steel Session

Presiding — C. K. DONOHO, American Cast Iron Pipe Co., Birmingham, Ala.
E. W. O'BRIEN, Oklahoma Steel Castings Div., American Steel & Pump Corp., Tulsa, Okla.

Secretary — W. A. KOPPI, International Nickel Co., New York.

Steel Castings for the Aircraft Industry, quality requirements,

Y. J. ELIZONDO, Chance-Vought Aircraft Co., Huntington Park, Calif.

Design and Welding of Alloy Cast Steel, steam turbine applications, L. W. SONGER, General Electric Co., Schenectady, N. Y.

Foundry Designing for Steel Castings, A. B. STECK, Metallurgical Associates, Inc., Melrose, Mass.

6:00 pm — AFS Alumni Dinner

Presiding — HARRY W. DIETERT, *Past President.*

FRIDAY, APRIL 17

8:00 am — Author-Chairman Breakfast

9:30 am — Ductile Iron Session

Presiding — K. D. MILLIS, International Nickel Co., New York.
H. W. RUF, Grede Foundries, Inc., Milwaukee.

Secretary — G. KRUMLAUF, Republic Steel Corp., Cleveland.

Heat Treatment of Ductile Iron, W. D. MCMILLAN, International Harvester Co., Chicago.

Carbon Flotation in Ductile Iron, A. H. RAUCH, J. B. PECK and G. F. THOMAS, Deere & Co., Moline, Ill.

Tin Effect on Structure and Properties of Flake and Nodular Graphite Cast Irons, E. C. ELLWOOD, Tin Research Institute, London, England. *Presented by* J. A. DAVIS, Battelle Memorial Institute. Columbus, Ohio.

9:30 am — Fundamental Papers Session

Presiding — M. C. FLEMINGS, Massachusetts Institute of Technology, Cambridge.
P. J. AHEARN, Watertown Arsenal, Watertown, Mass.

Secretary — S. Z. URAM, Massachusetts Institute of Technology, Cambridge.

Development of Low Alloy Steel Compositions Suitable for High Strength Steel Castings, H. R. LARSON and F. B. HERLIHY, American Brake Shoe Co., Mahwah, N. J.

Comparison of X-Ray Quality and Tensile Properties in Cast High Strength Steel, H. R. LARSON, H. W. LLOYD and F. B. HERLIHY, American Brake Shoe Co., Mahwah, N. J.

Structural Variables Influencing Mechanical Properties of High Strength Cast Steels, a progress report, M. C. FLEMINGS, R. GREEN and H. F. TAYLOR, Massachusetts Institute of Technology, Cambridge.

9:30 am — Gray Iron Session

Presiding — J. VARGA, JR., Battelle Memorial Institute, Columbus, Ohio.
W. J. SOMMER, The Plainville Casting Co., Plainville, Conn.

Secretary — R. W. FREEMAN, Union Carbide Metals Co., Niagara Falls, N. Y.

Factors Influencing Soundness of Gray Iron Castings, a review of some recent British work, I. C. H. HUGHES, K. E. L. NICHOLAS, A. G. FULLER and T. J. SZAJDA, British Cast Iron Research Association, Birmingham, England.

Mold Material Effect on Cooling Rate and Physical Properties of Cast Metals, report of subcommittee T.S.46, the Institute of British Foundrymen. *Presented by* J. HIRD, Chairman, Birmingham, England, assisted by R. W. RUDDLE, Foundry Services, Inc., Columbus, Ohio.

Copper in Cast Iron, principal considerations, A. DESY, University of Ghent, Belgium.

9:30 am — Heat Transfer Session

Presiding — R. C. SHNAY, Canada Iron Foundries, Ltd., Toronto, Ont., Can.

Secretary — W. K. BOCK, National Malleable & Steel Castings Co., Cleveland.

Modern Quenching Oils and Techniques for Foundry Heat Treatment, N. F. SQUIRE, Aldridge Industrial Oils, Inc., Cleveland.

Solidification Times of Simple Shaped Castings in Sand Molds, J. T. BERRY, Armour Research Foundation, Chicago; G. MARTIN, Industrial Research Laboratories, Honolulu Oil Corp., Los Angeles; and V. KONDIC, University of Birmingham, England.

9:30 am — Sand Session

Presiding — T. E. BARLOW, International Minerals & Chemical Corp., Skokie, Ill.
E. C. ZIRZOW, Werner G. Smith, Inc., Cleveland.

Secretary — R. F. DALTON, American Colloid Co., Skokie, Ill.

Timing of Expansion Scab Formation, J. E. HALLER, James B. Clow & Sons, Coshocton, Ohio.

Sand Movement and Compaction in Green Sand Molding, T. J. BOSWORTH, R. W. HEINE, University of Wisconsin, Madison; J. J. PARKER, SPO, Inc., Cleveland; E. H. KING and J. S. SCHUMACHER, The Hill and Griffith Co., Cincinnati, Ohio.

2:00 pm — Gray Iron Session

Presiding — J. L. CLARK, Canada Iron Foundries, Ltd., St. Thomas, Ont., Can.
T. GISZCZAK, Central Foundry Div., General Motors Corp., Defiance, Ohio.

Secretary — N. D. LILLYBECK, Brillion Iron Works, Brillion, Wis.

Martensitic White Irons for Abrasion Resistant Castings, T. E. NORMAN, Denver, Colo.; D. V. DOANE and A. SOLOMON, Climax Molybdenum Co., Detroit.

Molybdenum Effect on Gray Iron Elevated Temperature Properties, G. K. TURNBULL and J. F. WALLACE, Case Institute of Technology, Cleveland.

Ductility and Strength of High-Carbon Gray Irons, E. M. STEIN and H. L. MCINTIRE, Battelle Memorial Institute, Columbus, Ohio.

2:00 pm — Plant and Plant Equipment Session

Presiding — J. THOMSON, East Chicago, Ind.
H. W. JOHNSON, Wells Mfg. Co., Skokie, Ill.

The Foundry Plant Engineer, a 50 year Reminiscence, J. THOMSON, East Chicago, Ind.

Foundry Plant Engineering, panel discussion.

4:00 pm — 63d AFS Castings Congress and Engineered Castings Show Officially Closes

ANNUAL REPORT OF AFS GENERAL MANAGER

(For Fiscal Year 1958-59)

This report covers overall Administrative and Policy matters and is supported by separate reports of the AFS Treasurer, Secretary and Technical Director.

SURVEY OF YEAR

The fiscal year 1958-59 was not a good year financially for AFS, producing an Excess Expense of $188,000 against a budgeted excess of $116,800. Thus the 2-year period 1957-1959 ended with excess Expense of $60,400.

When the 1958-59 budget was approved in August 1958 it contemplated a 2-year excess Income of $10,700. This margin proved far too small, in view of the continuing effects of the recession. Last February the General Manager put into effect a revised budget, with the approval of the Finance Committee, raising the excess expense for 1958-59 from $116,800 to $148,000. This also proved too conservative, due mainly to further losses in Dues Income, Convention and Exhibit Income, MODERN CASTINGS Income, and increased Burden and Chapter Expense, . . . to a total of $57,000.

On the basis of 1958-59 financial results, it is obvious that AFS budgets for the 2-year period 1959-1961 must incorporate additional Income or lesser Expense if the Budget is to balance. While castings business has materially improved in most quarters since March 1959, the problem is one of estimating future business conditions in the budgeting of AFS Income and Expense.

Membership

The Membership Analysis Committee is proposing strong effort to build up the Sustaining and Company totals. A major problem still concerns the number of delinquent members dropped: 1,498 during the past year, compared with 1,521 new members. This "Mexican Stand-Off" is attributed mainly to business conditions, which means that every effort should now be made to regain the losses of the past two years.

Chapters

The expense of Chapter operations, totaling $75,000, should be considered in terms of Membership returns. Unless Dues Income materially improves in the next two years, the expense of Chapter operations should be held to present or lower levels.

Convention & Exhibit

Following the partial success of the 1st Engineered Castings Show in Cincinnati, the 1959 Show was expected to materially exceed the 1957 event. Instead, space sales for the 1959 Show were less than 50% of the 1957 Show, and can be attributed to the combination of business conditions and reluctance of castings producers to accept the event as a valuable marketing medium. The Technical Program in 1959, slanted strongly toward the purposes of the Congress, was not supported by the exhibit in the degree expected.

The General Manager's report for the fiscal year 1957-58 mentioned recommendations of three exhibitor groups re frequency of AFS Exhibits. These recommendations will now be resolved in terms of AFS Finance Committee recommendations for the two coming fiscal years.

Education

The Apprentice Contest and Foundry Instructors Seminar were both highly successful. The latter produced attendance of 154 instructors in foundry and patternmaking, compared with the previous high of 110 in 1956.

While entries in the Apprentice Contest dropped off somewhat from the previous year, the 1959 contest was remarkably well supported. Nevertheless, many companies still do not understand that the contest is not limited strictly to apprentices but is based on years of molding and patternmaking experience.

Training & Research Institute

T&RI Courses continue to be well attended and during the year four Regional courses were held with local Chapter cooperation. Two or three courses were cancelled because of lack of attendance but students continue to come from many states to attend courses held in the Chicago area.

Under Board authority of last February, the Building Committee was appointed and has now finalized plans for the proposed Foundry Training Center Building. The architect is completing final revisions, with the next step the preparation of detailed specifications for general contract bids. Board action of last February delegated to the Executive Committee of the Board final decision to proceed with the building, based on general contract bids received.

All 1958-59 AFS financial obligations to the Institute were liquidated as of July 1. Further obligations will be based on budgets approved for the two coming years.

"Modern Castings"

While the printing of "Transactions" in MODERN CASTINGS has been approved in many quarters, it has necessitated reduction of non-Transactions editorial content in an effort to meet the Board's requirement that the magazine "be operated at a profit." As a result, it has been necessary to curtail some of the available material upon which total readership of the magazine must depend.

The problem facing AFS in operating MODERN CASTINGS is maintenance of an editorial-advertising ratio that will enable publication at a profit, as opposed to the implied obligation of AFS as a technical society to publish the maximum volume of worth-while material in the interests of all members. The latter cannot be done under the present Board policy of "publication at a profit," and any other policy is bound to increase the expense of producing MODERN CASTINGS.

SH&AP Program

Of the six basic engineering manuals originally scheduled for SH&AP production, five have been published (Noise Control during past year) and only one (Control of Radiation Hazards) remains to publish, being scheduled for 1959-60. The SH&AP Committees are now on a stand-by basis for the present. All manuals produced to date will require future revision as new information develops.

The importance of SH&AP work continues to grow, particularly in the field of assistance to foundry groups and individual plants in the prevention of legislation harmful to foundry interests.

"Buyers Directory"

Response to the first AFS BUYERS DIRECTORY indicates the advisability of continuing this project, with publication every two years. All member associations of the National Castings Council responded to the invitation to provide copy for 2-page spreads on their activities. The Directory is scheduled for distribution in October.

Many problems have had to be solved in developing this first Directory but we believe it will be of major value to the Castings Industry. In future Directories it will be necessary to develop special material, to make each succeeding volume of sufficient value to replace previous editions. Plans are now being developed to that end. Present estimates are that the first edition will produce Net Income of approximately $10,000 and the second edition should do better.

Respectfully submitted,

WM. W. MALONEY
General Manager

Breakdown of AFS Membership

As of June 30, 1959

Chapter	Sustaining	Company	Personal	Affiliate	Associate	Junior	Honorary	Total
Birmingham	9	15	114	253	7	6	9	413
British Columbia	0	5	80	9	7	0	0	101
Canton District	0	19	49	91	4	4	0	167
Central Illinois	2	7	31	164	3	4	2	213
Central Indiana	3	27	74	206	7	12	1	330
Central Michigan	1	8	43	97	17	13	2	181
Central New York	1	17	66	98	2	0	0	184
Central Ohio	2	20	53	116	6	3	4	204
Chesapeake	2	13	55	32	21	2	3	128
Chicago	18	57	192	386	12	64	7	736
Cincinnati District	7	37	89	147	6	9	3	298
Connecticut	3	13	65	38	3	1	0	123
Corn Belt	0	4	10	20	2	1	0	37
Detroit	7	36	149	288	16	14	10	520
Eastern Canada	5	25	101	170	17	10	2	330
Eastern New York	1	8	16	47	5	1	0	78
Metropolitan	9	39	140	182	19	4	6	399
Mexico	0	4	41	17	0	9	0	71
Michiana	3	18	70	156	5	0	0	252
Mid-South	0	0	17	22	0	0	0	39
Mo-Kan	1	4	42	28	5	2	0	82
New England	2	24	123	80	20	5	3	257
Northeastern Ohio	15	48	190	391	13	10	8	675
Northern Calif.	0	20	124	127	11	2	0	284
N. Ill. & So. Wis.	0	14	25	77	1	3	0	120
Northwestern Pa.	0	14	51	95	2	3	3	168
Ontario	2	56	153	209	5	10	0	435
Oregon	1	9	45	62	4	4	0	125
Philadelphia	6	49	140	172	20	6	5	398
Piedmont	1	14	54	49	13	2	4	137
Pittsburgh	5	22	169	133	8	1	2	340
Quad City	3	17	43	122	6	4	1	196
Rochester	0	3	40	26	2	0	0	71
Saginaw Valley	6	13	23	284	2	51	4	383
St. Louis District	1	18	128	100	10	3	0	260
Southern Calif.	1	38	188	137	9	4	0	377
Tennessee	0	10	57	196	2	4	0	269
Texas	1	24	107	112	8	7	1	260
Timberline	1	0	36	15	3	1	0	56
Toledo	1	13	32	93	4	8	1	152
Tri-State	0	6	58	51	1	11	1	128
Twin City	0	18	90	98	9	6	0	221
Utah	0	4	19	51	1	0	0	75
Washington	0	4	54	11	15	1	1	86
Western Michigan	1	22	62	179	1	1	1	267
Western New York	1	18	70	109	2	3	2	205
Wisconsin	9	61	133	443	8	8	4	666
TOTAL IN REGULAR CHAPTERS	131	915	3,711	5,989	344	317	90	11,497
STUDENT CHAPTERS								
Massachusetts Institute of Technology						11		11
Michigan State University						14		14
Missouri School of Mines					2	79		81
Ohio State University						13		13
Oregon State College						14		14
Pennsylvania State University						26		26
Polytechnic Institute of Brooklyn						23		23
Texas A. & M. College					1	29		30
University of Alabama						31		31
University of Illinois						97		97
University of Michigan						6		6
University of Wisconsin						25		25
Wentworth Institute						48		48
TOTAL IN STUDENT CHAPTERS					3	416		419
TOTAL IN ALL CHAPTERS	131	915	3,711	5,989	347	733	90	11,916
Foreign		20	408	6	32	16	2	484
Non-Chapter	1	3	16	3	9	2	0	34
In Military Service								15
GRAND TOTAL	132	938	4,135	5,998	388	751	92	12,449

ANNUAL REPORT OF AFS SECRETARY

(For Fiscal Year 1958-59)

This report covers the two categories of Chapters and Membership. A separate report on Administrative matters and Policy has been prepared by the General Manager.

Membership

The Society's membership on June 30, 1959 reached 12,449 or 3.4% below the total a year ago. Membership target of 13,700 again was not reached, although 16 Chapters made their targets compared with 11 the previous year, and 21 in 1956-57.

Our new policy qualifies any Chapter for a membership Target Certificate if they reach their target *at any time* during the fiscal year. Actually, at the close of the fiscal year only 6 Chapters were on or above target, with four Chapters only three or less off their targets.

The past year's decline in business prosperity is definitely reflected in the membership picture. During the past year many companies regulated their policies for economy purposes and no longer assumed the expense of employees for membership and/or Chapter meetings. This factor accounts for a considerable portion of the membership loss, coupled with layoffs in the industry, resulting in non-renewals of memberships.

During the past 12 months, 1,536 new members were added, 2,248 were dropped, and 429 resigned; 750 of the dropped delinquents were reinstated, giving a net loss of 431 members for the year.

Membership losses occurred in four classes: 8 Sustaining, 45 Company, 36 Personal, 354 Affiliate. Gains were shown in two classes: 13 Associate, 3 Junior. The large loss in Affiliates results from the loss of Company and Sustaining memberships. The relatively few Personal members lost is due to a number of conversions from Company to Personal.

Membership Program 1959-60. A total of 13,500 has been set for 1959-60, with concentrated effort on retention of current members and reinstatement of delinquents. The figure of 1,001 net gain in membership is possibly optimistic considering net losses of 460 and 431 in the past two years, but with improved business, growing activities of the Society and concerted effort on the part of Chapter membership Chairmen and the Central Office, it is possible to reach this goal.

Chapters

No new regular Chapters were established during the fiscal year 1958-59, but a new Student Chapter was installed at Wentworth Institute in Boston on April 23, 1959, with 48 student members. The enthusiasm and interest of the students, faculty, administration and industry is exceptionally outstanding. The Society can look forward to an active student program at the school.

The 16th Annual Chapter Officers Conference was held June 11-12 with total attendance of 112, including 96 Chapter delegates representing all 47 Chapters. The first day of the Conference was held at the AFS Central Office in Des Plaines, with four separate and rotating workshops covering Membership, Education, Technical Activities and MODERN CASTINGS. In addition to the President and Vice-President, 14 National Directors attended.

Respectfully submitted,
ASHLEY B. SINNETT
Secretary

ANNUAL REPORT OF AFS TREASURER

(Fiscal Year Ended June 30, 1959)

As a part of the Treasurer's report, the following documents were forwarded to the Board of Directors for study prior to the Annual Board Meeting, August 6-7:

1) Annual Audit Report as of June 30, 1959.
2) Comparative Statement of Itemized Income & Expense Budget vs. Actuals, for years 1955-56 through 1958-59.
3) Budget Statements of Itemized Income & Expense – 2 years, 1959-1961. (Included in Comparative Statement.)

1958-59 SUMMARY ANALYSIS

Comparative Statement — Actual vs. Budget

Income Items		
Membership Dues	$ 8,800.	under
"MODERN CASTINGS"	33,400.	under
Publications (all)	8,300.	under
Convention & Exhibit	58,100.	under
"BUYERS DIRECTORY"	34,000.	under
Income from Investments	6,500.	under
Misc. Income	2,000.	over
Total Income "under" Budget	$134,000.	
Expense Items		
Salaries	$ 200.	over
Burdens	10,700.	over
Publications (all)	7,000.	under
"MODERN CASTINGS"	20,900.	under
Convention & Exhibit	9,000.	under
Chapter Expense	900.	under
Technical Expense (all)	3,500.	under
T&RI	19,100.	under
"BUYERS DIRECTORY"	13,500.	under
Total Expense "under" Budget	$ 63,000.	
Net Expense in excess of Budgeted Net Expense	$ 71,000.	
Net Expense as Budgeted	$116,800.	

Conclusion

Visible evidence is available in financial statements for 1958-59 to show how reduced Income activity affects Technical Activities of the Society, both in development of the theory for progress and distribution factors necessary to put available data into practice. For some years our supporting membership has been growing smaller and that portion "waiting for service" is larger.

Income & Expense Budgets 1959-61

Prior 2-year statements of Actual vs. Budget having been reviewed by the Finance Committee, Income & Expense budgets for 1959-1961 now are being recommended for Board approval, subject to review in July 1960 for the year 1960-61. The proposed 2-year Budget for 1959-1961 may be compared with Actual Excess Income or (Expense) of previous 2-year periods as follows:

2-Year Period	Excess Income or (Expense)
1951-1953 Actual	$126,040.
1953-1955 Actual	47,889.
1955-1957 Actual	41,963.
1957-1959 Actual	(60,388.)
1959-1961 Budget	1,000.
5-Period Average	$ 31,100.

Investments as of June 30, 1959

The Auditor's report summarizes Income from Investments held in the Society's Investment Agent Trust account at the Harris Trust & Savings Bank, Chicago. The report shows the following sources:

	1958-59	1957-58
Interest on U. S. Govt. Bonds	$ 331.07	$ 882.00
Interest on Canadian Govt. Bonds	831.25	831.25
Interest on Industrial Bonds	7,702.29	7,708.82
Dividends on Common Stocks	9,282.46	8,504.68
Discount on Treasury Bills	(170.67)	2,031.03
Gain (Loss) on Securities Sold	8,764.99	(1,873.65)
Write Down of Bonds to Maturity Value	(325.00)	
Interest on Bank Loans	(874.58)	
TOTAL	$25,541.81	$18,084.13

	Book Value Cost	Mkt. Value June 30	% Incr. or (Decr.)	% of Total (Mkt.)
U. S. Govt. Bonds	$ 21,924.	$ 21,010.	(4.16)	3.75
Canadian Govt. Bonds	24,827.	22,838.	(8.01)	4.08
Industrial Bonds	240,185.	213,312.	(11.18)	38.11
Common Stocks	161,096.	302,609.	87.84	54.06
TOTAL	$448,032.	$559,769.	24.93	100.00

Thus, although the Society incurred Net Operating Loss of $187,800 or $71,000 more than the Net Loss budgeted, AFS finances were still sound as of June 30 with an invested reserve fund totaling $559,769 market value. The invested reserve is maintained in an Investment Agent Trust and accumulated interest and dividends normally are "ploughed back" and reinvested.

This Agent Trust was created in December 1952 with an original principal of $80,000 Cash and $220,600 in Securities. Since then periodic deposits of surplus cash totaling $75,000 have been made to the Trust account for investment, exclusive of earned interest and dividends. This principal, as invested and reinvested within the Trust account thus has been appreciated 49.03% to $559,769 (market value) in six years.

Member Income and Expense

In calculating the Society's financial ability to increase or expand its services, the common denominator of Income and Expense per Member must be used. If net Expense per Member consistently exceeds realized Dues Income per Member, the Society cannot maintain a scale of expanding activities anymore than a foundry can consistently operate in the red. The following two tables show how average Dues Income per Member has *failed consistently to equal Excess Expense* per Member.

Avg. Income vs. Avg. Expense

8 Years, 1951-52 to 1958-59, Inclusive

	Total Members June 30	Avg. Dues Income per Member	Avg. Net Expense per Member	Avg. Excess (Expense) per Member
1958-59	12,449	$22.99	$31.11	$8.12
1957-58	12,880	23.15	29.00	5.85
1956-57	13,340	23.11	25.20	2.09
1955-56	12,488	23.79	25.65	1.86
1954-55	10,965	24.50	29.98	5.48
1953-54	11,551	20.01	22.83	2.82
1952-53	11,033	21.92	26.34	4.42
1951-52	10,066	21.41	24.26	2.85
8-Yr. Avg.	11,846	$22.61	$26.80	$4.19
8-Yr. Total Excess Expense				$49,634.

2-Year Avg. Dues Income Per Member vs. 2-Year Avg. Expense

8 Years, 1951-53 to 1957-59, Inclusive

2-Year Period	Avg. Dues Income per Member	Avg. Net Expense per Member	Avg. Excess (Expense) per Member
1957-1959	$23.07	$30.03	$6.96
1955-1957	23.44	25.42	1.98
1953-1955	22.20	26.30	4.10
1951-1953	21.68	25.35	3.67
4-Period Avg.	$11.30	$13.39	$2.09

From the above table it will be noted that Avg. Dues Income, Avg. Expense and Avg. Excess Expense per Member followed a precise 2-year pattern through 1956-57. Since then, however, Avg. Dues Income has *steadily declined* from $23.11 to $22.99 per Member. At the same time, Avg. Expense has *increased steadily and sharply* from $25.20 to $31.11, and Avg. Excess Expense per Member zoomed from $2.09 to $8.12. Result: servicing and average membership of 11,846 over the past 8 years cost the Society $49,634 more than was received in Member Dues.

AFS is by no means unique in facing this rather universal association problem. Obviously, however, the breadth and extent of Society services must be financed by income sources other than member dues. As will be noted in the following table, AFS must depend on three major sources of Net Income to maintain those services essential to foundry progress:

Condensed Operating Statement of Income & Expense Fiscal Year 1958-59

(All Expenses distributed to Major Activities—as Audited)

Net Income Activities	Net Income Amount	% of Total
Membership Services	$196,604.	92.4
"MODERN CASTINGS"	(575.)	(0.3)
Exhibit	(11,397.)	(5.4)
Investments & Misc.	28,075.	13.3
TOTAL	$212,707.	100.0

Net Expense Activities	Net Expense Amount	% of Total
Technical Activities	$120,394.	30.1
Publications	44,274.	11.0
Chapter Operations	93,731.	23.4
Convention	34,508.	8.6
"BUYERS DIRECTORY"	13,213.	3.3
General Administration	94,455.	23.6
TOTAL	$400,575.	100.0
Net Income or (Expense), Year	($187,868.)	

Because of the biennial nature of certain AFS revenues, the Society in 1954 adopted a policy of budgeting Income and Expense by 2-year periods. The table, below, compares Net Income and Net Expense activities for the past three 2-year periods. It vividly illustrates the significant fact that AFS Net Income in recent years has not been increasing sufficiently to enable a broadening of those essential services (e.g., fundamental research) which the Society is expected to perform.

In fact, the table reveals the rather startling fact that after six full years of operation, during which the Society's gross Income exceeded $4,680,000, the net result was *an earned Excess Income of only $8,431.*

Condensed Operating Statement of Income & Expense 2-Yr. Periods 1953-55, 1955-57, 1957-59

(All Expenses distributed to Major Activities—as Audited)

Net Incomes Activities	1953-1955	1955-1957	1957-1959
Membership Services	$387,246.	$434,166.	$403,880.
"MODERN CASTINGS"	24,206.	(12,336.)	14,470.
Exhibits	237,838.	228,510.	256,820.
Investments and Misc.	27,248.	48,277.	47,420.
TOTAL	$676,538.	$698,617.	$722,590.
Net Expense Activities		**Net Expense**	
Technical Activities[1]	$172,786.	$232,874.	$246,539.
Publications	68,635.	43,109.	76,219.
Chapter Operations	164,275.	190,143.	195,301.
Conventions	62,782.	57,034.	72,136.
"BUYERS DIRECTORY"[2]	—	—	22,026.
General Administration	181,203.	133,494.	170,758.
TOTAL	$649,681.	$656,654.	$782,979.
2-Yr. Excess Income (Expense)	$ 26,857.	$ 41,963.	$ (60,389.)

NOTES: 1) Includes Research, Contributions to Training & Research Institute, Safety Hygiene & Air Pollution Control program.

2) Initiated 1958, no major income until 1959-60 when completed and published.

10-Year Consistent Progress

In spite of the "tightrope financing" necessary for the past 20 years, AFS has constantly added new services in line with the industry's growing needs for information and assistance. At the same time, the Society's net worth has been steadily increased, as shown in the following table covering the 10-year period from July 1949 through June 1959:

Assets Increases — 10 Years

1949-1959, inclusive

As of June 30	(At Cost) Investments	Fixed Assets	Inventories	Fund Principals
1959	$449,180.	$341,594.	$44,605.	$774,225.
1958	465,298.	351,028.	47,974.	965,710.
1957	476,848.	301,230.	57,067.	838,349.
1956	560,960.	307,087.	52,218.	908,625.
1955	456,904.	304,490.	52,323.	801,021.
1954	564,573.	13,046.	37,135.	912,455.
1953	448,761.	14,616.	39,668.	679,974.
1952	330,600.	15,560.	32,470.	622,775.
1951	308,806.	14,765.	27,818.	406,812.
1950	118,806.	14,680.	35,355.	320,701.
10-Year Increase	$330,374.	$326,914.	$ 9,250.	$453,524.

General Observations

The following illustrates how AFS services to the Foundry Industry have been expanded in a 2-year period of reduced income (1957-1959):

Expanded Activities

	2-year Cost 1957-1959
Convention Papers, "MODERN CASTINGS" (Distribution increased 15,000)	$ 30,000.00
AFS News, "MC", increased coverage	16,000.00
Engineered Castings Show	12,000.00
Foundry Instructors Seminar	10,000.00

Invested for Future Progress

T&RI Land, Development and Fees	$ 55,000.00
"BUYERS DIRECTORY"	22,000.00
	$145,000.00
Thereby necessitating:	
Increased Liability	$102,000.00
Decreased Fund Principal	60,000.00
	$162,000.00

It can readily be seen from the above that reduced activity must be in order until ways and means are found to underwrite expanded and expanding activities. Last year, expansion of AFS service to the industry through continuation of the Engineered Castings Show and the new BUYERS DIRECTORY was unanimously approved in the hopes that exhibit income would carry the added expense. The BUYERS DIRECTORY will carry its own expense, but in a period of reduced income activity AFS must develop more "supporting" memberships (Company and Sustaining) to finance the Engineered Castings Show and educational activities. Ideas for increasing service are now temporarily withheld pending increased participation of the industry in sharing the expense of progressive and promotional efforts now available.

Respectfully submitted,
EDW. R. MAY
Treasurer

Comparative Condensed Balance Sheets

June 1956 to June 1959, inclusive (as Audited)

Assets	June 30 1956	June 30 1957	June 30 1958	June 30 1959	Incr. or (Decr.) Over 1958	4-Year Incr. or (Decr.)
Cash	$ 15,604.	$ 20,710.	$ 127,573.	$ 23,858.	$ (103,715.)	$ 8,254.
Investment Securities	560,960.	476,848.	465,298.	483,180.	17,882.	(77,780.)
Inventories	52,218.	57,067.	47,974.	44,605.	(3,369.)	(7,613.)
Accounts Receivable	10,284.	13,336.	13,647.	10,459.	(3,188.)	175.
Deferred & Prepaid Items	8,103.	13,141.	18,928.	16,885.	(2,043.)	8,782.
Furniture & Fixtures (net)	33,985.	33,326.	34,500.	30,651.	(3,849.)	(3,334.)
Land & Buildings	273,102.	267,905.	316,528.	310,943.	(5,585.)	37,841.
Total	$954,256.	$882,333.	$1,024,448.	$920,581.	$ (103,867.)	$ (33,675.)
Liabilities						
Current Liabilities	$ 45,631.	$ 43,984.	$ 58,739.	$146,356.	$ 87,617.	$ 100,725.
Fund Principals	908,625.	838,349.	965,709.	774,225.	(191,484.)	(134,400.)
Total	$954,256.	$882,333.	$1,024,448.	$920,581.	$ (103,867.)	$ (33,675.)

ANNUAL REPORT OF AFS TECHNICAL DIRECTOR

(Fiscal Year July 1, 1958 - June 30, 1959)

Convention Program

At the 63d Castings Congress held in Chicago, 102 papers were submitted for consideration and 99 approved for presentation. The program comprised 43 Technical Sessions, 9 Round-Table Luncheons, 5 Shop Courses and 4 Dinner meetings, including the Annual Banquet. Particular mention should be made of the excellent papers developed by the Steel Division. Interest in the technical papers was excellent, as evidenced by attendance at the sessions.

The Annual Lecture was delivered by Harry M. St. John on the subject "Control of Quality in the Brass Foundry."

For the second year, all papers received prior to the established deadline date were published in the "Transactions" section of MODERN CASTINGS and made available to 18,000 readers. From the marked increase in written discussions as well as oral participation following the presentation of the papers, it is evident that broad distribution of these papers has increased interest.

During 1958, 64 pages of technical papers were published monthly in MODERN CASTINGS. In January 1960, the annual TRANSACTIONS bound volume will be ready for distribution and will include all Convention papers together with written and oral discussion. This method of publishing the Convention pa-

pers should add materially to the member services rendered by the Society.

Apprentice Contest

The Annual Apprentice Contest met with good response in 1959, with a total of 474 entries from 189 companies. Entries were not as numerous as in 1958, probably reflecting general business conditions. From 15 local Chapter elimination contests and many individual plant contests, a total of 116 entries were judged in the national competition in the fields of gray iron, nonferrous and steel castings, and wood and metal patterns.

Publications Program

The 1958 annual TRANSACTIONS of the Society, vol. 66, were shipped in December 1958 to all pre-publication orders, and gratis copies were sent to all Company, Sustaining and Honorary Life members on request. A total of 2800 copies were printed—1224 sold, 626 gratis copies shipped, leaving a stock of 950 for future requirements.

During 1958-59 sales of Special Publications totaled $32,348, and income from royalties $1146. Sales during the year were somewhat less than in 1957-58, attributable to business conditions and the fact that practically no new publications were produced during the year.

The following books were published and made available for sale during the year: MELTING VARIABLE EFFECTS ON MALLEABLE IRON PROPERTIES; Tentative Standard Pattern Color Chart for New Patterns; FOUNDRY NOISE CONTROL MANUAL. In addition, the following publications were reprinted: STATISTICAL QUALITY CONTROL FOR FOUNDRIES; GLOSSARY OF FOUNDRY TERMS; FOUNDRY SAFETY MANUAL.

A revised edition of the PATTERNMAKER'S MANUAL is in process. A book on Radiation Hazards is in final stage of editing, for production in the near future. The MANUAL OF FOUNDRY REFRACTORIES has been completed for final review by the Refractories Committee. Two books — MOLDING METHODS & MATERIALS and the FOUNDRY SAND HANDBOOK — are in process of preparation by committees, and both should be available in 1959-60.

Research Projects

Research projects, under direction of AFS divisional research committees and sponsored by the Training and Research Institute, are making excellent progress. Reports on several projects were presented at the 1959 Convention. Since research develops new information, these reports are an excellent source of interest for the membership and provide fine material for local chapter meeting presentation.

Technical Committees

Nearly 675 men now serve on the 118 technical committees of the Society, in many instances on more than one committee. The two new divisions authorized by the Board of Directors last year have made an excellent start. Both the Die Casting & Permanent Mold and the Ductile Iron divisions have organized a number of committees and show evidence of real interest and activity.

AFS Training and Research Institute

The Board of Directors has been kept informed of the extensive training program being conducted by the Training and Research Institute. During 1958-59 a total of 15 courses were given on the following subjects: Gating & Risering of Castings, Cupola Melting of Iron, Melting of Copper-Base Alloys, Metallurgy of Ferrous Metals, Metallurgy of Gray Iron, Sand Control & Technology, Patternmaking, Product Development and Foundry Plant Layout.

Nine additional courses are scheduled in 1959 in the fields of Core Sand Practice, Patternmaking, Gating & Risering of Castings, Product Development & Marketing, Cutting Costs by Safe Practices, Preventive Maintenance, Permanent Mold & Die Casting, Sand Control & Technology, and an advanced course in Industrial Engineering. Three courses have been jointly sponsored with local chapters at Los Angeles and at Berkeley, Calif., and at Hamilton, Ont.

Since inception of the Training and Research Institute program, a total of 1006 students have registered from approximately 435 companies, and 37 per cent have sent "repeat" students. This is strong evidence that value must have been received by these organizations.

In accordance with Board decisions, a Building Committee and the Staff are energetically working with the architects in preparing detailed drawings and specifications for the proposed T&RI training building.

On June 18-20, an extremely successful Foundry Instructors Seminar, with 175 participants, was sponsored by the AFS Training and Research Institute in cooperation with the University of Illinois at Urbana. The FOUNDRY INSTRUCTORS GUIDE was completed and given to each instructor attending. This publication should prove of real value to instructors because of the large number and variety of casting projects described in the manual, and should be particularly helpful to teachers initiating a foundry program.

Safety, Hygiene & Air Pollution

The Director is cooperating with the Northern California Chapter on proposed Bay Area air pollution law. As chairman of the Committee on Foundries of the Air Pollution Control Association, he developed five technical reports for governmental control officers. Consultation was given two member foundries on air pollution and ventilation problems and, in addition, he has corresponded with numerous member foundries on problems in this field. The Director also served as an instructor on three T&RI courses and addressed nine chapters and regional conferences.

In addition to his specific activities for the SH&AP program, the Director has been serving as Acting Assistant Technical Director, representing the Technical Director at numerous technical committee meetings.

Technical Inquiries

The Central Office continues to receive an increasing number of technical inquiries from members of the Society and others on a broad range of subjects. Although considerable time and effort are spent in preparing suitable replies, this undoubtedly is an excellent means by which the Society is providing a valuable service to the industry as a whole.

Respectfully submitted,

S. C. MASSARI
Technical Director

Minutes

First Meeting of AFS Board of Directors 1958-59

Hotel Statler, Cleveland — May 24, 1958

ROLL CALL: President L. H. Durdin, *presiding*
Vice-President C. E. Nelson

Directors (Exp. 1959)
Chas. E. Drury
Roger W. Griswold
Al V. Martens
Alex W. Pirrie
Gerald R. Rusk

Directors (Exp. 1960)
Wm. D. Dunn
A. A. Hochrein
Karl L. Landgrebe, Jr.
Fred J. Pfarr
John R. Russo
Allen M. Slichter
Henry G. Stenberg

Directors (Exp. 1961)
David W. Boyd
Thomas W. Curry
Richard R. Deas, Jr.
Jake Dee
Webb L. Kammerer
Hillard M. Patton
Clyde A. Sanders

Wm. W. Maloney, *General Manager-Secretary*
S. C. Massari, *Technical Director*
Edw. R. May, *Treasurer*
Ashley B. Sinnett, *Asst. Secretary*

Absent: Directors H. Heaton (exp. 1959), G. P. Phillips (exp. 1959).

A quorum having been established, President Durdin called to order the first meeting of the 1958-59 Board of Directors and invited outgoing Directors to remain and participate in the discussions. He then introduced the new Vice-President C. E. Nelson and each of the newly elected Directors.

Election of Regional Vice-Presidents

The President announced that by letter ballot of the new Board the following Regional Vice-Presidents were declared elected for 1958-59:

Region 1 — H. G. Stenberg
Region 2 — R. W. Griswold
Region 3 — F. J. Pfarr
Region 4 — C. E. Drury
Region 5 — A. M. Slichter
Region 6 — K. L. Landgrebe, Jr.
Region 7 — J. R. Russo

The President requested Vice-President Nelson to organize Regional Administration Meetings in all Regions during the year, to be attended by all Directors in their respective Regions, with invitations to Chapter Chairmen and Vice-Chairmen. The President requested verification of dates of all RAM meetings to be held.

Directors Slichter and Russo stated they would issue all invitations and make all arrangements for RAM meetings in their territories. The Central Office was requested to handle similar details for the other five Regions.

Formation of Executive Committee

The President stated that under the new By-Laws all Regional Vice-Presidents serve as members of the Executive Committee, together with the President, Vice-President and immediate past-President.

Appointment of Board Committees

President Durdin stated that he was empowered under the By-Laws to appoint Board committees for the fiscal year 1958-59. On motion duly made, seconded and carried, appointment of the following bodies (where called for) was approved:

Regional Vice-Presidents

Region 1 — H. G. Stenberg
Region 2 — R. W. Griswold
Region 3 — F. J. Pfarr
Region 4 — C. E. Drury
Region 5 — A. M. Slichter
Region 6 — K. L. Landgrebe, Jr.
Region 7 — J. R. Russo

Executive Committee

L. H. Durdin, *Chairman*
H. W. Dietert
C. E. Drury
R. W. Griswold
K. L. Landgrebe, Jr.
C. E. Nelson
F. J. Pfarr
J. R. Russo
A. M. Slichter
H. G. Stenberg

Finance Committee

L. H. Durdin, *Chairman*
C. E. Nelson
H. W. Dietert
Wm. W. Maloney
E. R. May

Board of Awards

W. L. Seelbach, *Chairman* (1951-52)
I. R. Wagner (1952-53)
Collins L. Carter (1953-54)
Frank J. Dost (1954-55)
B. L. Simpson (1955-56)
F. W. Shipley (1956-57)
H. W. Dietert (1957-58)

Nominating Committee

H. W. Dietert, *Chairman*
F. W. Shipley
(6 others to be selected by Executive Committee)

Chapter Contacts Committee

C. E. Nelson, *Chairman*
(All Regional Vice-Presidents)
(All Directors, Members)

T&RI Trustees

H. Bornstein, *Chairman* (exp. 1961)
H. W. Dietert (exp. 1959)
I. R. Wagner (exp. 1959)
L. H. Durdin (exp. 1960)
B. C. Yearley (exp. 1960)
C. E. Nelson (exp. 1961)
R. A. Oster (exp. 1962)

Retirement Fund Trustees

R. S. Hammond, *Chairman*
Collins L. Carter
Ralph L. Lee
F. W. Shipley

Annual Lecture Committee

B. L. Simpson, *Chairman*
M. E. Brooks
A. L. Hunt
H. H. Wilder
V. E. Zang

Publications Committee

D. C. Colwell, *Chairman*
T. E. Barlow
T. T. Lloyd
F. L. Riddell
C. E. Sims
H. H. Wilder

Self-Appraisal Committee

L. H. Durdin, *Chairman*
C. E. Nelson
(4 Members from Industry, to be appointed by President)

Membership Analysis Committee

R. W. Griswold, *Chairman*
C. E. Drury
A. A. Hochrein
A. V. Martens
C. A. Sanders
F. J. Pfarr
T. W. Curry

Exhibits Analysis Committee

C. A. Sanders, *Chairman*
R. R. Deas, Jr.
R. W. Griswold
K. L. Landgrebe, Jr.
H. M. Patton
G. R. Rusk

Foundry Trust Committee

L. H. Durdin, *Chairman*
D. W. Boyd
W. L. Kammerer
A. V. Martens
G. R. Rusk
A. M. Slichter
Jake Dee

Board Nominating Committee

C. E. Nelson, *Chairman*
A. M. Slichter
H. G. Stenberg

National Castings Council Representatives

L. H. Durdin
C. E. Nelson

Foundry Educational Foundation Trustees

A. V. Martens (1957-1959)
W. D. Dunn (1958-1960)

International Representatives

L. H. Durdin
Wm. W. Maloney
Dr. A. B. Everest, European Repr.
Wm. A. Gibson, Australian Repr.

N.C.C. Representatives

The President stated that AFS, as a charter member of the National Castings Council, was obliged to name two official representatives, one of whom must be the incumbent President. On motion duly made, seconded and carried, Vice-President Nelson was named in addition to President Durdin.

Organization for Chapter Contacts

President Durdin stated that, while assignments of Directors for Chapter Contacts would be made as in previous years, he felt that such contacts should be on a voluntary basis. He expressed the hope that all contacts would be made if convenient, but stressed the importance of full Director attendance at Regional Administration Meetings.

The following schedule of proposed Chapter Contacts then was presented and accepted as "voluntary assignments" by all Directors present:

Chapter Contact Assignments — 1958-59

Region No. 1

Regional Vice-President: HENRY G. STENBERG

Chapter Group A
- Brooklyn Polytechnic Institute (A. A. Hochrein)
- Connecticut (H. G. Stenberg)
- Massachusetts Institute of Technology (H. G. Stenberg)
- Metropolitan (A. A. Hochrein)
- New England (H. G. Stenberg)

Chapter Group B
- Chesapeake (A. A. Hochrein)
- Philadelphia (T. W. Curry)
- Piedmont (T. W. Curry)

Region No. 2

Regional Vice-President: ROGER W. GRISWOLD

Chapter Group C
- Central New York (Wm. D. Dunn)
- Eastern New York (Wm. D. Dunn)
- Rochester (Wm. D. Dunn)

Chapter Group D
- Eastern Canada (To be appointed)
- Ontario (To be appointed)

Chapter Group E
Northwestern Pennsylvania (R. W. Griswold)
Pennsylvania State University (R. W. Griswold)
Pittsburgh (R. W. Griswold)
Western New York (R. W. Griswold)

Region No. 3

Regional Vice-President: FRED J. PFARR

Chapter Group F
Northeastern Ohio (F. J. Pfarr)

Chapter Group G
Canton District (F. J. Pfarr)
Central Ohio (G. R. Rusk)
Ohio State University (G. R. Rusk)
Toledo (G. R. Rusk)

Region No. 4

Regional Vice-President: CHARLES E. DRURY

Chapter Group H
Detroit (H. W. Dietert)
Saginaw Valley (C. E. Nelson)
University of Michigan (C. E. Nelson)

Chapter Group I
Central Michigan (D. W. Boyd)
Michiana (C. E. Drury)
Michigan State University (D. W. Boyd)
Western Michigan (D. W. Boyd)

Chapter Group J
Central Indiana (C. E. Drury)
Cincinnati District (R. R. Deas, Jr.)

Region No. 5

Regional Vice-President: ALLEN M. SLICHTER

Chapter Group K
Chicago (C. A. Sanders; G. P. Phillips)

Chapter Group L
University of Wisconsin (A. M. Slichter)
Wisconsin (A. M. Slichter)

Chapter Group M
Central Illinois (A. V. Martens)
N. Illinois-S. Wisconsin (A. M. Slichter)
Quad City (A. V. Martens)
Twin City (H. M. Patton)
University of Illinois (C. A. Sanders)

Region No. 6

Regional Vice-President: KARL L. LANDGREBE, JR.

Chapter Group N
Corn Belt (W. L. Kammerer)
Mo-Kan (W. L. Kammerer)
St. Louis District (W. L. Kammerer)
Timberline (Jake Dee)
Univ. of Missouri, School of Mines & Met. (W. L. Kammerer)

Chapter Group O
Birmingham District (L. H. Durdin)
Mid-South (K. L. Landgrebe)
Tennessee (K. L. Landgrebe)
University of Alabama (L. H. Durdin)

Chapter Group P
Mexico (To be assigned)
Texas (Jake Dee)
Texas A. & M. College (Jake Dee)
Tri-State (Jake Dee)

Region No. 7

Regional Vice-President: JOHN R. RUSSO

Chapter Group Q
Northern California (J. R. Russo)
Southern California (J. R. Russo)
Utah (J. R. Russo)

Chapter Group R
British Columbia (H. Heaton)
Oregon (H. Heaton)
Oregon State College (H. Heaton)
Washington (H. Heaton)

Next Meeting of Board of Directors

The President announced that the next meeting of the Board of Directors would be the Annual Board Meeting, to be held August 7-8 at Dearborn Inn, Dearborn, Mich.

There being no further business to be considered, the first meeting of the 1958-59 Board of Directors was declared adjourned.

Respectfully submitted,
WM. W. MALONEY
General Manager

APPROVED:
L. H. DURDIN, *President*
June 22, 1958

Minutes

Meeting of AFS Finance Committee

Union League Club, Chicago — July 21-22, 1958

ROLL CALL: President L. H. Durdin, *Presiding*
Vice-President C. E. Nelson
Past President H. W. Dietert
General Manager Wm. W. Maloney
Treasurer E. R. May
Technical Director, S. C. Massari
Asst. Secretary A. B. Sinnett

Annual Audit. Following study of the annual audit of the Society's financial books as of June 30, 1958, by Geo. V. Rountree & Co., certified public accountants, Chicago, the Committee recommended approval by the Board of Directors.

Budgets of Income & Expense, 1958-59. Following detailed study of individual items of Income and Expense for 1957-58, and Staff proposals for budgets for the final year of the 2-year fiscal period 1957-59, the Finance Committee recommended approval by the Board of Directors of the following revised budget:

	1957-58 Actual	1958-59 Budget	2-Year Budget 1957-1959
Income	$1,027,556.	$766,000.	$1,793,556.
Expense	900,077.	885,500.	1,785,577.
Excess Income	$ 127,479.	$119,500.	$ 7,979.

The budget previously set up for 1958-59 was revised to account for the following:

1) Lower income based on 1957-58 Actuals.
2) Reduction in expenses to correspond with lower income but to support present Society programs at minimum expense.
3) Withdrawal of Cash for operating funds from the Investment Reserve Account as follows:

Balance on 1957 Land Purchase (est.)	$34,000.
Utilization of 1958-59 Investment Income	19,000.
	$53,000.

Details of Budget Recommendations. The budget of Income & Expense for the 1958-59 fiscal year, as recommended by the Finance Committee, included the following bases:

(1) *Membership Dues.* The Committee recommended a conservative Dues budget based on the reduced income and lower membership total of 1957-58. Suggested that colored reminder card be sent to Chapter Membership Chairmen for personal follow-up or by forwarding cards to delinquent members. Suggested that firms of Personal members be notified that dues have not been paid.

(2) MODERN CASTINGS. Although advertising rates were not increased during the year because of business conditions, the Committee noted a lower net operating income. AFS News Section and "Transactions" Sections incorporated in the magazine during 1957-58 were discussed, it being the consensus that they be retained at the increased costs provided in the budget.

(3) *Special Publications.* The Finance Committee reviewed Publications Committee minutes on June 16, 1958, and after deliberation recommended $23,300 be budgeted for new books in process and reprints or revisions of present books. Promotion to maintain sales should be continued strongly.

(4) *Convention Registration Fees.* The Finance Committee recommended that 1959 Registration Fees be $2.00 for all registrants as in 1957, to minimize cost to Design Engineers and Castings Buyers attending the 2d Engineered Castings Show.

Publishing of a "MODERN CASTINGS Daily" during the Convention was considered, but disapproved.

Suggestion was made that two issues of the "Transactions" sections, in MODERN CASTINGS, containing all Convention papers of interest to design engineers, be made available gratis to design engineers registering at the 2d Engineered Castings Show.

(5) *Exhibit Space Rentals.* The Committee recommended a budget of $87,500 Income from the 2d Engineered Castings Show, $31,400 Expense. To gain the greatest income possible, no free space to be allocated. In order to conserve income, the Educational Exhibits originally incorporated in the 2-year budget were disapproved. Space Rental Fees of $3.50 per sq. ft. were recommended.

(6) BUYERS DIRECTORY. As approved by the Board in 1957, a BUYERS DIRECTORY will be published in September 1959. Budgeted Income of $40,000 was recommended to cover estimated cost of $40,000. Future editions are expected to produce substantial net income.

(7) *Income from Investments.* All Income on Investments, including rent on recently-acquired building, to be utilized for operating purposes during 1958-59 only.

(8) *Salaries.* Recommended for approval in executive session of the Committee.

(9) *Burden Items.* After careful consideration of each item and their component breakdowns, total of $86,000 recommended covering all Burden items.

(10) *Training & Research Institute.* Transfers to T&RI for all purposes to be grouped as contributions to T&RI. Total budgeted in 1958-59 for all T&RI contributions, $61,100. The Committee accepted recommendations of the T&RI Research Committee, with the exception of $10,000 requested as a contingency fund for a Light Metals project now under contract with the Ordnance Corps of the U.S. Army.

(11) *Capital Expenditures.* Plans for converting a Parking area on the Northwest Highway property recently acquired were presented and tabled.

Respectfully submitted,
EDW. R. MAY
Treasurer

Minutes

Annual Meeting AFS Board of Directors

Dearborn Inn, Dearborn, Mich. — Aug. 7-8, 1958

ROLL CALL: President Lewis H. Durdin, *presiding*
Vice-President Charles E. Nelson

Directors (Exp. 1959)	Directors (Exp. 1960)
Harry W. Dietert	Wm. D. Dunn
Chas. E. Drury	A. A. Hochrein
Roger W. Griswold	Karl L. Landgrebe, Jr.
Herbert Heaton	Fred J. Pfarr
Jack H. King	John R. Russo
Wm. D. McMillan	Allen M. Slichter
Gerald R. Rusk	Henry G. Stenberg

Directors (Exp. 1961)
Thos. W. Curry
Richard R. Deas, Jr.
Jake Dee
Webb L. Kammerer
Hillard M. Patton
Clyde A. Sanders

General Manager Wm. W. Maloney
Asst. Secretary Ashley B. Sinnett
Technical Director S. C. Massari
Treasurer Edw. R. May

Absent: Directors Garnet P. Phillips, ex officio (..xp. 1959), Al. V. Martens (Exp. 1959), David W. Boyd (Exp. 1961).

A quorum having been established, President Durdin called the Annual Meeting of the Board of Directors to order.

Minutes of the meeting of the Board of Directors held May 20, 1958 were read and approved on motion duly made, seconded and carried.

The General Manager presented and briefed minutes of the meeting of the Board of Directors held May 24, 1958, which, on motion duly made, seconded and carried, were approved with the following changes:

(a) An agreed change in Chapter contacts, Director Dee being assigned to visit the Mexico Chapter; Director Kammerer, the Timberline Chapter.

(b) It was announced by President Durdin, on agreement between Directors McMillan and Sanders, that Director McMillan would be the official Board contact with the Chicago Chapter.

Report of the General Manager

The General Manager's report for the fiscal year ended June 30, 1958, (see separate report) covered Membership, Chapters, Conventions & Exhibits, Education, MODERN CASTINGS, SH&AP, Training & Research Institute, International Affairs, BUYERS DIRECTORY and General Administration.

On motion made, seconded and carried, the report was accepted and is made a part of these Minutes.

Report of the Treasurer

The Treasurer's report for the fiscal year ended June 30, 1958 (see separate report) showed Income of $1,027,556.62, or $61,-393.38 under the budgeted $1,088,950 . . . Expense $900,077.39 or $20,272.61 under the budgeted $920,350 . . . Excess Expense $41,120.77 or 24.39% under budgeted Excess Income. Individual items of Income and Expense were examined in detail as audited by the firm of George V. Rountree & Co., Chicago.

An analysis of the 1957-58 financial status was presented in relationship to progress of AFS in the past 12 years. During the period 1946 to 1958 the Society's earned Excess Income averaged $19,000 annually; at the same time the Society's financial stability was very materially increased. For example, Invested Securities during the period increased $379,034, Fixed Assets increased $340,325, Inventories increased $14,807, and Fund Principals increased $868,373.

On motion duly made, seconded and carried, the report of the Treasurer was accepted and is made a part of these minutes.

Report of the Technical Director

The annual report of the Technical Director covered the 1958 Convention technical program, preprinting of "Transactions" papers in MODERN CASTINGS, demand for TRANSACTIONS bound volumes and Special Publications, Research projects and Committee activites.

On motion made, seconded and carried the Technical Director's report was accepted and is made a part of these minutes.

Reports of Board Committees

Chapter Contacts Committee. Vice-President Nelson urged continued efforts of the Directors in meeting with assigned Chapters, in addition to attending the Regional Administration Meetings.

Committee of Exhibitors. The General Manager read minutes of the Exhibitors Committee meeting held June 27, 1958. Director Sanders reviewed the findings of the Exhibits Analysis Committee meeting held July 23, 1958 to consider whether or not the frequency of AFS exhibits should be changed, a study instigated by requests from the Foundry Equipment Manufacturers Association, Foundry Facings Manufacturers Association and National Industrial Sand Association that AFS consider a change in the exhibit cycle.

Considerable discussion followed the findings of the committee, outlined in the following points as representing a consensus of the Board:

(a) Decision concerning exhibit frequency should be made solely by the AFS Board of Directors, representing as it does the Foundry industry as a whole.

(b) A further study on the effect of a change in exhibit frequency on the financing of the Society should be made.

(c) Attendance of AFS Exhibits every two years is needed by foundrymen, both new men as well as present men in the industry, for development and modernization of their plants, products and the castings industry as a whole.

(d) Competition within the industry exists from month to month, and an Exhibit every two years seems to be a maximum frequency. AFS should be primarily concerned with service to the membership and the industry on technical aspects of the industry, and there should be no change in the present AFS Exhibit policy as to frequency of these events.

Following Board study of data submitted by the Exhibits Analysis Committee, the following resolution was moved, seconded and approved:

RESOLVED that AFS in 1964 hold a technical Castings Congress, together with an Equipment and Supplies Exhibit, in a location to be selected.

The Board further agreed that President Durdin should arrange with the Exhibits Analysis Committee, separate meetings with the Foundry Equipment Manufacturers Association, the Foundry Facings Manufacturers Association and the National Industrial Sand Association so as to outline and discuss the findings of the Committee and the decision of the Board.

It was the decision of the Board that the Exhibits Analysis Committee should be dissolved and existing personnel reappointed as an Exhibits Promotion Committee, this committee to work with the Staff in an effort to better the Engineered Castings Shows and Equipment Shows in future years.

Publications Committee. The Technical Director reviewed minutes of the Publications Committee meeting held June 16, 1958. New publications recommended for 1958-59 include: Noise Control Manual and Radiation Hazards Manual. Publications to be reprinted and/or revised: STATISTICAL QUALITY CONTROL, PATTERNMAKER'S MANUAL, FOUNDRY CORE PRACTICE. It was suggested that the Society might consider investigating the possibility of translating and publishing some of the principal AFS books in Spanish.

Discussion developed on the need for a publication dealing with Metal Specifications, Testing (correlation and test bars, frequency, etc.), Casting Finish, Casting Dimensions, and Defects (rejects) resulting in scrap, with the primary objective of offering a new AFS service and increasing membership by preparing information for engineers, purchasing people, and quality control departments in consumer plants. It was the consensus of the Board that such an activity should be developed by a "Technical Information Committee" with Director Curry as Chairman, Directors Stenberg and Sanders, Technical Director Massari, and H. W. Ruf of Grede Foundries. It was suggested that Mr. Curry contact members of A.S.T.M. Committee A-3 to serve. It was suggested that the committee's main work should be "what the consumer of castings can expect from the foundry."

Question was raised on the advisability of continued free distribution of TRANSACTIONS to all Company and Sustaining members. President Durdin delegated study of the suggestion to the Membership Analysis Committee.

Annual Lecture Committee. The Technical Director reported that the 1957-58 Annual Lecture Committee had chosen H. M. St. John as the annual lecturer for 1959.

National Castings Council. President Durdin highlighted minutes of the National Castings Council meeting of May 12, 1958, as previously mailed to all Directors for study.

F.E.F. Trustees. Director Dunn presented a written report on the F.E.F. Trustees meeting held May 22, 1958, dealing with scholarship commitments, finances, cooperation with government scholarship programs and assistance for graduate students so as to provide top level specialists for future leadership.

The relation between AFS and F.E.F., insofar as two-year technical institute programs are concerned, was raised and discussed. It was pointed out that the existing AFS Education Policy is to (a) "support and encourage educational activities at all levels of education," and (b) "confine its educational activities at this time primarily to foundry plant personnel training programs and to schools at the secondary level of education." It was stated that no official discussions have been held to indicate whether two-year technical institute programs lie within the field of AFS or F.E.F., but that F.E.F. has explored this field for possible support. Following discussion, a resolution was moved and seconded:

RESOLVED THAT AFS reaffirms its original Educational Policy and instructs the AFS Trustees to F.E.F. to oppose the granting of F.E.F. scholarships at institutions not qualified to offer Bachelor degrees in the sciences.

Motion carried, two Directors dissenting.

Staff Retirement Plan. The General Manager reported that the Retirement Plan Trustees are now currently reviewing a new AFS Staff Retirement Plan, for recommendations to the Board.

Training & Research Institute

Director Massari reported on minutes of the May 20, 1958 meeting of Trustees. The Trustees expressed favorable comments on the number of students enrolled in courses thus far and on finances of the Institute. Budgetary requirements for the year 1958-59 were reviewed. The Director commented that as of June 30, 538 students had been enrolled in 14 courses, resulting in tuition fees income of $25,610, and direct course expense of $15,900.

The General Manager suggested that further consideration of permanent building facilities for T&RI should be postponed until such time as T&RI can obtain a proper tax exempt status. He stated that a nearby conference had been scheduled with the Internal Revenue Department for further discussions.

1959 Engineered Castings Show

The General Manager requested Board decision on rules of exhibitor eligibility for the 1959 Engineered Castings Show. He stated that the 1957 Committee of Exhibitors had recommended that exhibits in 1959 be limited strictly to producers of castings for sale. This recommendation was tabled at the Annual Board Meeting in August 1957, for consideration by the current Board.

He pointed out that the 1957 Committee of Exhibitors included, for the first time, seven foundry exhibitors from the 1957 Engineered Castings Show, and that a meeting of the same committee was held June 27, 1958. Since only one of the appointed seven foundry exhibitors attended this meeting, the General Manager presented the following Staff recommendations for 1959:

(a) That 1959 exhibits be limited to: Producers of Castings for sale; producers of Patterns for sale; producers of metals and alloys essential to castings quality; and manufacturers of laboratory, testing and inspection equipment for the control of castings quality. Design engineers and castings buyers being interested in all four types of exhibits, undue restrictions on the 1959 Show would limit the value of the Show for specialized, non-foundry attendance.

Following discussion, the above recommendation was accepted on motion made, seconded and unanimously carried.

The General Manager then pointed out that only 25,000 square feet of floor space was available at Chicago in 1959, and that the 1958-59 Income budget called for sale of the entire square footage available. Consequently, he said, the Finance Committee recommended (a) that no free exhibit space be made available to the foundry trade associations, and (b) the expediture of $15,000 for educational exhibits depicting engineering properties of all cast metals be dropped from the budget.

Following discussion, the above Finance Committee recommendations were accepted on motion made, seconded and carried.

Approval of Annual Audit

President Durdin reported on the Society's Income and Expense for 1957-58, and asked that the Board approve the audit of Society finances as of June 30, 1958, prepared by George V. Rountree & Co., certified public accountants, Chicago. Approved, on motion made, seconded and carried.

Death of Director Alex Pirrie

President Durdin announced with deep regret the untimely death on May 22 of Director Alex Pirrie, American Standard Products Ltd., Toronto, Ont. The President reported expressions of sympathy had been expressed on behalf of the entire

Board of AFS to Director Pirrie's family and associates, and directed the Secretary to spread upon these minutes an expression of the Society's highest regard and respect for Alex Pirrie.

Announcement of New Directors

President Durdin announced that, under the new by-laws of the Society effective July 1, Director G. P. Phillips had been elected to Director *ex officio* status by letter ballot of the Board, due to inability to attend Board meetings because of continuing ill health. He also stated that, by the same letter ballot, W. D. McMillan, Supervisor of Metallurgy, McCormick Works, International Harvester Co., Chicago, had been elected Director for the balance of Director Phillips' term of office, expiring April 1959.

The President stated that ex officio Director Phillips would continue to receive all Board announcements and that he and Director McMillan had agreed to consult on all matters requiring Board decision. The President stated that, by agreement, Director McMillan would be the official Board contact with the Chicago Chapter.

President Durdin announced that J. H. King, Vice-President of Archer-Daniels-Midland Co. (Canada), Ltd., Toronto, Ont., had been elected Director by letter ballot of the Board to fill the unexpired term of Alex Pirrie, deceased, to April 17, 1959. He added that Director King would be the official Board contact with the Ontario and Eastern Canada Chapters.

The President pointed out that Directors McMillan and King had been elected by letter ballot so that both might be able to attend the Annual Board Meeting. Both Directors then were introduced and welcomed.

Election of Staff Officers

In executive session, the following Staff officers were elected for the year 1958-59, in accordance with the new by-laws:

General Manager — Wm. W. Maloney, reelected.
Secretary — A. B. Sinnett, formerly Asst. Secretary.
Treasurer — E. R. May, reelected.
Technical Director — S. C. Massari, reelected.

Staff Salaries and Compensations

In executive session of the Board, recommendations of the Finance Committee for Staff salaries and compensations for the fiscal year 1958-59 were approved, with certain revisions.

Budgets of Estimated Income and Expense

Recommendations of the Finance Committee for budgets of Estimated Income and Expense for the year 1958-59 were considered in detail. Specific items were explained in detail, in relation to 1958-59 activities and as a part of the 2-year budget for 1957-1959. Following full discussion, budgets of Estimated Income and Expense for the year 1958-59 and the revised 2-year budget 1957-1959 were approved as recommended by the Finance Committee, with certain revisions.

Approval of Resolutions

The following resolutions required annually by banks of deposit for the deposit and withdrawal of Society funds and documents were approved on blanket motion made, seconded and carried:

RESOLVED that resolutions required by the Harris Trust & Savings Bank of Chicago, Illinois, and by the First National Bank of Des Plaines, Illinois, authorizing the withdrawal of Society funds, are hereby approved and the General Manager authorized to certify thereto.

RESOLVED that checks for the withdrawal of funds deposited in the name of the Society with depository banks, including all Checking accounts and Interest Savings accounts, and for the disposition of all temporary Securities held in the various funds of the Society by the Harris Trust & Savings Bank of Chicago and/or the First National Bank of Des Plaines, shall require the signature of any two of the following: President, Vice-President, General Manager, Secretary, Treasurer.

RESOLVED that directives to the Harris Trust & Savings Bank concerning the AFS Investment Trust Account shall be only by means of a letter or resolution bearing the personal signatures of all members of the AFS Finance Committee. Where immediate action is necessary, the Treasurer of AFS is designated to so inform the Harris Trust, but such information must be promptly confirmed by personal signatures.

RESOLVED that the Secretary be authorized to maintain, if deemed feasible, a Safety Deposit Box in the name of the Society at some convenient location for the safekeeping of Society documents, and that any two of the following have authority to obtain access to such safety deposit box: President, Vice-President, General Manager, Secretary, Treasurer.

RESOLVED that the General Manager be authorized to execute all Contracts for the administration of Society affairs, subject to specific approval by the Board of Directors. In the case of AFS-sponsored research projects, approval of projects by the Board of Directors includes authority for the General Manager to execute contracts for the performance of such projects on a bid basis.

RESOLVED that the General Manager be authorized to reimburse Travel expenses for members of the Society in attendance at any regularly called meeting of the Board of Directors, official Administrative and Special committees, and Technical committees, with the following exception: No expenses shall be paid to Directors or committee members for attendance at meetings held during the week of the Annual Convention of the Society, unless specifically authorized by the Board of Directors.

RESOLVED that the General Manager be authorized to negotiate the compensations of necessary Staff employees below $6,000 per year, and that Finance Committee approval be required on all compensations of $6,000 per year or more.

RESOLVED that Indemnity Bonds be maintained covering responsible financial Officers of the Society, as follows: Blanket Position Indemnity Bond covering all Staff Members and including all Officers of the Society, whether compensated or not, in the amount of $30,000 each. All premiums for such Indemnity Bonds to be paid by the Society.

Washington Chapter Inquiry

The General Manager presented a letter from Spokane, Wash. expressing interest in forming a new AFS Chapter there and requesting information on Chapter requirements. The President requested Director Heaton and Regional Vice-President Russo to meet with representatives of the Spokane group to determine the degree of interest in terms of AFS Chapter requirements.

Official Delegates to International Congress

The General Manager reported that, by appointment of the President, AFS would be officially represented at the 1958 International Foundry Congress in Belgium by Director R. R. Deas, Jr. and Dr. A. B. Everest of the Mond Nickel Co.

AFS Participation at Internationals

The General Manager recommended that AFS, as an important member of the International Committee of Foundry Technical Associations, should attempt to organize an American party for future Internationals held abroad, particularly for the 1959 International in Spain and the 1960 International in Switzerland. He stated that sizable U.S. representation at these two events should materially assist in developing foreign attendance at the 1962 International which AFS is committed to sponsor in Detroit. He stated that the President had approved organization of official tours in both years.

Appointment of 1958-59 AFS Nominating Committee

The President announced appointment by the Executive Committee of the 1958-59 Nominating Committee, in meeting held August 7.

It was pointed out that hitherto those Directors elected by the Board of Directors had been considered "official representatives" of the Chapter Groups wherein they reside. Since these Directors now are officially termed "Directors-at-large" under the new AFS by-laws, the President said the Executive Committee had agreed that Directors-at-large should not be considered as Board representatives of any particular Chapter group. Accordingly, the following resolution of the Executive Committee was presented for Board approval:

RESOLVED, that Article XI of the Society's by-laws effective July 1, 1958, should be interpreted that Board-elected Directors-at-large shall not be considered henceforth by AFS Nominating Committees as representing, of themselves, those Chapter Groups in which they may reside.

Resolution approved on motion made, seconded and carried, and the Secretary instructed to notify the 1958-59 Nominating Committee.

Announcement of Next Board Meeting

The President announced that the next meeting of the Board of Directors would be held at the Palmer House, Chicago, February 11, 1959.

There being no further business to be considered, the Annual Meeting of the Board of Directors was declared adjourned.

Respectfully submitted,
ASHLEY B. SINNETT
Secretary

APPROVED:
L. H. DURDIN, *President*
August 16, 1958

Minutes

Special Meeting of Executive Committee

Dearborn, Mich. — Aug. 7, 1958

ROLL CALL: President L. H. Durdin, *presiding*
Vice-President C. E. Nelson
Past President H. W. Dietert
Regnl. Vice-Presidents
H. G. Stenberg
R. W. Griswold
F. J. Pfarr
C. E. Drury
A. M. Slichter
K. L. Landgrebe, Jr.
J. R. Russo

A quorum having been established, President Durdin called the meeting to order.

The new By-Laws of the Society were read, describing the method by which the Nominating Committee is appointed for the election of a President, a Vice-President and new Directors of the Society at the 1958 Convention. Names of various candidates were presented from lists submitted by 13 eligible Chapters. It was reported that 11 eligible Chapters had not submitted the names of candidates, 23 Chapters being "ineligible" under the By-Laws.

1959-60 Nominating Committee

In accordance with the By-Laws, six members were appointed by the Executive Committee to form, together with Past Presidents H. W. Dietert and F. W. Shipley, the 1958 Nominating Committee, as follows:

Past President Harry W. Dietert, Chairman of the Board, Harry W. Dietert Co., Detroit, *Chairman.*

Past President Frank W. Shipley, Fdy. Mgr., Caterpillar Tractor Co., Peoria, Ill.

Erwin W. Deutschlander, Supt., Worthington Corp., Buffalo, N.Y., (rep. Region I, Chapter Group E—Western New York Chapter and Gray Iron).

F. A. Dun, Foreman, Babcock & Wilcox Co., Barberton, Ohio, (rep. Region III, Chapter Group G—Canton District Chapter and Gray Iron and Steel).

Donald G. Schmidt, Met. Engr., H. Kramer & Co., Chicago, Ill., (rep. Region V, Chapter Group K—Chicago Chapter and Supplies).

James D. Tracy, President, Salmon Bay Foundry Co., Seattle, Wash., (rep. Region VII, Chapter Group R—Washington Chapter and Gray Iron).

Marcel Trottier, Tech. Sales Rep., Quebec Iron & Titanium Corp., Sorel, P.Q., Canada, (rep. Region II, Chapter Group D—Eastern Canada Chapter and Supplies).

L. J. Woehlke, Plant Mgr., Grede Foundries, Inc., Waukesha, Wis., (rep. Region V, Chapter Group L—Wisconsin Chapter and Gray Iron and Steel).

The General Manager was instructed to notify immediately those selected and to obtain prompt acceptance.

Chapter Group Representation

The point was raised as to whether election of one Director annually by the Board should properly constitute "Chapter Group Representation" on the Board. It was pointed out that the new By-Laws now refer to a Board-elected Director as a "Director-at-large." It therefore was moved, seconded and carried as a recommendation for Board approval, that Art. XI of the By-Laws "shall be so interpreted that Board-elected Directors-at-large shall not be considered by the Nominating Committee as representing, of themselves, the Chapter Groups in which they may reside."

There being no further business to be considered, the meeting was adjourned.

Respectfully submitted,
ASHLEY B. SINNETT
Secretary

APPROVED:
L. H. DURDIN, *President*
August 16, 1958

Minutes

Meeting of 1958-59 Nominating Committee

Hotel Sherman, Chicago — December 8, 1958

ROLL CALL: Chairman H. W. Dietert, Past President, *Presiding*
F. W. Shipley, Past President
Region 2 — Marcel Trottier, Sorel, Que.
Region 2 — E. W. Deutschlander, Buffalo, N. Y.
Region 3 — F. A. Dun, Barberton, Ohio
Region 5 — D. G. Schmidt, Chicago
Region 5 — L. E. Woehlke, Milwaukee
Region 7 — J. D. Tracy, Seattle, Wash.

A quorum having been established, Chairman Dietert called the meeting to order.

Nomination of President

On motion duly made, seconded and unanimously carried, the incumbent AFS Vice-President, CHARLES E. NELSON, Technical Director, Magnesium Div., Dow Chemical Co., Midland, Mich., was nominated President of the Society for the year 1959-60.

Nomination of Vice-President

Following submission of names of candidates and on motion duly made, seconded and unanimously carried, NORMAN J. DUNBECK, Vice-President, Industrial Minerals Div., International Minerals & Chemical Corp., Inc., Skokie, Ill., was nominated Vice-President of the Society for the year 1959-60.

The Chairman announced that Mr. Dunbeck has expressed willingness to accept nomination only if it were understood and made known publicly that the AFS President and Vice-President henceforth may function precisely as stated in the Society's By-Laws, namely, as Executive Officers of the Society, with minimum travel requirements. All succeeding Nominating Committee actions were taken with this principle of operation known and acceptable to the Committee.

Nominations of Directors

By-Laws of AFS were read as relating to organization of the Board of Directors and Nominating Committee procedures. The

Chairman pointed out that the Committee was expected to name six National Directors, each for a term of three years commencing April 18, 1959, and each from a Chapter Group which, without action by the Committee, would not have continuous representation on the National Board.

The Chairman also stated that by action of the AFS Board, Directors-at-large elected by the Board of Directors under the by-laws should not be considered by the Nominating Committee as providing, in themselves, "Chapter Group representation" on the Board. Seven Chapter Groups then were listed for Nominating Committee consideration.

Candidates by the Chapters, as requested in accordance with the By-Laws, were presented. It was stated that of 16 Chapters "eligible" to submit candidates, one had not submitted names to the Committee, and 31 Chapters were considered "ineligible" by reason of continuing Board representation after April 1959.

Following submission of all candidate names, the Nominating Committee nominated the following AFS members to serve as National Directors for terms of office indicated, and representing Chapter Groups, Chapters and Industrial Interests as shown:

Chapter Group D (Region 2) — Eastern Canada Chapter
A. J. Moore, Exec. Vice-President, Canadian Bronze, Ltd., Montreal, Que., Canada – representing Brass & Bronze.

Chapter Group E (Region 2) — Western New York Chapter
Wm. H. Oliver, Manager, Bond Plant, American Radiator & Standard Sanitary Corp., Buffalo, N. Y. – representing Gray Iron.

Chapter Group G (Region 3) — Canton District Chapter
Robert E. Mittlestead, Met., Lectromelt Casting Div., Akron Standard Mold Co., Barberton, Ohio – representing Steel, Gray Iron.

Chapter Group H (Region 4) — Detroit Chapter
Cecil N. King, Plant Mgr., Chrysler Fdy., Detroit Chrysler Corp., Detroit, Mich. – representing Gray Iron, Steel, Brass & Bronze, Light Metals.

Chapter Group L (Region 5) — Wisconsin Chapter
Norman N. Amrhein, President, Federal Malleable Co., West Allis, Wis. – representing Malleable Iron.

Chapter Group R (Region 7) — Washington Chapter
James N. Wessel, Supv. Materials Engineer, Puget Sound Naval Shipyard, Bremerton, Wash. – representing Gray Iron, Steel, Brass & Bronze.

Acceptances

All nominees selected by the Nominating Committee were contacted by telephone and acceptances personally obtained.

General

The Chairman stated that all names of candidates submitted by the Chapters and not acted upon by the Nominating Committee would be referred to a special Board Nominating Committee, for Board election of one Director-at-large as provided in the by-laws. He stated that special attention would be called to the fact that the Nominating Committee, in naming nominees from six eligible Chapter Groups, had been unable to nominate anyone from eligible Chapter Group K (Chicago Chapter), presently represented by Director-at-large Sanders and Vice-President nominee Dunbeck.

It was the consensus that more information on the AFS nominating procedure should be made available to the Chapters, and that all eligible Chapters should be urged to submit names of candidates annually, preferably representing several industrial interests.

Election Procedures

The Chairman pointed out that under the by-laws additional nominations may be made, by written petition signed by 200 members in good standing, at any time 45 days prior to the Society's Annual Business Meeting. Newly elected Officers and Directors assume office on the day following the close of the Society's Annual Meeting.

Members of the Committee were requested to hold all nominating discussions strictly confidential, first publication of names of nominees to appear in MODERN CASTINGS.

There being no further business to be considered, the meeting was adjourned.

Respectfully submitted,
HARRY W. DIETERT, *Chairman*

December 12, 1958
APPROVED:
F. W. SHIPLEY, *Vice-Chmn.*
Feb. 2, 1959

Minutes

Meeting of AFS Board of Awards 1958-59

Union League Club, Chicago — Dec. 9, 1958

ROLL CALL: Past President W. L. Seelbach (1951-52), *Presiding*
Past Presidents:
I. R. Wagner (1952-53)
Collins L. Carter (1953-54)
Frank J. Dost (1954-55)
B. L. Simpson (1955-56)
Frank W. Shipley (1956-57)
H. W. Dietert (1957-58)
Wm. W. Maloney, Board of Awards Secretary

The Chairman thanked all present for another 100% attendance, and asked that minutes of the 1957 Board of Awards meeting be read, as previously approved.

Agreement on Procedures

In advance of consideration of specific candidates for awards, a number of procedures were agreed upon:

(a) Following study of Award fund accumulations and past frequency of medal awards, agreed that the Penton, Whiting and Simpson medals would be considered for 1959 awards.

(b) Suggestions for the various awards as orginally indicated by donors were studied, it then being agreed that all medals would be considered interchangeable for purposes of 1959 awards.

(c) Agreed that the names of all recipients of an Award of Scientific Merit or Service Citation be considered eligible for future Board of Awards consideration.

(d) Agreed that all persons submitting nominations for award consideration, not acted upon by the Board of Awards, be thanked for interest and cooperation over the signature of the Chairman, and urged to continue submission of new candidates.

(e) Agreed that the Secretary would prepare and maintain a single, permanent book of all nominees submitted in previous years, kept up-to-date in terms of award decisions; and that the Board of Awards receive full data on all new and continuing candidates in advance of each annual meeting.

Gold Medal Awards

Following consideration of all candidates submitted, the Board of Awards recommended, on motions made, seconded and carried, the following to receive AFS Gold Medals in 1959 for the reasons stated, subject to approval by the Board of Directors:

(a) To HAROLD W. LOWNIE, JR. (Chief, Metallurgy Research Div., Battelle Memorial Institute, Columbus, Ohio) The John H. Whiting Gold Medal . . . "for outstanding contributions to the Society and to metallurgical progress, especially in the field of Gray Cast Iron."

(b) To JOHN A. RASSENFOSS (Manager, Manufacturing Research Laboratory, American Steel Foundries, East Chicago, Ind.) The Peter L. Simpson Gold Medal . . . "for exceptional contributions to the Society and the Steel Castings Industry, by elevating technical endeavor in steel foundry research."

(c) To FRED J. WALLS (Retired former District Manager, International Nickel Co., Detroit) The John A. Penton Gold Medal . . . "for outstanding contributions to the Society and the Castings Industry, especially in the field of Gray Iron

metallurgy and foundry practice, and for life-time devotion to the industry's potentialities."

Awards of Scientific Merit

Following consideration of candidates, the following Awards of Scientific Merit for 1959 were recommended for approval by the Board of Directors, on motions made, seconded and carried:

(a) To ROBERT H. MOONEY (Division Works Engineer, Central Foundry Div., General Motors Corp., Saginaw, Mich.) . . . "for technological contributions to production and working conditions in the Malleable Iron industry."

(b) To HOWARD H. WILDER (Manager of Engineering Sales, Vanadium Corp. of America, Chicago) . . . "for noteworthy contributions to the Society and its membership in furthering the technology of Ferrous Castings and their acceptance."

(c) To ELMER C. ZIRZOW (Director of Foundry Service, Werner G. Smith, Inc., Cleveland) . . . "for long and conscientious effort on behalf of the Society's Sand Division in the development and application of Sand technology."

Service Citation Awards

Following consideration of candidates, the following AFS Service Citations for 1959 were recommended for Board of Directors approval, on motions made, seconded and carried:

(a) To JAMES R. ALLAN (President, Allan Industries, Melbourne, Fla.) (Formerly Manager, Industrial Engineering & Construction Dept., International Harvester Co., Chicago — Retired.) . . . "for unswerving devotion to making the Foundry a better and a safer place to work, and for his leadership in developing practical engineering standards for the improvement of foundry working conditions."

(b) To BERNARD D. CLAFFEY (Vice-President, Dayton Malleable Iron Co., Dayton, Ohio) . . . "for outstanding service to the Society and the Castings Industry in the work of its Chapters, Technical Committees and Educational development."

(c) To ROY W. SCHROEDER (Professor, Dept. of Mechanical Engineering, University of Illinois, Chicago) . . . "for long and devoted service to the Society's activities in Apprentice Training and Foundry Instruction, and for his unceasing guidance of young men toward foundry careers."

Honorary Life Member Award

On motion made, seconded and unanimously carried, the following Honorary Life Membership award was made subject to Board of Directors approval:

To LEWIS H. DURDIN, on completion of his term of office as AFS President 1958-59.

It was pointed out that the 1959 Annual Lecturer, Harry M. St. John, had received Honorary Life Membership in 1947, as the recipient that year of the Society's McFadden Gold Medal. The Secretary therefore was instructed, in accordance with the past practice, to arrange a suitable gift or emolument for presentation to Mr. St. John in lieu of Life Membership.

General

The Board of Awards agreed that all 1959 awards would be presented as follows: Awards of Scientific Merit and Service Citations, at the Annual Business Meeting; Gold Medal awards at the Annual Banquet; all awards presented by Chairman Seelbach; all award recipients to receive leather-bound copy of programs of their respective meetings.

It was agreed that all award recipients would be notified of their selection, following approval by the Board of Directors, by means of telegrams signed by Chairman Seelbach, and that first publicity on selectees would appear in MODERN CASTINGS in the earliest possible issue.

There being no further business to be considered, the meeting was adjourned.

Respectfully submitted,
WM. W. MALONEY
Board of Awards Secretary

APPROVED:
W. L. SEELBACH, *Chairman*
Nov. 4, 1959

Minutes

Special Meeting of AFS Board of Directors

AFS Central Office, Des Plaines, Ill. — Tues. Feb. 10, 1959

ROLL CALL: President Lewis H. Durdin, *presiding*
Vice-President Charles E. Nelson
H. Bornstein, Chairman, T&RI Trustees
General Manager Wm. W. Maloney
Secretary Ashley B. Sinnett
Technical Director S. C. Massari
Treasurer Edw. R. May

Directors (exp. 1959)
Charles E. Drury
Roger W. Griswold
Herbert Heaton
A. V. Martens
Wm. D. McMillan

Directors (exp. 1960)
Wm. D. Dunn
A. A. Hochrein
Karl L. Landgrebe, Jr.
Fred J. Pfarr
John R. Russo
Allen M. Slichter
Henry G. Stenberg

Directors (exp. 1961)
David W. Boyd
Richard R. Deas, Jr.
Jake Dee
Webb L. Kammerer
Hillard M. Patton
Clyde A. Sanders

Officer and Director Nominees (on invitation)
Vice-Pres. Nominee Norman J. Dunback
Director Nominees:
Norman N. Amrhein
Donald L. Colwell
A. Jack Moore
Wm. H. Oliver
James N. Wessel

Absent: Directors H. W. Dietert, J. H. King, G. R. Rusk and G. P. Phillips (exp. 1959); Director T. W. Currey (exp. 1961); Director Nominees Cecil N. King and Robert E. Mittlestead.

A quorum having been established, President Durdin announced that Director Dietert was unable to attend because he was out of the country; that the 100% attendance otherwise expected had proved impossible because bad weather had prevented several (Messrs. J. H. King, Rusk, Curry, Mittlestead) from flying or driving to Chicago; and that Director Nominee C. N. King was forced to cancel because of sudden labor negotiations. (Director Nominees Mittlestead and J. H. King arrived later.)

The meeting was opened by introduction of Vice-President Nominee N. J. Dunbeck and Director Nominees N. N. Amrhein, D. L. Colwell, A. J. Moore, W. H. Oliver and J. N. Wessel, attending on invitation.

Purpose of Meeting

The President thanked all Directors and Nominees present for attending the Special Meeting of the Board and reiterated the meeting's purpose, namely, to acquaint all with the program and objectives of the Training and Research Institute and to consider certain recommendations of the Institute Trustees. He stated that, while all Director Nominees were urged to enter fully into the discussions, voting on any motions made would necessarily be confined to incumbent Officers and Directors.

Statement by Trustees Chairman

President Durdin then stated that he had personally invited H. Bornstein, Chairman of T&RI Trustees, to attend and present the Trustees' second annual report and recommendations, in person. Chairman Bornstein then was introduced and presented the following statement:

"I am pleased to be invited to your meeting today. Lew Durdin has asked me, as Chairman of Trustees of T&RI,

to present a statement to you and to be available for discussion.

"You already have the brochure prepared by the Staff and also the Trustees annual report. A review of the activities of T&RI may be helpful, particularly to those of you who were not on the AFS Board of Directors at the inception of this project.

"The Staff of AFS issued a brochure dated October 26, 1956, describing a program for the development of a Training and Research Institute. This proposed program included a building at the AFS Headquarters site at an estimated cost of $200,000. The AFS Board of Directors approved the proposed program at its meeting on November 16, 1956. Then the Trustees of T&RI were selected. As you know, three of the Trustees are from the AFS Board: the President, Vice-President and Past-President. The other four Trustees are elected by the AFS Board and these trustees serve terms of 4 years.

"The first meeting of the Trustees was held December 21, 1956. It soon appeared that there were serious differences of opinion. Furthermore, only four of the seven Trustees were present. It was decided that another meeting should be held soon, with full attendance if possible.

"The second meeting was held February 24, 1957 with 100% attendance. Again, there were serious differences of opinion. Some Trustees wanted to proceed immediately with the construction of a building on the site of AFS Headquarters. Other Trustees felt that this was moving too fast, since we did not know how well the courses would be received and whether or not students would travel considerable distances to a central location. Also, there was some question as to the actual cost of the building. Some Trustees felt that the figure of $200,000 might be too low.

"After an all day meeting, the Trustees reached unanimous agreement as follows:

"(a) Appointed S. C. Massari Director of the Institute.
"(b) Authorized Mr. Massari to employ a Training Supervisor.
"(c) Appointed a committee to investigate and study plans for building.
"(d) Provided for trial courses at several locations as follows:
(a) Foundry Sand — at Detroit
(b) Cupola Melting — at Chicago
(c) Industrial Engineering — at Milwaukee

"It was decided that each course should be given several times. The results obtained from these trial courses would determine further action.

"Another meeting of the Trustees was held during the AFS Convention on May 7, 1957. At that meeting it was decided that the T&RI building should be located on the site of the AFS Headquarters. Authorization was given to the Building Committee to employ an architect to prepare drawings and specifications at a cost not to exceed $7,000. Mr. Massari outlined plans for further courses.

"The Building Committee selected an architect and plans were submitted at the December 1957 meeting of the Trustees. These plans and specifications were worked out in collaboration with Mr. Massari and other members of the Staff. The architect's estimate of the building was $409,000. To this figure should be added about $100,000 for equipment.

"Because the results obtained from the trial courses were very satisfactory, the Trustees at the December 1957 meeting, by unanimous action, recommended to the AFS Board of Directors that the Board now consider proceeding with the construction of the proposed Foundry Training Center building.

"You will recall that I appeared before you at your meeting last February and called attention to the recommendations of the Trustees. At that time decision was deferred because of business conditions.

"Now, a word about tax exemption. One of the governing reasons for setting up T&RI as a separate organization was the tax situation. AFS is not an exempt organization under Section 501 (3)(c) of the Internal Revenue Code in 1954. Our attorneys believe that T&RI would have such exemption. This exemption is required to secure funds such as those offered by the Pangborn Foundation.

"The Internal Revenue Department has ruled against us. As indicated in my annual report, we have hired a firm of Washington tax attorneys to represent us on appeal. There have been delays and we do not know when a decision will be made.

"Another year has passed since the recommendation of the Trustees to the AFS Board of Directors. More courses were given in 1958 and again the results were favorable. The Trustees recognize that more and better courses could be given if adequate facilities were available. This again brings up the question of a building. At the December 1958 meeting of the Trustees a resolution was passed by unanimous vote, recommending that the Board of AFS authorize construction as soon as possible. It is recognized that it will be at least two years after such Board authorization that the building would be available.

"In my opinion, the Trustees have acted conservatively yet expeditiously on the Institute program. In manufacturing a new item it is usual to start with an experimental program. If successful this is followed by pilot plant operation, and then by production in quantity.

"We had our experimental program on courses in 1957. This was followed by pilot plant operation in 1958, which will be continued in 1959. We say to you that we are ready for production, provided you furnish the necessary facilities.

"Gentlemen, it is your next move."

President Durdin read for the group the action of the Trustees at their December 1958 meeting where they voted unanimously the following recommendation:

"That the Trustees of the AFS Training and Research Institute believe that greatly increased demands for qualified personnel will be made by all industry in the next few years, and that the need for broader training of all occupational groups will steadily increase. The Trustees desire to point out that the proposed Training Center building is definitely needed by the Institute to broaden and improve its program. The T&RI Trustees therefore recommend that the AFS Board of Directors authorize construction of the proposed Training Center at the earliest possible date. It probably will take at least two years to complete the project after authorization is made by the AFS Board."

Mr. Bornstein then called attention to the growing need for manpower training and stated that if the Board should give the building an immediate "green light" it would still require nearly two years to complete the proposed building, by adhering to the following "Estimated Timetable for Construction":

ESTIMATED CONSTRUCTION TIMETABLE

Feb. 1959 — Final approval of design and plans.
March 1959 — Preparation of detailed specifications begun.
July 1959 — Approval of detailed specifications.
Aug. 1959 — Specifications mailed to selected general contractors.
Nov. 1959 — Deadline for general contract bids.
Nov. 1959 — Selection of general contractor.
Dec. 1959 — Arrange financing.
Jan. 1960 — Commence solicitation of contributions.
April 1960 — Commence building construction.
Oct. 1961 — Completion of building.
Nov. 1961 — First T&RI Courses in new building.

Discussion of Overall T&RI Program

President Durdin stated that all phases of the Institute program had been approved in November 1956, with the exception of actual decision on construction of the Foundry Training Center building. Particular attention was called to the following paragraph from the Enabling Resolution adopted by AFS and T&RI in December 1956:

"BE IT RESOLVED that the Society shall contribute such sums of money and property as the Board of Directors of the Society shall from time to time determine upon and as shall be usable by the Institute; that the Society and the Institute may solicit contributions of money and materials for the purposes of the Institute; and that the Society *by approval of the Board* shall construct, equip and make available to the Institute an adequate building for the purpose of establishing and maintaining its program."

Discussion then was requested by the President and all present urged to participate and to seek clarification on any phase of the overall T&RI program. He expressed the hope that all had studied the extensive, detailed brochure mailed by the Central Office well in advance of the meeting, prepared in view of

the fact that relatively few of the present Directors had to date had an opportunity to examine the program in detail.

Following luncheon served in the AFS Library, discussion continued until 4:00 p.m. The President then pointed out the necessity of returning to the Palmer House for dinner and stated that he would entertain a motion to table action on the T&RI recommendations and would call for a vote as the first order of business at the regular Board meeting the following morning. Motion to table was made, seconded and carried.

Adjourned Special Meeting of the Board of Directors

Palmer House, Chicago — Wed. Feb. 11, 1959

President Durdin called the meeting to order and announced that Roll Call would be dispensed with, but welcomed the arrival of Director J. H. King and Director Nominee R. E. Mittlestead, both delayed on attending by atrocious weather.

The President reviewed briefly the tenor of discussions held the previous day and requested action on the recommendations of the T&RI Trustees to proceed with immediate construction of the proposed Foundry Training Center building.

Further T&RI Discussion

Several Directors stated that insufficient publicity had been issued to date on the overall T&RI program, with the result that foundry top management was not aware of the basic purposes, need and plans for the proposed building. The General Manager stated that this had been done in all good faith so as not to embarrass the Board with apparent announcement of a project still to be approved. He stated that the numerous questions developed at the previous day's meeting would be developed in Question and Answer form and made freely available to all concerned.

In response to query as to how it was proposed to solicit building contributions from the industry, the President stated that it was planned to hold a series of meetings with top management in foundry areas throughout the country and to develop local "working committees" for following up thereafter; that a series of articles would be printed in MODERN CASTINGS; and that all other available publicity channels would be utilized.

Action on T&RI Building

All present having been given opportunity for discussion, the following motion was made, seconded and carried unanimously:

> THAT the AFS Board authorizes construction of the Foundry Training Center building to proceed promptly in accordance with the "Estimated Timetable" presented, subject to Executive Committee approval of General Contractor bids.

Following passage of the above motion, the Staff was urged to publicize strongly the entire T&RI program. The General Manager agreed, and pointed out that fullest promotion of the Institute and its program necessarily included stressing the Board's progressive action authorizing construction of the Foundry Training Center building.

There being no further business in reference to the Training and Research Institute, the Special Meeting of the Board of Directors was declared adjourned.

Respectfully submitted,
ASHLEY B. SINNETT
Secretary

APPROVED:
L. H. DURDIN, *President*
April 7, 1959

Minutes
Meeting of AFS Board of Directors
Palmer House, Chicago — Feb. 11, 1959

ROLL CALL: President Lewis H. Durdin, *presiding*
Vice-President Charles E. Nelson
General Manager Wm. W. Maloney
Secretary Ashley B. Sinnett
Technical Director S. C. Massari
Treasurer Edw. R. May

Directors (exp. 1959)
Charles E. Drury
Roger W. Griswold
Herbert Heaton
Al V. Martens
Wm. D. McMillan
John H. King

Directors (exp. 1960)
Wm. D. Dunn
A. A. Hochrein
Karl L. Landgrebe, Jr.
Fred J. Pfarr
John R. Russo
Allen M. Slichter
Henry G. Stenberg

Directors (exp. 1961)
David W. Boyd
Richard R. Deas, Jr.
Jake Dee
Webb L. Kammerer
Hillard M. Patton
Clyde A. Sanders

Officer and Director Nominees (on invitation)
Vice-President Nominee Norman J. Dunbeck
Director Nominees:
Norman N. Amrhein
Donald L. Colwell
Robert E. Mittlestead
A. Jack Moore
Wm. H. Oliver
James N. Wessel

Absent: Directors Harry W. Dietert (exp. 1959); Garnet P. Phillips, ex-officio (exp. 1959); Thomas W. Curry (exp. 1961); Director Nominee Cecil N. King.

A quorum having been established, President Durdin called the meeting to order. He thanked the Directors for their fine attendance and of the Vice-President Nominee and Director Nominees. He urged Director Nominees to enter fully into the discussions, but stated that voting on any motions would necessarily be confined to incumbent officers and directors.

President Durdin announced that the Minutes of the Annual Meeting of the Board of Directors held Aug. 7-8, and the meeting of the Board of Awards held Dec. 9, had both been approved by letter ballot of the Board.

Report of the General Manager

The General Manager's report for the fiscal year to date included Membership, Chapters, Convention & Exhibit, MODERN CASTINGS, BUYERS DIRECTORY, Administrative and International Cooperation.

Membership held up well through November 30, reaching 13,000; as of December 31 it dropped to 12,875 or 5 below the June 30 total of 12,880. Major membership changes included loss of 3 Sustaining members and 12 Company memberships, net gain of 23 Personal memberships, and net loss of 133 Affiliate memberships. The three Sustaining members either canceled or converted to a lower class. Although 29 Company members canceled and six converted to Personal membership, 21 additional Company members have been recorded to date.

Chapters. No new Chapters were established during July-December. Regional Administration Meetings were held in all seven Regions, with a second such meeting scheduled March 12 in Region 7.

Convention and Exhibit. Report on the 1959 Engineered Castings Show indicated a Show smaller than contemplated. The Show is being held strictly to the four exhibit categories approved by the Board, with no "borderline" cases accepted. The General Manager commented that most foundrymen still are not sufficiently promotion minded to make the Engineered Castings Show a success on its present basis.

A fine technical program was assured not only foundrymen but also designers and castings buyers. All Divisions are pointing their programs toward the purposes of the Engineered Castings Show, resulting in outstanding selection of Convention papers.

MODERN CASTINGS. It was reported that the new "standard" size of MODERN CASTINGS has been universally popular, both in dimensions and number of pages. Reader inquiries continue at a high rate with some 24,855 inquiries received in 1958, 5730 more than in 1957. As with all magazines, MODERN CASTINGS ad-

vertising fell off sharply during 1958, more noticeably in the 7-month period July-January 1959, as a result of the declining business conditions.

International Cooperation. As agreed by the Board, official tours will be organized to attend the International Foundry Congress in Madrid, Spain. Two official AFS delegates must be appointed, he said, to attend this Congress personally in addition to the AFS European Representative.

Report of the Treasurer

The Treasurer's report noted that as of December 31 all major items of Income were less than forecasted. Income for the first six months totaled $95,896 or $41,276 under Forecast; Expense $169,477 or $23,400 under Forecast; Net Expense $17,875 over Forecast.

Training and Research Institute expense showed $28,000 below forecast on December 31. Development and Promotion included $5,000 for attorney fees in seeking tax exemption, an activity still in process. Research expenditures to December 20 totaled $9,336.67, which will not appear on the AFS statement until AFS makes the necessary cash transfer to T&RI as a contribution. Transfers of cash contributions are being made as T&RI needs arise, as previously agreed with the Trustees.

Report of the Technical Director

The Technical Director reported approximately 60 Convention papers received for 1959, with an additional 30 expected, resulting in a Technical Program compatible with the established standards of the AFS Castings Congress. Convention "Transactions" ase now included in each issue of MODERN CASTINGS, two such sections of 64 pages being published in the January and February issues. It is intended to publish practically all Convention papers in the magazine.

The Annual TRANSACTIONS for 1958 were mailed to all pre-publication orders as well as gratis copies to Company, Sustaining and Honorary Life members on request. A total of 936 copies were sold to date, and 671 copies given gratis, leaving 1239 copies for stock.

Sales of Special Publications to December 31 totaled $17,083, compared with $21,482 for the similar period a year ago. Lower sales are a reflection of prevailing business conditions, plus the fact that few new publications were produced in the past six months.

It was noted that technical affairs of the Society continue at an unusually high pitch, probably accented by business conditions creating an awareness that technological development is of primary importance. AFS technical activities involve the participation of over 630 men serving on 120 committees of ten technical divisions and nine general interest committees.

Director of the SH&AP program has given talks in some ten Chapters and has been successful in obtaining modifications of air pollution ordinances in Newark, N.J., and Plymouth, Ind. As a result of his work, the air pollution law of Chicago saw final passage as an equitable and reasonable law. In addition to his SH&AP duties, the Director is acting as Assistant Technical Director, rendering considerable service in conjunction with many technical committees.

Reports of Board Committees

Chapter Contacts Committee. Vice-President Nelson expressed thanks for the contacts made by Directors with their respective Chapters. He noted that practically all Chapters had been contacted, in addition to attending the Regional Administration Meetings. He urged that liaison between the National organization and the Chapters be continued as strongly as in the past.

Exhibits Promotion Committee. Chairman Sanders reported that his committee was unable to hold a meeting prior to the Board meeting in order to offer recommendations for Board consideration. He stated that considerable study had been made of the exhibit problem and several items were being prepared for discussion at the next meeting.

Report of F.E.F. Trustees. Director Dunn reported on the Trustees meeting held at Massachusetts Institute of Technology. An F.E.F. committee, he said, has recognized the effect on the scholarship program that might develop because of government action concerning scholarships at both the national and state levels. It was the decision to table future activity of this committee until such time as government policy has been published.

The committee on technical institutes, he reported, are not in favor of scholarship programs at this level. It is desirous of securing two-year certificate holders for industry rather than encouraging students to continue on a four-year college program. It was decided not to allocate funds specifically for scholarships at this level.

It was recommended by this committee, Director Dunn reported, that F.E.F. undertake a technical institute program at once, initially confined to two schools, in the east and central west. It was recommended that the eastern school be the Erie County Technical Institute, and that Michigan University, Penn State or Purdue be also considered.

It was reported that F.E.F. is publishing a booklet entitled "Your Career in Metal," to be distributed nationally.

International Representatives. The General Manager reported that Dr. A. B. Everest of the Mond Nickel Co., Ltd., London, has accepted appointment as the new AFS European Representative. A. R. Parkes of the "Foundry Trade Journal" has been named AFS European Technical Correspondent.

Regional Administration Minutes. The President announced that a meeting of the Executive Committee would be held March 19 for the express purpose of reviewing all minutes of Regional Administration Meetings and making recommendations to the Board of Directors.

Report of the Nominating Committee

For information of the Directors, Minutes of the Nominating Committee meeting held December 8 were read, announcing selection of officers and directors to be elected at the 1959 Convention.

Report of Board Nominating Committee

Vice-President Nelson, Chairman of the Board Nominating Committee, stated that the Board had approved by letter ballot the election of Donald L. Colwell, Vice-President, Apex Smelting Co., Cleveland, as a Director-at-large for a 3-year term.

Finance Committee Recommendations

President Durdin stated that it was necessary for the Society to obtain an additional $30,000 operating capital until such time as final billings for the 1960 Show are received. He asked for approval of a 60-day loan in the amount stated from the Harris Trust and Savings Bank, Chicago. On motion duly made, seconded and carried the Board authorized the Treasurer to act accordingly.

Purchase and Sale of Real Estate

The General Manager presented a proposal for the sale and purchase of certain properties adjacent to land procured in 1958 and adjoining the Headquarters building. It was the consensus that real estate today is a sound investment for the Society. After full discussion, on motion duly made, seconded and carried the Board authorized the proposed land purchase by transfer of cash funds accumulated in the Investment Securities account, without selling any present securities.

Obligations of AFS Officials

Vice-President Nelson and Vice-President Nominee Dunbeck requested Board discussion concerning responsibilities of the elected officers of the Society. It was their firm belief, unanimously supported by the Board, that the Executive Committee should function more frequently and that the officers of the Society should be allowed to serve as chief executives rather than as "traveling salesmen." It was agreed that this conception of official duties and responsibilities conforms to the intent of the By-Laws of the Society, and should be followed in the future. Messrs. Nelson and Dunbeck asked that this understanding be publicized to the Chapters and the industry so as to avoid misunderstandings.

Selection of Auditors

On motion duly made, seconded and carried the Board of Directors approved George V. Rountree & Co., Chicago, as auditors for the fiscal year 1958-59.

New AFS Trustee to F.E.F.

The General Manager announced that Director Kammerer had accepted appointment as the new Trustee to the Foundry Educational Foundation, as approved by the Board in May 1958, to serve during 1959-1961 and replacing A. V. Martens. Director Dunn continues as a Trustee for the term 1958-1960.

New Membership Class

The General Manager recommended establishment of a new membership class for those who have retired from the industry after many years of service and yet desire to maintain contact with the industry. Following discussion, it was agreed that individuals accepted for such memberships should be active members of the American Foundrymen's Society for twenty years minimum, and should not thereafter be actively engaged in the industry. The following resolution then was approved on motion duly made, seconded and carried:

THAT a Retirement or Service Life Membership of AFS be granted individuals who have retired from active service in the Castings Industry after holding AFS membership for 20 consecutive years, no membership dues to be charged thereafter.

Student Delegates to AFS Conventions

Director Slichter proposed that AFS consider a program to subsidize student delegates from Student Chapters to AFS Conventions. Following discussion, it was the consensus that the Chapters and the Foundry Educational Foundation are already carrying on such activities, and that no additional action need be taken by AFS at this time.

Student Chapter Possibilities

The Board was polled on possibilities for new Student Chapters in their respective areas. Director Stenberg, commenting on increasing foundry interest at Wentworth Institute, Boston, agreed to explore a Student Chapter there. President Durdin asked that all Regional Vice-Presidents study Student Chapter possibilities within their areas and report back to the Board at a later date.

1961 and 1963 Convention Cities

The General Manager reported that, following discussions with the New England Chapter and the Boston Convention Bureau, it was agreed that the new Boston Convention Hall would not be finished in time for AFS to hold an Engineered Castings Show and Castings Congress in that city in May 1961. He stated that action had been immediately taken to cover proper dates in May 1961 at San Francisco, and at Boston in 1963. He reported that both the Boston Convention Bureau and the New England Chapter preferred to wait until 1963 to do the job properly.

Following Board discussion, the following motion was duly made, seconded and carried:

THAT the 1961 AFS Castings Congress and Engineered Castings Show be held in San Francisco instead of Boston, and that the 1963 Convention and Exhibit be held in Boston in the new Municipal Auditorium now under construction.

Selection of 1964 Convention City

The General Manager stated that Atlantic City and Chicago were both being considered for a 1964 Convention and Exhibit site. He explained that no exhibit policy had as yet been published by Chicago's Exposition Authority for the new Chicago Auditorium, but that occupancy rates probably would approximate twice those in other major convention cities such as Cleveland and Atlantic City. He pointed out the undesirability of confirming dates in a hall still under construction with no established rate policy, and therefore recommended Atlantic City as the 1964 Convention city.

A number of Directors finding it necessary to leave the meeting because of pressing travel plans and the late hour, a quorum was no longer possible. Selection of the 1964 Convention city was therefore tabled.

Announcement of Next Board Meeting

The President announced that the next meeting of the Board of Directors would be held at the Sherman Hotel, Chicago, on April 14, 1959, during the AFS Castings Congress and Engineered Castings Show.

There being no further business, the meeting was declared adjourned.

Respectfully submitted,
ASHLEY B. SINNETT
Secretary, AFS

APPROVED:
LEWIS H. DURDIN, *President*
April 6, 1959

Minutes

Meeting of AFS Executive Committee

Statler Hotel, Cleveland — March 19, 1959

ROLL CALL: AFS Vice-President C. E. Nelson, *presiding*
R. W. Griswold, Regnl. Vice-Pres., Region 2
Fred Pfarr, Regnl. Vice-Pres., Region 3
Karl L. Landgrebe, Regnl. Vice-Pres., Region 6
T. W. Curry, AFS Director
A. V. Martens, AFS Director
C. A. Sanders, AFS Director
A. B. Sinnett, AFS Secretary

Absent: L. H. Durdin, AFS President; H. Stenberg, Regnl. Vice-Pres., Region 1; C. E. Drury, Regnl. Vice-Pres., Region 4; A. M. Slichter, Regnl. Vice-Pres., Region 5; J. R. Russo, Regnl., Vice-Pres., Region 7.

Vice-President Nelson thanked the Regional Vice-Presidents, for their attendance and suggested that all Directors present at the morning meeting on Membership Analysis remain for the discussions on Regional Administration Meetings. He explained that he was acting on the request and in the absence of President Durdin, attending the Texas Regional in San Antonio. The Chairman stated that the purpose of the meeting was to study all Regional Aministration Meeting minutes for recommendations to the Board.

Conduct of Regional Administration Meetings

After complete discussion on the value and merits of the Regional Administration Meetings, conducted for the past two years in all AFS Regions, it was unanimously recommended,

THAT Regional Administration Meetings should be conducted in the future on the same basis as in the past two years.

It was suggested that the Chapter Officers Conference should give particular emphasis to the importance of Regional meetings in an effort to maintain good attendance.

Management Interest in AFS

Following discussion and review of all minutes, it was unanimously agreed that AFS should develop and follow through a definite program to create more management interest in the Society and its activities. It was stated that until the management interest is obtained, membership of technical operating people, who form the backbone of the AFS membership, cannot be obtained. It was suggested that AFS survey and Chapters and, as a beginning, adopt the successful procedures that have been used by the Chapters so far.

Regional T&RI Courses

Considerable discussion was reported in many Regional meetings on the holding of Regional courses by the Training and Research Institute. It was the consensus that the membership, and people availing themselves of the T&RI courses, cannot obtain fullest value from T&RI courses until a central program is developed and put into effect. Coupled with the central program for T&RI, it was the unanimous recommendation of the Executive Committee:

THAT T&RI continue to conduct Regional courses in Chapter areas.

Broadened Research Program

Vice-President Nelson reported many references in the seven Regional meetings for a broadened research program than is currently being conducted by AFS. All agreed that AFS should

expand their Research program, it being commented that current research funds were not being expended, due to some degree to lack of projects submitted by the divisions. The following recommendation resulted:

THAT a committee of the Board be appointed to consider long-range research interest and stimulate research projects, to be conducted by AFS Divisions under the present organization.

To assist this work it was suggested that Company and Sustaining members of the Society be solicited for suggested Research projects for T&RI consideration.

As a means of developing greater technical recognition of AFS, it was suggested that greater cooperation be established with other technical societies in the same area wherever possible.

Invoicing Member Dues

Most of the Regional meetings discussed the initial invoicing of member dues. It was felt that invoice mailings contain too much literature not related to the primary purpose of the invoice. As a result of discussion it was recommended:

THAT initial membership dues invoices be mailed separately with only a small card explaining the merits of membership in the Society, and that careful study be made to determine whether or not effective. If proven not effective, then invoicing should return to the present method of including several stuffers with initial invoices.

Communications With the Membership

Vice-President Nelson and other Directors commented that many members do not know the full value of membership and AFS services available; that communications between the Central Office and the membership are inadequate. It was agreed that planned programs in the Chapters should be initiated by the Regional Vice-Presidents, the Directors and Staff personnel through direct contacts at Chapter meetings.

The Secretary outlined plans currently in effect for developing closer communications, including a strip film being developed by T&RI to be shown before management groups and Chapters throughout the country. It was the consensus that this would have a double-barrelled effect in that both AFS and membership can be sold to management at the same time. The Executive Committee expressed enthusiasm over this proposed program and felt it would do much to bridge the present gap of communications between membership and management.

Local Publicity

Vice-President Nelson presented a recommendation from Director Slichter for closer contacts with local newspapers in Chapter communities. It was felt that continual liaison between the Chapter Reporter and local newspapers would result in better public acceptance of the industry, and at the same time develop better industrial recognition of the foundry industry and the local Chapter.

Self-Appraisal Committee

Question was raised as to action by the Board Self-Appraisal Committee. Vice-President Nelson reported that no action or work had been done to date, but accepted the following recommendation:

THAT the Self-Appraisal Committee be appointed and commence studies as soon as possible.

It was the consensus of the Executive Committee that the work of this committee could go far toward solving some of the problems confronting the Society.

There being no further business, Chairman Nelson declared the meeting adjourned.

Respectfully submitted,
A. B. SINNETT
Secretary

APPROVED:
C. E. NELSON, *Chairman*
April 8, 1959

Minutes
Final Meeting of AFS Board of Directors 1958-59
Sherman Hotel, Chicago — April 14, 1959

ROLL CALL: President L. H. Durdin, *presiding*
Vice-President C. E. Nelson

Directors (Exp. 1959)
H. W. Dietert
C. E. Drury
R. W. Griswold
J. H. King
A. V. Martens
W. D. McMillan
G. R. Rusk

Directors (Exp. 1960)
Wm. D. Dunn
A. A. Hochrein
K. L. Landgrebe, Jr.
F. J. Pfarr
A. M. Slichter
H. G. Stenberg

Directors (Exp. 1961)
D. W. Boyd
J. Dee
T. W. Curry
W. L. Kammerer
H. M. Patton
C. A. Sanders

Incoming Directors (Exp. 1962)
N. N. Amrhein
D. L. Colwell
A. J. Moore
Wm. H. Oliver
L. J. Pedicini
J. N. Wessel

Staff Officers: Wm. W. Maloney, General Manager
S. C. Massari, Technical Director
A. B. Sinnett, Secretary
E. R. May, Treasurer

Absent: Directors H. Heaton (exp. 1959), G. P. Phillips (exp. 1959), J. R. Russo (exp. 1960), R. R. Deas, Jr. (exp. 1961), Incoming Director R. E. Mittlestead.

President Durdin introduced the Training & Research Institute Trustees, who joined the Board for luncheon. He expressed pleasure and commendation for the excellent work they have accomplished to date and to the proposed expansion of T&RI activities.

President Durdin welcomed the newly elected Directors of the Society and invited all to participate in the Board discussions.

A quorum having been established, Minutes of the Special Board Meeting held February 10, 1959, were approved on motion duly made, seconded and carried.

Minutes of the Executive Committee Meeting held March 19, 1959 were read and approved on motion duly made, seconded and carried. The following recommendations were presented to the Board and approved as part of these minutes.

THAT Regional Administration Meetings should be conducted in the future on the same basis as the past two years.

THAT AFS and T&RI continue to conduct regional courses in Chapter areas.

THAT a committee of the Board be instituted to consider long-range Research interest and stimulate Research projects to be conducted by AFS Divisions under sponsorship of T&RI.

THAT initial dues invoice be mailed separately with card explaining merits of membership and that a study be made of effectiveness. If not effective, the invoicing should return to present method of mailings.

THAT a Self-Appraisal Committee of the Board be appointed and begin to operate.

Reports of Staff Officers

Interim reports of the General Manager, Secretary, Treasurer and Technical Director were presented and accepted on motions made, seconded and carried.

Reports of Board Committees

Membership Analysis Committee. Chairman Griswold presented minutes of a meeting held March 19, with the following recommendations; that there be no alteration of the present

dues structure; that there be a renewed effort to obtain more Company and Sustaining members; that dues refunds not be altered; that a master file of the committee's activities be kept in the Central Office.

Foundry Educational Foundation Trustees. Director Dunn, reporting on the annual meeting of F.E.F. Trustees held in March, briefed the following points: F.E.F. has reduced its budget without curtailing any activities; 241 scholarships were granted and 163 graduated at the end of the year; request for N.F.A. affiliation with F.E.F. was referred to the Executive Committee for further study; no definite arrangements made concerning a Technical Institute program, still under study; election of new F.E.F. officers for 1959-60; President, Frank G. Steinebach, Penton Publishing Co., Cleveland, Vice-President, R. B. Parker, American Brake Shoe Co., New York.

Exhibits Promotion Committee. Chairman Sanders reported his committee had been active, but no formal report was to be presented at this time. A joint meeting was requested with the Finance Committee on July 22, to completely present a program as worked out.

Report on Training & Research Institute

The General Manager briefed minutes of the Trustees meeting of April 14, reporting that detailed drawings and specifications were under way following a meeting of the Building Committee on April 3. The Building Committee: S. C. Massari, *Chairman;* F. C. Fluegge, International Harvester Co., Chicago; Axel H. Granath, National Engineering Co., Chicago; L. D. McDowell, American Steel Foundries, Chicago; R. A. Oster, Beloit Adult and Vocational School, Beloit, Wis. The Committee named Holmes & Fox of Des Plaines, Ill., as official architects for the project.

It was reported that a Course Advisory Committee has been appointed to assist the T&RI Director in selecting future courses to meet the industry's needs, as follows: H. H. Wilder, Vanadium Corp. of America, Chicago; T. E. Barlow, Eastern Clay Products Dept., International Minerals & Chemical Corp., Skokie, Ill.; J. L. Leach, University of Illinois, Urbana, Ill.; T. T. Lloyd, Albion Malleable Iron Co., Albion, Mich.; J. F. Wallace, Case Institute of Technology, Cleveland; and E. Trela, Apex Smelting Co., Cleveland.

The General Manager reported on the current tax status of T&RI, stating that the most recent action of the attorneys included filing a protest to a decision of the Department of Internal Revenue on March 19, 1959.

1964 Convention City

Staff recommendations for selection of 1964 Convention City were presented by the General Manager. However, in view of a forthcoming report for the Exhibits Promotion Committee, the Board, on motion duly made, seconded and carried, voted to table any decision until the Annual Board Meeting in August.

Recommendations of Retiring President

Before retiring from office, President Durdin expressed his sincere appreciation for the cooperation given him by the Board of Directors. For the incoming 1959-60 Board he offered the following recommendations for their consideration: Develop more and better publicity for the Society with an extended effort in Public Relations; the Executive Committees of the Divisions should be requested to develop more research activity; with evidence of the foundry industry lagging, a cooperative program with the foundry trade associations should be developed to put the industry back in its rightful place in the national economy.

President Durdin expressed his gratitude to all members of the Central Office staff for their cooperation and assistance.

Induction of New Officers

President Durdin welcomed the newly elected Directors to the AFS Board, and asked their cooperation in continuing to build the Society under the purposes and policies for which it was organized.

There being no further business to be considered, President Durdin declared the 1958-59 Board adjourned and turned the gavel over to incoming President Charles E. Nelson.

Respectfully submitted,
ASHLEY B. SINNETT
Secretary

APPROVED:
L. H. DURDIN, *President*
July 31, 1959

Minutes

Meeting of AFS Building Committee

AFS Headquarters, Des Plaines, Ill. — April 3, 1959

ROLL CALL: S. C. Massari, T&RI Director, *Presiding*
F. C. Fluegge
Axel Granath
R. A. Oster
Wm. W. Maloney, *Secretary*
Absent: L. D. McDowell

Director Massari reviewed the past history of the Training and Research Institute as well as the status of the present program for construction of an Institute building. The committee was informed that the Trustees and the AFS Board of Directors had authorized the selection of an architect and the preparation of detailed drawings and specifications followed by obtaining firm bids from reputable contractors.

The Building Committee was informed that it would be responsible for the final approval of design, construction details and selection of materials.

It was questioned whether or not plans for the proposed building are subject to approval by the State of Illinois since it is to be used for school purposes. The Des Plaines building commissioner was promptly contacted and advised that such action was largely perfunctory as long as the City of Des Plaines approves the design and construction of the building, but that four copies of the building plans would have to be prepared and sent to the State fire commissioner. Final approval by that office, he said, is also largely perfunctory.

It was suggested that in all probability equipment for fume and particulate matter removal would be subject to State approval and should be in full compliance with recommendations of the AFS Safety, Hygiene and Air Pollution Control Committee.

The Committee unanimously selected the firm of Holmes & Fox, Des Plaines, Ill., as architects for the T&RI building and agreed that they should be advised to proceed with preparation of detailed drawings and specifications. In the afternoon the committee discussed several matters relating to plans with Messrs. Holmes and Fox present. It was suggested that the architects consider two bridge cranes of monorail type rather than a type hung from a single bridge, as a means of lowering the cost for this particular equipment. It was further suggested that the architects contact Dean Scott, Lowden Machine Co., Chicago; and Abe Wright of Chicago Tramrail Co., to further pursue this matter.

Other actions and recommendations of the Committee were:

(a) The floor slab should be designed to carry a load of 200 pounds per square foot.

(b) It was also suggested that the architects study further the location and arrangement of the elevator so that it will serve both levels in the building rather than the lower level only.

Due to the inability of L. D. McDowell to attend the April 3 meeting, he subsequently met with the Director at AFS Headquarters on April 8. Briefed on the status of the project, he approved the selection of Holmes & Fox as architects.

The Committee was supplied with drawings of the building as prepared to date with the request that they give them further critical consideration, any major changes involved to be informed to Director Massari within a week or so that such information could be transmitted to the architect.

The Committee agreed that either Thursdays or Fridays

would be best for future Committee meetings, with the exception of the last Thursday in each month.

The Committee was advised that as progressive plans and specifications are developed by the architects, copies will be sent to the Committee members several days in advance of any contemplated meeting so as to afford them time for careful review and to come prepared to discuss them with the architect.

It was agreed that the next meeting of the Committee would be called when the architects have prepared sufficient material to justify Committee consideration.

There being no further business to be considered, the meeting was adjourned.

Respectfully submitted,
S. C. MASSARI
Chairman

Minutes

Meeting of AFS Technical Council

Hotel Sherman, Chicago — May 22, 1959

ROLL CALL: Chairman H. H. Wilder, *presiding*
Vice-Chairman C. E. Nelson, AFS President

F. C. Bennett	S. Lipson
W. K. Bock	H. W. Lownie
R. A. Clark	D. Matter
R. A. Colton	J. G. Mezoff
R. F. Dalton	R. L. Olson
R. P. Dunn	L. J. Pedicini
R. B. Fischer	V. M. Rowell
G. E. Garvey	R. W. Schroeder
H. G. Haines	J. Thomson
Harvey Henderson	J. Wallace
F. W. Jacobs	E. Welander
A. J. Kiesler	C. E. Westover
W. A. Koppi	(repr. L. H. Durdin)
J. M. Kreiner	E. C. Zirzow

Guests:
H. W. Gorman
W. E. Mason

Staff:
S. C. Massari, AFS Technical Director
R. E. Betterley, T&RI Training Supervisor
J. H. Schaum, Editor, MODERN CASTINGS
H. J. Weber, Director, Safety, Hygiene & Air Pollution

Absent: B. L. Bevis, D. L. Colwell, C. W. Gilchrist, R. J. Keeley, D. N. Rosenblatt, J. P. Holt, L. R. Jenkins, C. F. Joseph, L. W. Lehmann, K. M. Smith.

Reports of Chairmen

Committee activities of the past year and programs for the coming year were briefly described by the chairmen in order to keep each other informed and correlate all technical activities without duplication or overlapping of effort.

AFS-Sponsored Research

The Technical Director explained the method by which the Divisions may obtain research funds, and the Society's policy regarding research. Divisional research requests then were presented, as follows:

Gray Iron Division—requested renewal of a previous grant to continue and complete work on Gating & Risering. Enough information has already come from this work to justify a paper on "Feeding Distance of Cast Iron" for the 1960 Convention.

A Committee on Gating & Risering has recently been formed in the Gray Iron Division and intends to apply the results of the research at Case Institute to actual shop practice, in order to determine if the work can be field checked. For this purpose the Committee requested $2,500 to cover the cost of necessary radiography.

Light Metals Division—wishes to continue work at Battelle Institute on the freezing and feeding of light metals and requested $10,000 in the event that U.S. Ordnance funds are not available. Should the continuation of this work be underwritten by Ordnance, the findings will still be available to AFS.

Brass & Bronze Division—outlined the following program: (a) Continuation of the principal theme of present research; (b) investigation of closely related problems which would enhance the value of the present program; (c) investigation of new channels suggested by research to date. The Division requested $5,000 to determine thermal gradients in bars, plates and complex sections. A detailed proposal of future research, requiring an additional $4,500, was filed with the Technical Director.

Education Division—requested funds to produce a new motion picture entitled "The Place of Castings in Our Everyday Life." Script is already prepared, cost estimated at $5,000. It was pointed out that this is not research and that the money should not be taken from the Society's research funds.

Malleable Division—requested $8,500 to continue research at the University of Wisconsin on the casting of heavy section malleable castings.

Pattern Division—made no request for research funds.

Cupola Advisory Committee—intends to revise the book "The Cupola and Its Operation" in the next two years. In order that needed information may be incorporated in the revised edition, the committee requested $15,000 to cover research on "Gas Flow and Related Conditions in the Cupola."

Sand Division—reported that six of its eleven committees need a total of $15,800 to further their research work, the money to be used as follows: $1,000 for flasks, patterns, core plates, etc.; $1,500 to the Mold Surface Committee to evaluate mold surface; $500 for photographs needed for the revised edition of the "Castings Defects Handbook"; $300 to the Basic Concepts Committee for glass beads; $500 to Committee 8-J for supplies for a 5-day work session to determine the causes of veining; $12,000 to the Shell Mold & Core Committee to sponsor a project at a research institute in order to determine the causes and control of expansion.

Steel Division—requested $8,500 to continue research on snotters so as to complete final stages at the University of Michigan.

Die Castings & Permanent Mold Division—made no request for funds.

Ductile Iron Division—desires to sponsor a research project on "The effect of Specific Variables in Raw Material and Processing on the Eutectic Composition of Ductile Iron." The Division will prepare an estimate of cost in order to obtain approval on the project for the next fiscal year.

The 1960 Convention Program

The Gray Iron Division requested another session for 4 p.m. Tuesday.

The Malleable Division requested a third session for 4 p.m. Tuesday.

The Brass & Bronze Division requested the following changes in the program: (a) Eliminate sessions 1 and 2 on Monday; (b) Present a symposium jointly with the Light Metals Division from 2 p.m. to 5:30 p.m. on Monday; (c) Hold one session Tuesday afternoon after its luncheon meeting.

Program & Papers Committee Procedures

The Technical Director explained the procedures to be followed in obtaining Convention papers, and distributed copies of the Guide to Authors, Offer of Technical Paper and Review of Technical Paper.

Training & Research Institute

Messrs. Massari and Betterley reported briefly on courses given and scheduled in 1959. The status of the proposed T&RI building and plans for construction were described.

General Suggestions and Requests

It was suggested that session chairmen be given more publicity by supplying them with colored ribbons at the Convention. The Technical Director stated that ribbons were available at the 1957 Convention but that so few chairmen called for them it

was assumed they were not interested and the practice was dropped in 1958.

It was suggested that the National Committee Personnel roster should give a better explanation of how divisional officers are selected.

It was proposed that the Editor of MODERN CASTINGS write an article on "The Birth of an AFS Paper." It was agreed that the activities of the Gray Iron Division will be reported to the Editor for this purpose.

It was suggested that MODERN CASTINGS publish a list of all AFS Committees.

It was proposed that the National Committee Personnel roster appear in MODERN CASTINGS as a "bonus" section.

It was suggested that some educational institution should be retained to inform AFS where castings research is needed and what problems should be explored.

Remarks of President Nelson

President Nelson assured the Council that the National Officers will carefully consider the suggestions made and that the matter of publicity will be given appropriate attention. He stated that AFS does not do enough research, but that research requires funds which cannot be made available unless the research plan is well detailed and outlined.

The Board wants more ideas of good research projects, he said, preferably in the field of basic investigation, and projects should not follow divisional lines too closely. He stated that AFS spends $20,000 to $50,000 a year on research but that the actual benefit to AFS from committee work would run into millions. The President thanked the Council for its service to the Society.

There being no further business to be considered, the meeting was adjourned.

Respectfully submitted,
H. J. WEBER
Acting Assistant
Technical Director

Minutes

Annual Meeting of Trustees AFS Training & Research Institute

Union League Club, Chicago — Dec. 10, 1958

ROLL CALL: Chairman H. Bornstein, *presiding* (1957-1961)
Vice-Chairman C. E. Nelson (1958-1961)
H. W. Dietert (1957-1959)
L. H. Durdin (1957-1960)
I. R. Wagner (1957-1959)
B. C. Yearley (1957-1960)

Wm. W. Maloney, Secretary T&RI
S. C. Massari, Director T&RI
R. E. Betterley, Training Supvr. T&RI
A. B. Sinnett, AFS Secretary

Absent: R. A. Oster (1958-1962)

A quorum having been established, Chairman Bornstein called the meeting to order.

Tax Exemption Status

Chairman Bornstein and Messrs. Maloney and Massari reviewed the current status of the T&RI tax exemption appeal. They reported on a meeting held in Washington in September with officials of the Internal Revenue Dept., attended by Messrs. Bornstein, Durdin and Massari, and Messrs. Peterson and Cardon of Lee, Toomey & Kent, Washington attorneys handling the T&RI exemption case. It was stated that new Internal Revenue Dept. regulations governing tax-exempt organizations are in process but not yet completed, the AFS attorneys continuing their efforts to obtain proper tax exemption for the Institute.

Foundry Training Center Building

The Chairman reviewed development of plans to construct a building for T&RI on the present AFS site in Des Plaines. At the February 1957 meeting of the Trustees, he said it was decided to recommend deferral of building plans in order to evaluate the need. Trial courses during 1957 showed the desirability of such a building, on the AFS site.

At the May 1957 meeting of Trustees an expenditure of not to exceed $7,000 was authorized to employ an architect to prepare drawings and general specifications. These were presented at the December 11, 1957 meeting of the Trustees, and in the annual report to the AFS Board dated December 23, 1957, the Trustees recommended that the AFS Board be prepared to finance the project. The Chairman of Trustees appeared at the February 1958 meeting of the AFS Board and presented the recommendations of the trustees. At that time the AFS Board deferred action due to existing business conditions.

The Chairman requested all Trustees present to restate their attitudes on the need for a Foundry Training Center building. It was pointed out that it probably would take at least two years to complete the project, and that the industry's need for the building would then be even greater than at present. All agreed that the AFS Board should be urged to proceed immediately with construction plans.

The General Manager presented an estimated "Timetable" for construction of the Training Center and it was the consensus of the Trustees that this Timetable should accompany the request to the Board for construction of the building, as follows:

TIMETABLE OF DEVELOPMENT

Feb.	1959	— Final approval of design and plans.
March	1959	— Preparation of detailed specifications begun.
July	1959	— Approval of detailed specifications.
Aug.	1959	— Specifications mailed to selected general contractors.
Nov.	1959	— Deadline for general contract bids.
Nov.	1959	— Selection of general contractor.
Dec.	1959	— Arrange financing.
Jan.	1960	— Commence solicitation of contributions.
April	1960	— Commence building construction.
Oct.	1961	— Completion of building.
Nov.	1961	— First T&RI Courses in new building.

Accordingly, the following Resolution was moved, seconded and carried as a unanimous recommendation of the T&RI Trustees to the AFS Board of Directors:

Recommendation

The Trustees of the AFS Training and Research Institute believe that greatly increased demands for qualified personnel will be made by all industry in the next few years, and that the need for broader training of all occupational groups will steadily increase. The Trustees desire to point out that the proposed Training Center building is definitely needed by the Institute to broaden and improve its program. The T&RI Trustees therefore recommend that the AFS Board of Directors authorize construction of the proposed Training Center at the earliest possible date. It probably will take at least two years to complete the project after authorization is made by the AFS Board.

It was agreed that the Staff would revise the original T&RI brochure and mail it to all Directors at least two weeks in advance of the February Board meeting.

Review of T&RI Courses

Messrs. Massari and Betterley analyzed acceptance of training courses held during the past two years and presented the following cumulative data:

	1957	1958
No. of Courses presented	9	12
No. of Registered Students	403	368
Avg. Enrollment per Course	44.8	30.7
Avg. Miles Traveled per Student	833.6	610
No. of Regional Courses given	0	2

	1957	1958
No. of Courses canceled	0	4
No. of Companies represented	338*	226*
No. Foundries represented, %	89	81.4
No. of States represented	25	27
No. Students from outside U.S.A.	25	57
Total Companies represented to date		357**

*Some companies duplicated.
**All separate companies.

Courses for 1959

The Director presented a tentative schedule of 1959 Training Courses for approval, proposing 15 courses including Regional training courses at Birmingham, Ala., Hamilton, Ont. and consecutive courses at San Francisco and Los Angeles immediately preceding the California Regional Conference in March 1959. He stated that from experience gained with the 1957 and 1958 courses, four of the 1959 courses were being scheduled as 5-day events, 11 as 3-day events.

The need for sponsoring additional Regional courses was brought out and the fact that it is impracticable, under present limited Staff personnel, to hold more than four such courses annually. It was the consensus, however, that expansion of T&RI "field programs" is highly desirable when feasible. Betterley and Trustee Yearley reported on meetings with the U.S. Bureau of Apprenticeship and the Bureau's expressed interest in Regional Training Courses. They felt that the type, location and number of Regional courses might be affected by a forthcoming Bureau survey on foundry training needs.

Betterley presented data from the Bureau of Apprenticeship on 1956 and 1957 foundry surveys made with a National Foundry Advisory Committee, indicating that castings production is expected to increase strongly by 1960. Personnel needs to meet this increased production, he said, are expected to increase as follows:

Jobbing Plants: Engineers 30% increase, Technicians 15%, Service Group 12%, Operating Group 8%, Administrative 5%, Foremen 2%.

Production Plants: Technicians 34% increase, Administrative 26%, Engineers 22%, Service Group 22%, Operating Group 21%, Foremen 11%.

The Bureau's announced long-range program includes the objective "To encourage and promote wider participation in AFS Apprentice Contests and attendance at AFS Training and Research Institutes." The report of the National Foundry Advisory Committee also contains two significant points:

a) Repeatedly changing technology in castings production with a variety of alloys and customer demands for high quality have accentuated the need for highly skilled workers.
b) Off-job training should be used to supplement the on-job phase and increase the technical knowledge of all employees.

Course Advisory Committee

The Trustees discussed whether or not a Training Course Advisory Committee was needed to assist the Staff in developing future T&RI Courses. It was the consensus that such a committee might be helpful from the standpoint of industry-wide acceptance of course selection but that the committee should operate purely in an advisory capacity, with the Technical Director finally responsible for course schedules.

Institute Finances

The Treasurer presented financial statements for T&RI covering the period Dec. 1, 1957 to Nov. 20, 1958, showing excess Expense of $17,617 and including tuition Income of $21,630 and Expense items as follows:

Course Expense:		
Direct expense	$9,542.86	
Indirect expense	7,119.88	$16,661.74
Research expense		19,742.06
Foundry Instructors Seminar expense		2,849.11
Total Expense		$39,268.56

The financial statements also showed current assets of $17,058, Fixed assets $25,000, total assets $42,058; current liabilities $1,225, Fund Principal $40,832.

There being no further business to be considered, the meeting was adjourned.

Respectfully submitted,
WM. W. MALONEY
Secretary

APPROVED:
H. BORNSTEIN, *Chairman*
Training & Research Institute
Jan. 13, 1959

Minutes

Meeting of Trustees AFS Training and Research Institute

Sherman Hotel, Chicago — April 14, 1959

ROLL CALL: Chairman H. Bornstein, *Presiding*
H. W. Dietert
L. H. Durdin
C. E. Nelson
R. A. Oster
I. R. Wagner
B. C. Yearley
B. L. Bevis, Chairman AFS Education Division
N. J. Dunbeck, AFS Vice-President elect
Wm. W. Maloney, Secretary
S. C. Massari, Director
E. R. May, Treasurer
R. E. Betterley, Training Supervisor

The Secretary reported on action taken by the AFS Board of Directors in February authorizing completion of detailed drawings and specifications for the Institute building and subsequently taking firm bids from reputable contractors, the firm bids subject to final approval by the Executive Committee of the Board.

The status of courses given during 1959 was briefly reviewed and reference made to the statistical analyses covering courses given up to the present time.

A new strip film produced by the Associated Business Publications was shown as typical of the type of film intended to be produced for promotion of the T&RI program.

Director Massari announced appointment of a Course Advisory Committee to assist the staff in preparing future programs and specific course outlines for the year 1960. The Trustees were informed that President Durdin had appointed an AFS Building Committee for the purpose of assisting the Staff and architect in deciding details of design as well as construction materials for the proposed Institute building. The Director stated that the Committee on April 3 had selected as architects the firm of Holmes & Fox, Des Plaines, Ill. The Director stated that this Committee had reviewed drawings prepared to date and had agreed to submit their comments so that these in turn may be transmitted to the architects for inclusion in the final plans. Progressively, as plans and specifications are developed, the Committee will again meet with the architect, for progressive review.

The Secretary reported on the tax exemption status of the Training and Research Institute and the recent action of the Institute's attorneys in filing a protest to an adverse decision of the Department of Internal Revenue. Copies of this protest as prepared by Lee, Toomey & Kent and presented to the Internal Revenue Department on March 19, 1959, were supplied to all Trustees.

Financial statements for the 3-month period ending March 20, 1959, as prepared by the Treasurer, were provided to the Trustees for discussion.

By unanimous action of the Trustees, H. Bornstein was again elected as Chairman of the Trustees for the ensuing year, on motion duly made, seconded and carried.

There being no further business to be considered, the meeting was adjourned.

APPROVED:
H. BORNSTEIN, *Chairman*
April 25, 1959

Respectfully submitted,
S. C. MASSARI
Director

16th Annual AFS Chapter Officers Conference

AFS Headquarters and Sherman Hotel, Chicago — June 11-12, 1959

Group Buffet Breakfast (Sherman Hotel)
Group Conference Photo and Welcome ..... Pres. Nelson
Chapter Workshops — 1st Series
Group 1 — Technical Activities ..... Massari
Group 2 — "MC" Operations ..... Schaum-Eckert
Group 3 — Educational Programs ..... Betterley
Group 4 — Membership Services ..... Sinnett-May
Chapter Workshops — 2d Series
Buffet Luncheon
Chapter Workshops — 3d Series
Chapter Workshops — 4th Series
Evening Sessions (Sherman Hotel)
Conference Dinner
Guest Speaker — George Gilbert Groman
Subject: "How Other People Talk"
Special Delegate Meetings
June 12 (Sherman Hotel)
Group Buffet Breakfast
Chapter Programs that Work ..... *Panel:* Maloney-Massari
Organizing the Regional Conference ..... Maloney
Regional Administration Meetings ..... Vice-Pres. Dunbeck
AFS Conventions & Exhibits ..... Sinnett-Davis
SH&AP Program Developments ..... Weber
Good Chapter Administration ..... Maloney
The A-B-C of AFS Finances ..... Maloney-May
Luncheon
AFS Training & Research Institute ..... Massari-Betterley
The Shakeout ..... Pres. Nelson

ATTENDANCE

AFS President & Conference Chairman — Charles E. Nelson, Technical Director, Magnesium Div., Dow Chemical Co., Midland, Mich.
AFS Vice-President — Norman J. Dunbeck, Vice-President, International Minerals & Chemical Corp., Skokie, Ill.
AFS General Manager — Wm. W. Maloney

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William D. Dunn, Asst. to Pres., Oberdorfer Foundries, Inc., Syracuse, N. Y.
Webb L. Kammerer, Pres., Midvale Mining & Mfg. Co., St. Louis.
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A. Jack Moore, Exec. Vice-Pres., Canadian Bronze Ltd., Montreal, Que.
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Fred J. Pfarr, Mgr., Lake City Malleable Co., Cleveland.
John R. Russo, Pres., Russo Foundry Equipment Co., Oakland, Calif.
Clyde A. Sanders, Vice-Pres., American Colloid Co., Skokie, Ill.
James N. Wessel, Supv. Mtls. Engr., Puget Sound Naval Shipyard, Bremerton, Wash.

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Secretary Ralph A. Petersen, Dist. Mgr., Pangborn Corp., Birmingham, Ala.
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Educational Chairman George E. Steimle, Fdy. Apprentice Instr., Caterpillar Tractor Co., Peoria, Ill.

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FOUNDRIES CAN PRODUCE THEIR OWN CAST IRON DIRECTLY FROM ORE

By H. W. Lownie, Jr. and A. J. Stone

ABSTRACT

Foundries might increase their profits by using some recently developed direct-reduction processes for producing their iron. Some of these processes are suitable for integration into a foundry to replace the conventional cupola facilities and other melting auxiliaries. Under some conditions which exist in iron foundries today, these processes can be used to produce and melt iron more cheaply than by the present conventional methods. The use of direct-reduction processes would also reduce the foundry's dependence on outside sources of pig iron, scrap and coke.

The background of direct-reduction processes is given. Typical processes are described, and their advantages and disadvantages are discussed. Costs of producing iron by direct-reduction are summarized, and are compared with those involved in melting in a cupola.

DEVELOPMENT OF DIRECT-REDUCTION PROCESSES FOR IRON ORE

Recently developed processes for making iron from ore offer to foundries melting 100 tons of iron or more per day the opportunity to reduce their cost of iron and their dependence on outside sources of pig iron and scrap. This paper describes general aspects of these processes and shows how the processes could fit into foundry operations.

The blast furnace has been for many years (and still is) the only commercially important producer of pig iron in the United States. Huge blast furnaces today produce pig iron for about $38 or $40 per gross ton.

Continuing growth in the number and total output of pig iron blast furnaces, however, requires the solution of a number of problems which have been arising to plague the operators of blast furnaces.

A new blast furnace must produce about 1,200 tons or more of iron per day to be competitive with existing furnaces. The furnace and its auxiliaries, including coke ovens, would cost about $40,000,000.

High-quality coke is as important to the operation of blast furnaces as it is to the operation of cupolas. Blast furnaces consumed about 69,000,000 tons of metallurgical coke in 1957 as compared to an estimated 2,500,000 tons in cupolas. Large amounts of relatively scarce coking coal are required for this coke. The diminishing supply of good coking coals has necessitated large investments for improved methods of mining, treating and blending coals.

In 1957, over one quarter of the iron ores used in the United States were imported because domestic direct-shipping (unbeneficiated) ores are being rapidly depleted. Expensive plants for the beneficiation of taconites and other ores low in iron content have been installed to assure future supplies of domestic ores.

Everything associated with the blast furnace, therefore, must be on a very large scale for economical operation. The need for units that can produce iron economically in smaller quantities using poorly coking coals has been the driving force behind the development of other reduction processes.

Although the term is neither properly descriptive nor accurate, these nonblast furnace processes today are being called "direct-reduction" processes. This paper discusses the use of direct-reduction processes in foundries. When using such a process the foundry would receive iron ore, not pig iron, as a raw material.

FOUNDRY PROBLEMS BACKGROUND IN THE MELTING OF IRON

In order to analyze the probable value of direct-reduction processes to the foundryman, the cost of producing molten iron of the desired composition ready for pouring into molds should be known for both conventional methods and direct-reduction methods. For the conventional cupola operation, a foundry can be expected to pay (in mid-1958) about $40 to $60 per ton for its metal charge (exclusive of alloys). Melting costs (conversion costs) including labor, coke, maintenance of equipment, depreciation of capital equipment and overhead can add another $15 to $50 per ton (depending upon the type of foundry and tonnage melted).

Thus, the cost of molten iron in the ladle is about $55 to $110 per ton by conventional cupola methods. If iron of the same composition and temperature might be produced by another method at lower cost, progressive foundries should be familiar with the

H. W. LOWNIE is Div. Chief and A. J. STONE is Asst. Div. Chief, Process Met. Rsch., Battelle Memorial Institute, Columbus, Ohio.

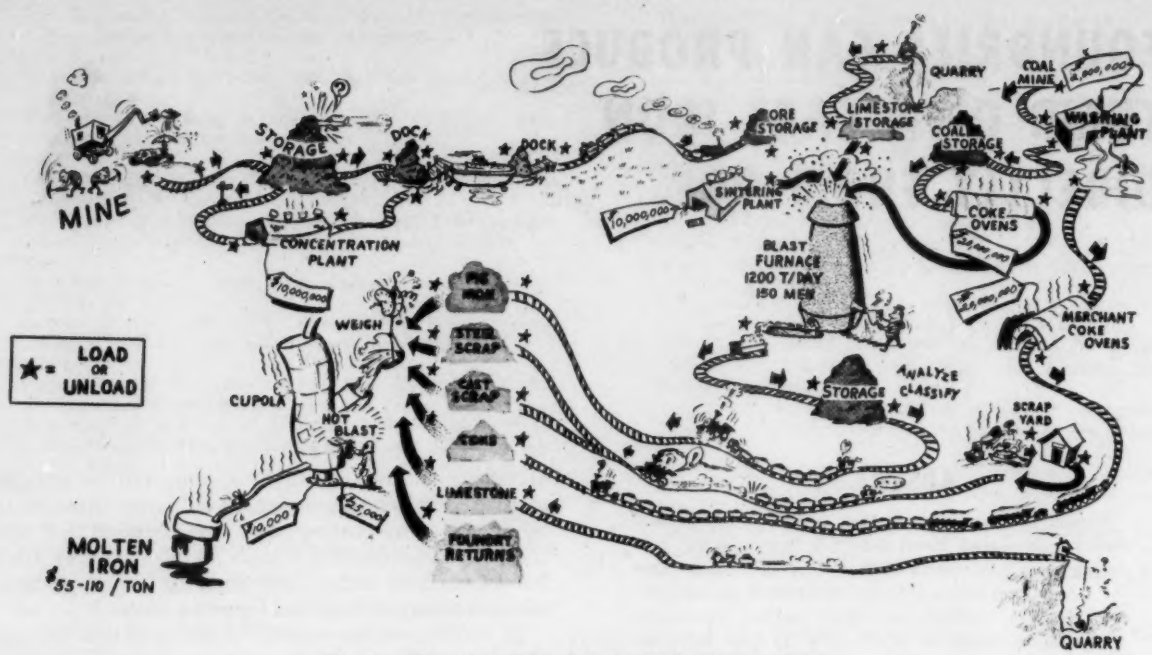


Fig. 1 — Present method of producing iron for castings.

other method so that they understand what their competitors are thinking about.

Direct-reduction processes can influence or eliminate other problems and situations which are inherent in conventional cupola melting in most foundries. Some of these are discussed below.

The prices of ferrous scrap fluctuate greatly. Although cycles appear prices are difficult to predict. Just as one example, consider what happened to the quoted price of No. 1 cupola cast iron scrap at Cleveland in 1957. It started the year at about $61 per gross ton, fell gradually to about $48 in April and May, turned up in June to reach $56 in July and August, then fell off to $38 in December. Steel scrap for cupola use varied even more. In contrast to this variable price of metallic scrap, the quoted price of iron ore remained fairly stable and consistent. Mesabi nonbessemer ore at Cleveland started 1957 at $10.85 per gross ton and finished the year at $11.45.

In the past five years (mid-1953 to mid-1958), the price of Mesabi nonbessemer iron ore has increased gradually from $9.90 to $11.45 in line with the gradual change in general economic conditions. This is an average change of about 15 per cent in 5 years, and always in the same direction in small steps. During the same period the price of cast scrap for cupolas varied over a range of almost 2 to 1, and this variation was erratic. Increased stability of the price of major raw materials for foundries would appear to be a definite advantage of direct-reduction processes.

The major sources of the ferrous metals used as raw materials by foundries fluctuate in availability in a manner which is certainly harmful to foundries. When the demand for castings is low pig iron is plentiful and ferrous scrap is abundant and cheap. When the demand for castings swings to a high level (as in time of war or threat of war) pig iron becomes hard to find in adequate amounts, sources of scrap dry up and the quality of foundry coke takes a nose dive. None of these materials, however, is needed in direct-reduction processes. The materials which are needed in the new processes remain at a more stable level of availability and cost.

Foundry coke is a problem of its own. It is expensive. Historically, quality has been variable. One of the obvious advantages of direct-reduction processes is that they require no foundry coke. Many of the processes require no coke and no coking coal.

The use of a direct-reduction process can make a foundry less dependent upon its traditional sources of melting stock. The foundry becomes more nearly an integrated producer.

Today, it appears that some new and some old processes are worthy of serious consideration of foundries. Foundrymen will ask, however, 1) can we make the kind of iron we want, and 2) how much will it cost?

In answer to the first question, virtually any useful composition of iron can be made by some direct-reduction process. Not every process will necessarily make every desired composition, but there is probably a process to make any desired type of iron. Some of the direct-reduction processes involve an electric furnace as the final stage. The proper operation of the electric furnace permits the production of irons with a wider range of compositions than is obtainable from a blast furnace or a cupola. Also, electric furnaces are noted for their ease of control as compared to a blast furnace or a cupola.

COMPARISON OF PRESENT PROCESS WITH DIRECT-REDUCTION PROCESSES

The conventional process used today for producing iron for castings is complex. It involves much han-

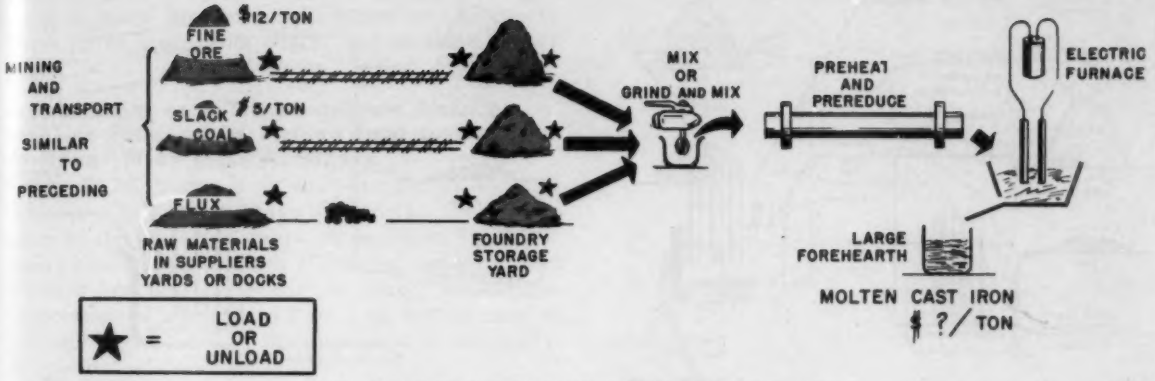


Fig. 2 — Proposed method of producing iron for castings.

dling, many people and much equipment. If one will stand back from the situation and take a look, the present situation looks something like that shown in Fig. 1.

The description of what is involved in a direct-reduction process has been deferred until later in this paper, but the elements of a typical process for use by a foundry are shown in Fig. 2. Certainly this process has some obvious advantages over the conventional method. Repeated loading and reloading of materials before they reach the foundry is minimized. Fewer agencies participate in providing raw materials, so fewer profits and markups are involved. Many miles of transportation costs are saved. Obviously, however, there are new costs incurred to offset some of these savings. The question is—how do the savings and new costs balance out against each other?

Because foundries are interested in getting molten metal, the situation on costs will be illustrated for direct-reduction processes which furnish molten metal from an electric furnace. In some types of direct-reduction processes other types of melting units might be used, but usually the electric-arc furnace has advantages which recommend it in this type of application.

To serve as a background for a discussion of costs, the general features of several types of direct-reduction processes will be described.

GENERAL DESCRIPTION OF DIRECT-REDUCTION PROCESSES

The simplest description of the direct-reduction processes of interest here is that they consist of two stages, 1) a pretreatment stage and 2) a melting stage.

Pretreatment usually involves preheating of the charge, plus complete or partial reduction of the iron oxide in the ore. Pretreatment stages usually are designed around one of the following types of equipment:

1) A rotary kiln.
2) A traveling-grate sintering machine.
3) A fluidized-bed reactor.

If reduction is completed during the pretreatment, the process is known as a "sponge-iron" process. Depending upon the quality of iron ore and reducing agent, the sponge iron may contain small or large amounts of impurities, slag and gangue. Sponge iron that is relatively low in nonmetallics may be melted much as scrap in a steelmaking furnace or as steel scrap in a cupola. Sponge iron relatively high in nonmetallics requires additional treatment to convert it to suitable melting stock.

The similarities of and differences between various processes will become clearer as five types of processes are described. These types are as follows:

1) Direct reduction in an electric furnace without thermal pretreatment of the charge.
2) Pretreatment in a rotary kiln, followed by melting (and some smelting) in an electric furnace.
3) Pretreatment on a traveling-grate sintering machine, followed by melting (and some smelting) in an electric furnace.
4) Reduction in a rotary kiln to produce some type of sponge iron which is melted in an electric furnace.
5) Reduction in a fluidized bed to produce some type of sponge iron which is melted in an electric furnace.

These types of processes will be described briefly to illustrate the variations which are possible in processing and in equipment. The intention of the descriptions is to give the foundryman a general understanding of the nature of processes of this type.

There is no intention to imply superiority of one process over another. A foundry must consider all of the processes and select the one best suited to its specific location, product and needs. This selection can require much study and analysis by people who are expert in the specific technology and economics of both the new processes and the needs of foundries.

The processes will be described on the basis of the authors' understanding of them. Some types of processes include several variations. All variations will not be discussed.

Direct Reduction in an Electric Furnace Without Thermal Pretreatment of Charge

Iron ore has been smelted in electric furnaces for many years. The most common furnace for this purpose is the Tysland-Hole furnace. About 30 Tysland-Hole furnaces are in use and about 20 more under

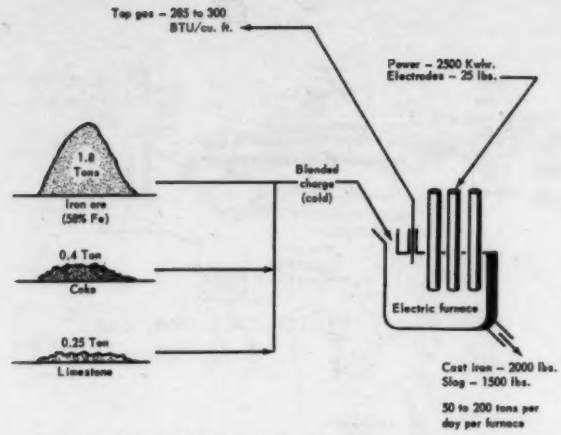


Fig. 3 — Simplified example of direct reduction in an electric furnace without thermal pretreatment of charge.

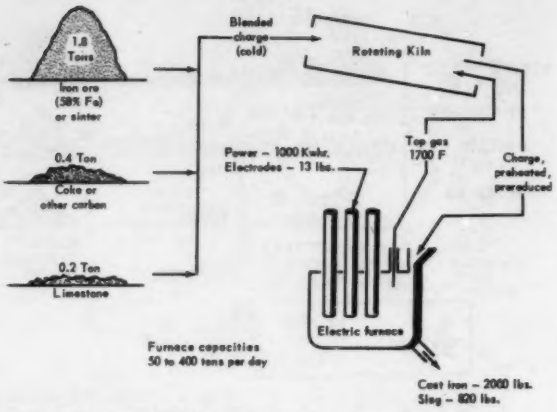


Fig. 4 — Simplified example of a process involving pretreatment in a rotary kiln followed by melting (and some smelting) in an electric furnace.

construction in Norway, Sweden, Finland, Italy, Japan, Switzerland, India, Peru, Spain, Venezuela and Yugoslavia. The largest furnaces developed so far produce about 200 tons of iron per day. These are powered by 33,000 kva of transformer capacity. The furnaces themselves are rated at 20,000 kw.

Cold ore and low-grade coke or char are charged directly into the furnace, as shown in Fig. 3. Smelting requires about 2,400-2,800 kwh per gross ton of iron. This is about 2½ times the power requirement for the newer processes which involve pretreatment of the charge. For this reason, this process is usually limited to locations where the following conditions occur simultaneously:

1) Electrical energy is cheap.
2) Metallurgical coke is relatively expensive or is not available.
3) Output is limited to a need for up to 200 or 800 tons of iron per day.

These conditions have not existed in the United States, so the process has not been used commerically here. The newer pretreatment processes have been developed to overcome some of the high cost of smelting ore directly. Because of the success of the new processes, it is unlikely that the smelting of untreated ore will receive serious attention for use in the United States. The Tysland-Hole electric furnace itself will undoubtedly be considered further because it has a long and good record of performance, but the furnace will usually be preceded by a pretreatment process in modern plants which will be considered for erection in the United States.

Cost calculations have indicated that in the United States it is probable that the cost of electric smelting of iron ore without pretreatment will be about $15 per ton of iron more than with pretreatment of the charge.

Pretreatment in a Rotary Kiln Followed by Melting (and some Smelting) in an Electric Furnace

Processes of this type are logical extensions of the simple smelting of untreated ore. Iron ore, fuel and reducing agent are fed to a rotating kiln, as shown in Fig. 4. Flux may be added with the charge to the kiln. Depending upon the method by which it is operated, the kiln can be used 1) only to preheat the charge, or 2) to preheat and partially reduce the charge. *The hot material from the kiln is discharged directly into the electric furnace* for the completion of reduction and for melting of the iron. Lime and coke or char may be added in the electric furnace. If the electric furnace is covered the rich gases resulting from reduction of the ore in the furnace are channeled back through the kiln to provide heat, and possibly some reduction.

By this type of operation, the power requirements per ton of iron in the electric furnace can be cut by 50 per cent or more. Furthermore, a particular furnace and transformer setup which will produce 200 tons of iron per day by direct treatment without preprocessing can produce 400 tons of iron per day when pretreatment of this type is added.

A distinguishing characteristic of processes of this type is the fact that the hot products from the kiln are dumped directly into the electric furnace. This saves the sensible heat in the product from the kiln. Processing between the kiln and the electric furnace is avoided.

Processes of this type are generally the most likely to fit smoothly into integration with the operations of a foundry.

The Elkem (Elektrokemisk) Process is the best known and probably the most completely developed of several processes of this type. The Stratmat Process resembles the Elkem Process.

The Orcarb Process is the same general type but makes good use of coking coal in the charge to the kiln. The coal is coked or charred in such a way that it bonds the ore and reduced iron into nodules, which are charged while hot into the electric furnace. The residual carbon in the nodules serves as the reducing agent in the electric furnace.

The rotary kiln performs two main functions in processes of this type. The kiln is used 1) to preheat the charge with relatively cheap fuel, and 2) to perform some of the reduction of the ore using heat from sources cheaper than electrical energy.

Depending upon the specific process, the degree of reduction of the iron ore which leaves the kiln can be expected to range from almost zero to about 70 per cent.

Pretreatment on a Traveling-Grate Sintering Machine Followed by Melting (and some Smelting) in an Electric Furnace

The DLM Process is the only one known to be of this type. It is quite similar in general principle to those in the preceding group. As shown in Fig. 5, the rotary kiln is replaced with a sintering machine. Fine ore and fine non-coking coal are mixed together in proportions of about 2 to 1 or 3 to 1. Flux may or may not be included in the mixture. The mixture is dampened and pelletized.

The pellets are fed to the grate of a continuous sintering machine where they are progressively 1) dried, 2) ignited and 3) carbonized. During this last step, the pellets are preheated to about 1500 F and about 60 per cent of the oxygen is removed from the ore. The hot pellets are charged directly to the electric furnace. As in the Orcarb Process, residual carbon in the pellets provides all or most of the reducing agent needed in the electric furnace.

It has been stated that the product from the electric furnace can be controlled to give a choice of 1) a pig iron containing 4 per cent of carbon, 2) a cast iron containing 2.5 to 3.8 per cent of carbon or 3) a steel containing 0.5 per cent or less of carbon. The contents of other elements such as silicon, manganese, phosphorus and sulfur can also be controlled to desired levels by 1) selection of the ore and coal, 2) control of the process and 3) operation of the electric furnace.

One feature of the DLM Process is that it requires no coke or coking coal. It can make effective use of very fine ore. Another feature is the high rate of throughput on the sintering machine. Residence time of the charge on the sintering machine is about 8 to 15 min. The processes which use a rotary kiln appear to require a residence time of about 1 to 4 hr to do the same job.

As with the processes in the previous classification, the traveling-grate process appears to be suitable for integration into a foundry operation.

The DLM Process is in the pilot-plant stage. It has not yet been used commercially, but is being considered for commercial applications.

Reduction in a Rotary Kiln to Produce Some Type of Sponge Iron Which Is Remelted in an Electric Furnace

The sponge-iron processes are characterized by the use of a processing step intermediate between the kiln and the electric furnace, as shown in Fig. 6. The Krupp-Renn Process is probably the best known of early processes. It has been used commercially in Europe, Asia and Africa.

The iron ore is reduced almost completely in the kiln. Coal, coke and/or a suitable gas is used in the kiln as 1) fuel and 2) reducing agent.

The product from the kiln is processed by mechanical, chemical or magnetic means to separate the reduced iron from gangue, ash, slag and other nonmetallics. Often this separation must be preceded by crushing and grinding to free the iron. Sometimes the separated iron is briquetted to simplify further handling and to avoid reoxidation. The final product from the extra processing step can be handled in any melting unit much as steel scrap would be handled.

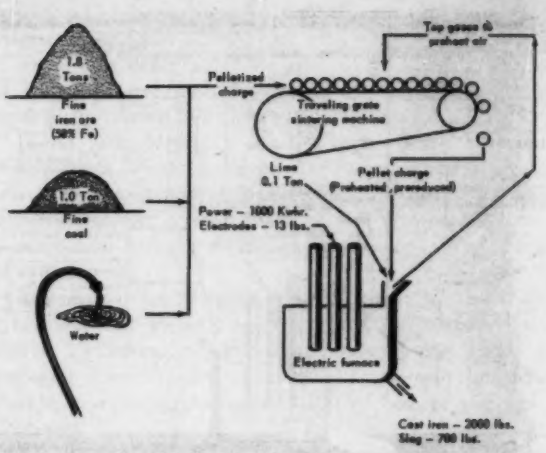


Fig. 5 — Simplified example of a process involving pretreatment on a traveling grate followed by melting (and some smelting) in an electric furnace.

When sponge iron is charged to a melting unit, the melting unit is not called upon to perform any reduction of the iron ore. (Recall that in each of the three preceding classifications, the melting unit was also called upon to perform all or part of the reduction of the ore.)

There appears to be no technical reason for promoting the integration of a sponge-iron process with an iron foundry. No sensible heat is retained from kiln to melting unit. The sponge iron can be made at a distant site and shipped to the foundry. There may be, in some situations, however, economic advantages to integration of a sponge-iron process with a foundry.

One early process of this type was developed by the U.S. Bureau of Mines. More recently, the Domnar-

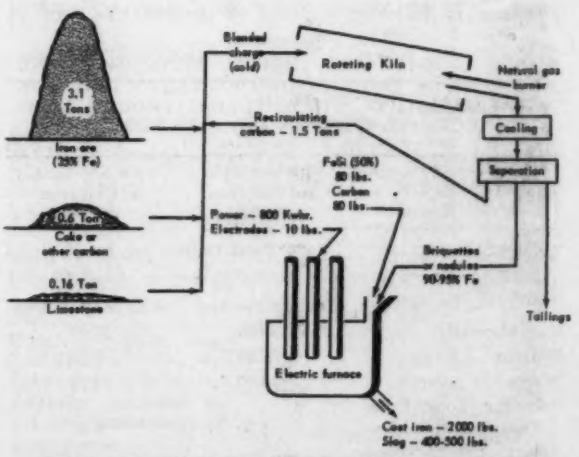


Fig. 6 — Simplified example of a process involving a rotary kiln to produce sponge iron which is melted in an electric furnace.

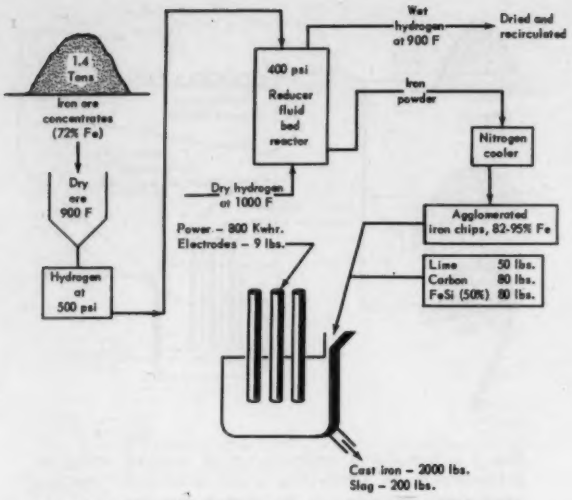


Fig. 7 — Simplified example of a process involving reduction in a fluidizer bed to produce sponge iron which is melted in an electric furnace.

fvet Process was developed in Sweden. The latest entry into the field is the R-N Process.

Reduction in a Fluidized Bed to Produce Some Type of Sponge Iron Which Is Remelted in an Electric Furnace

There are sponge-iron processes similar to those in the preceding classification, but use a modern fluidized-bed reactor instead of a rotary kiln. Whereas the rotary-kiln processes can make good use of ores of relatively low quality, generally the fluidized-bed processes require ores of very good quality.

Reduction of the ore and fluidization of the bed are accomplished by a gas (frequently hydrogen, carbon monoxide or dissociated natural gas).

The development of processes in this classification has been hastened by steel companies which are looking for cheaper sources of low-carbon iron for use in steelmaking. This interest by steel companies is a good indication of the potential of these processes.

The H-Iron Process of this general type has been thoroughly discussed in many printed articles. One company is building a plant to produce 50 tons of iron powder per day by the H-Iron Process. This iron will not be intended for melting stock, but will be sold for premium applications.

Figure 7 illustrates the authors' idea of how such a process might be used to produce sponge iron for use in a foundry. As in the previous classification, however, there appears to be only weak technical reasons for integrating such a process with the other operations of a foundry.

Some other processes known to utilize fluid-bed reactors and gaseous reductants include the Nu-Iron (U.S. Steel), Esso-Little, Stelling and Madaras Processes.

The Nu-Iron and Esso-Little Processes are believed to be in pilot-plant operation. The Madaras Process based on cracked natural gas has been (or is being) used to produce metal for a small Mexican steel company.

The situation on the cost of producing molten cast iron by direct reduction will be described on the basis of a process involving two integrated stages: 1) continuous preheating and some reduction in one piece of equipment, followed by 2) continuous smelting of the hot, partially reduced charge in an electric furnace to produce a common composition of gray iron.

TABLE 1 — ESTIMATED CAPITAL INVESTMENT AND NET COST OF PRODUCTION OF MOLTEN IRON FOR A PREREDUCTION AND SMELTING PROCESS INTEGRATED WITH AN IRON FOUNDRY

Cost	100 Net Tons of Cast Iron Per Day	400 Net Tons of Cast Iron Per Day
	Total Dollars for Plant	
Capital Investment	2,500,000	6,000,000
	Dollars Per Net Ton of Molten Iron	
Raw Materials	30.90	30.90
Electrical Energy	7.50	7.00
Wages and Salaries	16.00	4.20
Insurance, Local Taxes, Repairs, Incidentals ...	3.50	2.60
Amortization on Investment	3.60	2.10
Net Cost of Production..	61.50	46.80

COSTS

The costs which will be given are based on calculations made for several different processes of this general type. The costs purposely are presented in such a form as to avoid identification with any specific process. The specific processes differ technically in numerous ways, but differences between the processes in one application can be less than differences between the same process in various situations. There is an excellent chance that the actual cost of production of molten iron in a specific foundry may be materially less than the costs which are given here.

The cost estimates given here apply to plants and processes of a type and size suitable for integration with an iron foundry. If similar processes are used on a larger scale to produce pig iron for sale, the estimated costs will generally be somewhat lower.

The capacity of the smelting plant has a large influence on its capital cost and cost of production. Table 1 shows that a prereduction and smelting plant for 100 tons of iron per day can be expected to cost about $2,500,000, and a plant for 400 tons per day can be expected to cost about $6,000,000. The smaller plant will produce molten iron for about $61 per net ton. The larger plant will produce for about $47 per net ton.

Details of the cost estimates for the 400-ton plant are given in Table 2. These same bases were used to calculate the costs for other processes mentioned here.

How do these costs compare with the cost of melting in a cupola? That depends on how much iron you want and on how much you are now paying to melt in a cupola.

Before you make a comparison, be sure you know your present cost of melting and have included all

the items listed in Table 2. If your direct-labor rate is more or less than $2.50 per man-hour, correct Table 2 for this and other differences. Make sure that costs for amortization and taxes on your entire melting system (including building and storage yard) are included in your metal costs. Include power costs for blowers and charging equipment. Include at least part of the cost of your chemical laboratory.

Use delivered costs of raw materials, and include additions of silicon or manganese to the charge. Allow for melting losses. Figure the cost of your returns as equal to the costs of purchased No. 1 cupola cast scrap. If you have a water-cooled cupola, what does the water and pumping cost? If you have a hot-blast cupola, what does the fuel cost? Do not overlook the original and maintenance costs on instruments. Figure coke on overall consumption, not on your coke ratio per charge.

After you calculate your cost of producing a ton of iron ready to pour, your analysis might shape up something like one of those given in Table 3.

If the total cost of a net ton of molten iron from your cupola is calculated on the foregoing basis to be less than about $55, either 1) you are doing an unusually effective job of efficient and economical purchasing and melting or 2) you have overlooked some of the costs.

After your cost of melting in a cupola has been established, and Table 2 has been adjusted to your particular situation, you are then ready to use Table 4 (or some modification of it) to estimate whether you should give further consideration to one of the new processes.

If your analysis makes the new processes come even close to your present cost of melting in a cupola, the new processes deserve your further study. The costs given for the new processes are by no means the lowest which can be obtained under favorable conditions.

TABLE 2 – ESTIMATED CAPITAL INVESTMENT AND NET COST OF PRODUCTION OF MOLTEN IRON FOR A PREREDUCTION AND SMELTING PROCESS INTEGRATED WITH AN IRON FOUNDRY AND PRODUCING 400 NET TONS OF MOLTEN IRON PER DAY

Item	Basis	Cost	Subtotal or Total
Capacity	400 net tons of molten iron per day, 350 days/year, 24 hr/day, 140,000 net tons per year		
			Total Dollars for Plant
Capital Investment	Facilities for receiving, storage, grinding, mixing, preheating, and prereduction. One electric furnace at 33,000 kva. Pig machine, buildings, utilities, mixing ladles, cranes, laboratory. Construction, installation, and engineering. Trackage. Electrical substation.		6,000,000
			Dollars Per Net Ton of Product
Raw Materials			
Iron ore	$12 per GT delivered	20.20	
Coal	$ 6 per NT delivered	6.30	
Limestone	$ 3 per NT delivered	1.20	
FeMn	$245 per NT	0.90	
Electrodes	Prebaked carbon at $0.14 delivered	1.70	
Cooling water		0.40	
Refractories		0.20	
			30.90
Electrical Energy	Includes motors, smelting, and incidental. 7 mils per kwh		7.00
Wages and Salaries	Direct labor—54 men at $2.50/hr	2.70	
	Supervision—10% of direct labor	0.30	
	Laboratory	0.10	
	Indirect labor, payroll expenses, and overhead (40% of direct labor)	1.10	
			4.20
Insurance, Local Taxes	Insurance and taxes (2½% of capital)	1.10	
Repair, Incidentals	Miscellaneous	1.50	
			2.60
Amortization	20-year basis		2.10
Gas Credit	None taken. Potential credit of $1.15 per net ton		0.00
Total Cost of Production			46.80

SOME PROBLEMS INVOLVED IN DIRECT REDUCTION

The integration of a direct-reduction process with a foundry operation will not be without its problems. Each foundry must decide whether the potential advantages outweigh the problems.

The foundry will require a supply of iron ore of suitable quality and characteristics. Suitable arrangements and contracts will have to be made, but the supply of iron ore will be only a normal purchasing and scheduling problem. The foundry may find it advisable, however, to stock a reserve of iron ore in its own yard. The plant and labor force on which Table 2 is based includes facilities for grinding and screening ore and coal, if this is necessary. Facilities also include storage and handling of ore and coal.

The major problem which a foundry will face in integration is probably that of handling and scheduling molten metal. The electric smelting furnace should run continuously, certainly for 5 days a week, preferably 7 days a week and around the clock. Shut-

TABLE 3 – COST ANALYSIS

	Dollars Per Net Ton of Unalloyed Iron Melted in a Cupola and Ready to Pour		
	Small Foundry With Relatively High Costs	Large Foundry With Relatively Low Costs	Possible Average Foundry
Metal for cupola	61.00	37.00	40.00
Direct labor & Direct supervision	9.40	3.70	5.00
Coke	20.00	5.00	6.40
Supplies, repairs, misc. materials	7.60	1.60	3.10
Prorated share of general foundry overhead, including amortization of equipment	12.00	2.90	3.90
Conversion cost	49.00	13.20	18.40
Total Cost	110.00	50.20	58.40

TABLE 4 – ESTIMATED POTENTIAL SAVINGS AND RETURNS ON INVESTMENT FOR PREREDUCTION AND SMELTING PROCESSES INTEGATED WITH AN IRON FOUNDRY

100 Net Tons of Iron Per Day				
Cost of Iron Melted in a Cupola, dollars/net ton..	60	70	80	90
Cost of Iron Produced by New Process, dollars/net ton	62	62	62	62
Saving, dollars per net ton	none	8	18	28
Saving, dollars per year....	none	280,000	630,000	980,000
Annual Return on Investment (before Federal taxes), per cent of capital investment	none	11	25	39
400 Net Tons of Iron Per Day				
Cost of Iron Melted in a Cupola, dollars/net ton..	50	60	70	80
Cost of Iron Produced by New Process, dollars/net ton	47	47	47	47
Saving, dollars per net ton	3	13	23	33
Saving, dollars per year....	420,000	1,820,000	3,220,000	4,620,000
Annual Return on Investment (before Federal taxes), per cent of capital investment	7	30	53	77

downs over a week end may be feasible, shutdowns overnight probably will not be. In continuous operation, the furnace should require shutdowns for relining only once every year or two.

If the foundry operates on only one or two shifts, sufficient mixer (forehearth) capacity can be provided to store molten iron until the next shift starts. The equipment and labor in Table 2 are calculated for a plant in which the electric furnace operates 24 hr per day and the foundry pours metal 12 hr a day. With experience, a foundry might learn to vary the rate of output of the electric furnace to approach the foundry's demand for metal. It will almost certainly be necessary, however, for the process to include a forehearth to store metal to take care of a fluctuating demand by the foundry.

The plant should also be provided with a pigging machine to dispose of excess metal. Such equipment was included in the plant in Table 2. The disposition of this pigged metal will present about the same problem as the disposition of the foundry's return scrap.

The use of a process based on iron ore as a raw material can raise a question about what a foundry will do with its returns. Today, a foundry using the conventional cupola process will remelt its returns simply because this is the cheapest thing to do. With an electric furnace spewing out relatively cheap hot metal, the foundry will find that it has several outlets for returns. They can be melted in the electric furnace along with the ore charge, they can be sold, or part can be melted and part can be sold. Electric power for melting cast iron will probably cost about $4 per ton. With the ability to consume or market its returns as it sees fit, the foundry can turn periods of high scrap prices into another source of income.

As another alternative, the foundry might provide a second and smaller electric furnace for melting only. This furnace could be used for melting pigged metal or returns (both of which would be about the same composition), for superheating some molten metal, or for producing irons of special compositions.

CONCLUSION

These processes are available for foundries to produce their own iron directly from ore. Foundrymen owe thanks to many investigators who have, over a period of a century, developed the processes and put them into pilot plants (and, in a few cases, into commercial operation). The investigators were laying groundwork for an industrial change in the future. So far as some foundries are concerned, the future may be here today.

SOME FOUNDRY PROBLEMS IN THE DEVELOPMENT OF A NEW MARINE PROPELLER ALLOY

By A. J. Smith

ABSTRACT

The development of a casting practice for a new alloy is sometimes difficult because of the peculiar characteristics of the new material. This is especially true when the castings are large in size and weight such as large ship propellers. Experimental work must necessarily progress slowly. Some of the problems encountered by the foundry and their eventual solution are described. Material, melting, molding, refractories, fluxes, gas and feeding, are discussed as applied to nickel aluminum bronze. A comparison of engineering properties and service performances of the standard manganese bronze propeller alloy and the new superior nickel aluminum bronze is outlined.

INTRODUCTION

Shortly after World War II, the oil transportation industry began a program of building supertankers to replace the outmoded ships of the pre-war and war years. Shipyards throughout the United States were given the boost needed to revive an industry suffering from post-war palsy. These new high-powered ships were equipped with manganese bronze propellers which had long been recognized as the standard alloy for such marine applications.

After the ships had been in service for a few years the propellers exhibited more than the usual amount of wear. This condition was not entirely unexpected, and can readily be appreciated when a comparison of the requirements for the pre-war and post-war propellers is made. For example, a relatively high-powered, high-speed, pre-war tanker was equipped with a 6,000 hp turbo-electric engine driving its 19 ft 6 in. diameter propeller at a speed of 90 rpm. The post-war 28,000 ton tanker was equipped with a 15,000 hp turbine engine driving its 22 ft diameter propeller at a speed of 112 rpm.

Differences greater than these can also be demonstrated when comparisons are made between the pre-war reciprocating engine-driven vessels with propeller speeds of 65-80 rpm, and some of the modern passenger liners having propeller speeds as high as 155 rpm. The amount of water pushed by a propeller is directly proportional to the horsepower, which in turn is proportional to the amount of scrubbing action on the propeller surface.

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A program was initiated by the Central Technical Department of the author's company[1] to follow the wear characteristics of a large number of modern tanker propellers with varying lengths of service and under many conditions of operation. The results of this study (Fig. 1) establish the wear characteristics which could be expected from manganese bronze for such service.

As the propeller is an extremely costly item, and vital to the operation of the ship, it was obvious that a material with better wear characteristics was needed. The efficient operation of a ship means a minimum of tie-up time, and even a few unscheduled days in drydock can spell the difference between profit and loss for the operator.

A relatively small number of ships were driven by propellers made of other alloys such as aluminum bronze and stainless steel. For harbor craft, cast iron or steel was used, but they were for the most part unsatisfactory for large ocean-going vessels.

The development of a new propeller alloy pre-

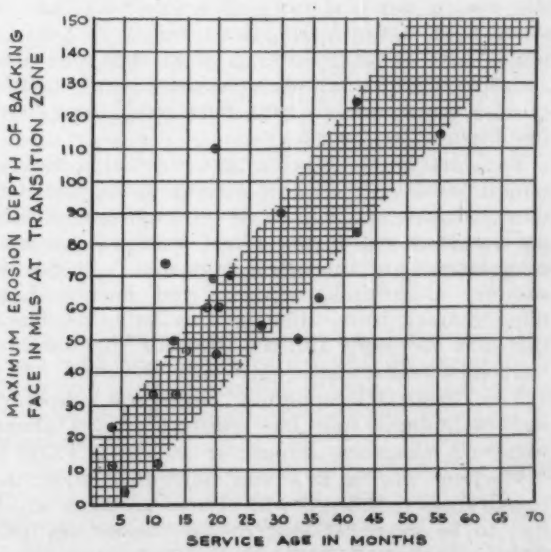


Fig. 1 — Corrosion vs. time plot of manganese bronze propellers on large single-screw vessels. Each point represents one vessel. Note that most of the values fall within the comparatively narrow shaded band.

sents a major problem to the foundry. Castings of this size and cost (the author's company's foundry produces propeller castings ranging from 6,000-120,000 lb rough weight) preclude any hasty manufacturing plans. The vital part which the propeller plays in the performance and safety of a ship would prevent both manufacturer and ship owner from jumping to a premature selection of an untried material. Earlier efforts had been made to confine the manganese bronze chemistry to a more limited range, and to make slight modifications in the composition. Modifications such as additions of nickel or tin to improve wear resistance and mechanical properties were made. Although some minor improvements in tensile properties and wear resistance were developed, it was felt that more definite improvements could be made with a new alloy.

It is the intent of this paper to present a general review of some of the more pertinent foundry problems encountered in developing a new propeller alloy designed to meet the requirements of the manufacturer, user and the various marine classification societies throughout the world.

EARLY INVESTIGATIONS

Prior to any metallurgical or foundry development work a cooperative research program was established between Bethlehem Steel Co., Shipbuilding Div. and the International Nickel Co.[2] to select a group of reputable alloys, and a method of comparing their wear resistance under laboratory conditions simulating propeller service. Because of the extensive and complex nature of this program, time and space will allow only a brief description of some of the general findings.

Several alloys chosen for preliminary comparative tests were selected on the basis of prior research, experience and a survey of the available literature.

The method for evaluating the wear characteristics was to spin flat cast, machined blocks in sea water at a nominal tip velocity of 27 ft per sec for 60 days. Although test specimen size, spin time and spin equipment have undergone some modification since these early runs, later results have been in substantial agreement.

The final choice of alloy was made after casting several 200-300 lb crucible furnace heats to determine handling qualities of the various materials under consideration. It was found that nickel aluminum bronze had the best combination of wear resistance, mechanical properties and foundry handling characteristics, although this latter consideration was far more difficult than for the standard manganese bronze alloy.

The chemical composition, mechanical properties and preliminary spin test results of nickel aluminum and manganese bronze are shown in Table 1.

Working with a new and unfamiliar alloy, the foundry was confronted with several problems which had to be solved at least in part before any production castings could be made. Such general subjects as raw material specifications, metallurgical control, melting, molding materials, risering and refractories were given primary consideration.

Production propeller heats are run in large oil-fired reverberatory melting furnaces having a combined capacity of 130,000 lb. However, it was felt that small oil-fired crucible furnaces would enable the author's company to develop the answers to some of the problems. No difficulty was experienced in melting in this latter type furnace, and progress was made in developing the properties of test specimens, influence of pouring temperature upon mechanical properties, use of fluxes, castability, etc.

It was found that reproducible mechanical properties could be obtained from test coupons made to A.S.T.M. B208-54 (Fig. 1 (a)*) the double keel block (Fig. 3*) and the double horizontal full-web type coupon.[3] Slightly better properties were obtained with the latter bar. Smaller gates than those used for manganese bronze were necessary to obtain maximum properties.

The influence of pouring temperature upon mechanical properties was found to be insignificant, as for manganese bronze. The results of a typical set of test data, which has since been reproduced several times, are shown in Table 2. It must be appreciated, however, that gas solubility increases with an increase of temperature and, in that respect temperature may influence the mechanical properties.

*Figures referred to are A.S.T.M. figures.

TABLE 1 — TEST RESULTS OF NICKEL ALUMINUM AND MANGANESE BRONZE

Item	Nickel Aluminum Bronze	Manganese Bronze
Chemical Composition, %		
Copper	78 — 81	55 — 60
Zinc	In "all others"	Remainder
Nickel	4.5 — 5.5	—
Iron	3.5 — 5.5	0.9 — 2.0
Manganese	0.5 — 1.0	0.3 — 0.9
Aluminum	9.0 — 10.3	0.7 — 1.0
Lead	0.01 max.	0.4 max.
Tin	In "all others"	1.5 max.
All Others	0.5 max.	—
Mechanical Properties (Normal Range)		
Yield	35-43,000 psi	27-33,000 psi
Tensile	80-95,000 psi	60-72,000 psi
% Elong. in 2 in.	15-30	20-35
Proof stress	28,000 psi	14-16,000 psi
Brinell hardness no.	152 — 190	112 — 130
Fatigue — air	21-25,000 psi	9-14,000 psi
Fatigue — salt water	18-22,000 psi	9-12,500 psi
Density — lb/cu in.	0.273	0.297
Spin Test		
Weight loss — grams	2.48	9.21
Loss — Mg/in.^{2}/day	6.5	24.0
Mg/dm^2/day	99.0	376.0
In. pentration/yr.	0.019	0.063
At tip — in.	0.005	0.012

Spin Test Data:
Nominal tip velocity — 27 ft/sec.
Exposed area — 6.5 sq in.
Average sea water temp.:
First 18 days — 14.7 C (58.5 F) Range: 9.0-19.5 C (50-66 F)
Last 42 days — 27.5 C (81.5 F) Range: 18.0-31.5 C (64-88 F)

NOTE: Yield measured by 0.5% offset method.

In general, the metal did not handle as easily as manganese bronze. There appeared to be less fluidity and a greater tendency toward dross (aluminum oxide) entrapment in the castings.

Shrinkage was somewhat greater than for manganese bronze, and it was apparent from the work done on the small castings that risering would require considerable care if sound castings were to be made. Insulating and exothermic materials used for riser sleeves and padding worked satisfactorily as for manganese bronze, and no difficulty was anticipated with these materials. Later, however, experience on production size castings discounted the results of this early experimental work.

Certain proprietary fluxes were found beneficial in reducing melting loss and in producing cleaner castings. Here, however, this experimental work did not aid in developing production practices, as fluxing in the reverberatory furnace proved to be an entirely different matter.

EXPERIMENTAL FURNACE WORK AS APPLIED TO PRODUCTION MELTING

The first production heats indicated that considerably more experimental work was in order before satisfactory castings could be made. It was evident that melting nickel aluminum bronze in crucible-type furnaces was not indicative of the results that could be obtained in the larger reverberatory furnaces, and that many operational problems remained to be solved. This fact led to a short but concentrated experimental and development program on melting in a scale model reverberatory furnace whose interior dimensions were reduced 6 to 1 from the average interior dimensions of the largest production furnace. An adjustable tilting oil burner was fitted to the firing end.

For the most part these developments on melting in the small furnace were applicable to production furnace operation and resulted in a comparatively inexpensive and rapid solution to many problems, some of which are discussed briefly below.

Burner Position. It was found that nickel aluminum bronze could be melted more rapidly when the burner flame impinged on the metal. The production furnaces were designed so that the burner flame would be quite far removed from the metal to prevent excessive zinc losses in manganese bronze. A new adjustable tilting burner was added to these furnaces to allow flexible operation of melting both alloys without further modifications. It is appreciated here that there exists a variety of burner locations for reverberatory furnaces from one foundry to another, and that such modifications are not necessary in all foundries. It can be generally stated, however, that furnaces equipped with burners close to the metal line will melt nickel aluminum bronze more efficiently than burners remote from this area. This is not as obvious as it might appear to be as, theoretically, melting in a reverberatory furnace is accomplished by direct contact with the gas and heat radiated from the refractory.

Combustion. Most of the results of the pilot-plant furnace study were not applicable to production heats. Although it was well appreciated that a neutral furnace atmosphere was most desirable for efficient melting, such a condition was almost impossible to maintain. Theoretically, a neutral or oxidizing atmosphere would prevent hydrogen absorption by the metal. Likewise, a reducing atmosphere would prevent oxidation of the aluminum resulting in the formation of dross.

However, it was found that combustion conditions were constantly changing during the melting period because of natural air leaks in the furnace. These fluctuations could be prevented only by extreme oxidizing or reducing atmospheres both of which were most undesirable. Good furnace practice required a careful check and manipulation of the burners to prevent excessive gas absorption or oxide formation. By such operation, which can be considered somewhat of an art, rapid meltdown and elimination of many combustion problems can be accomplished.

Fluxes. The primary functions of a flux are, 1) to react with the oxides present on the metal bath, 2) to prevent the formation of additional oxides and 3) to form a fluid slag to prevent gas absorption by the metal.

Slag covers are intended to perform the third function and in addition, for high zinc alloys, to prevent the volitilization of the zinc.

In discussing the subject of fluxes with people actively engaged in smelting and refining, a wide divergence of opinion on the use of flux at all or in the type of flux to use was found. None had

TABLE 2 – THE INFLUENCE OF POURING TEMPERATURES UPON THE MECHANICAL PROPERTIES OF NICKEL ALUMINUM BRONZE

Temp. F and (C)	Yield Strength psi	Tensile Strength psi	% Elon. in 2 in.
2260	36,200	91,550	27.0
(1238)	37,150	91,000	27.0
2200	38,150	91,300	26.5
(1204)	36,500	90,800	27.0
2150	38,800	90,950	23.5
(1177)	42,100	92,050	28.0
2100	39,750	91,950	26.5
(1149)	41,700	92,550	27.0
2050	36,950	93,350	28.0
(1121)	41,050	92,950	26.5
2000	36,550	92,850	26.5
(1093)	37,750	92,500	26.0
1950	39,200	89,500	19.5
(1066)	37,350	84,800	16.5
1940 (1060)	Froze in Ladle		

NOTE: Temperatures were taken at the mold and represent the metal temperature at the time it entered the test bar cavity. In pouring a propeller with test bars attached, the metal may be quite cold when it reaches the test bar cavity. Therefore, allowances must be made for this. Test coupons were of the A.S.T.M. 208-54 (Fig. 1a) double keel block test bar.

any experience in fluxing aluminum bronze in reverberatory furnaces as such. Ingot is always made in rotary-type or low-frequency induction furnaces. All agreed, however, that fluxes were essential when melting charges of dirty scrap.

The most effective fluxes for rotary furnace treatment of aluminum bronzes appear to be the fluorides (potassium titanium fluoride or anhydrous potassium fluoride) used singly or mixed. Sodium chloride and potassium chloride, borax salt mixes, sal-ammoniac and manganese sulfate are also used. Such compounds are generally the basis for some proprietary fluxes which had been used with marked success in the crucible furnaces, but with an equal lack of success in the reverberatory furnace.

Broken bottle glass has no actual fluxing action and is intended only as a protective cover. It is commonly used in foundries in this country and abroad to minimize loss of volatile elements, to reduce oxidation and to minimize gas absorption.

Charcoal is another material commonly used as a cover for aluminum bronze. It has the advantage of being a good deoxidizer for copper. Its chief disadvantages are in its faculty for introducing moisture into the metal if added to the molten charge, and in its ability to combine with the iron in the material if left in contact with the molten metal for any extended time.[4]

Perhaps one of the most objectionable characteristics of fluxes and covers is their ability to attack the furnace linings forming additional and difficult to control refractory slags. From the standpoint of the author's company no greater hindrance could be introduced into the reverberatory furnace melting practice.

Table 3 summarizes some of the pertinent properties of available fluxes. It will be noted from the table that fluorspar has a high melting point of 2480 F (1360 C). Since one of the reasons for a flux is to get an early liquid slag, it is evident that something else is required.

TABLE 3 – FLUX PROPERTIES

Material	Melt Point, F and (C)	Al_2O_3 "Solvent"	Erode Refractory	Silica "Solvent"
Fluorspar	2480 (1360)	Yes	Yes	Yes
Glass	1500 (816)	No	Some	Yes
Borax	1366 (741)	No	Yes	Yes
50% Borax, 50% Caustic Soda	Low	No	Yes	Yes
Caustic Soda	640 (338)	No	Yes	Yes
Potassium Fluoride	1616 (880)	Yes	Yes	Yes
Sodium Fluoride	1827 (997)	Yes	Yes	Yes
Cryolite	1832 (1000)	Yes	Yes	Yes

While the melting point of glass is about 1500 F (816 C) this is really a softening point, and it is felt that the glass would make a quite viscous slag until heated to well over this temperature.

Borax has a relatively low melting point of 1366 F (741 C), but does not have an active affinity for aluminum oxide. The same is true of caustic soda or 50 per cent borax – 50 per cent caustic soda.

It was felt that a flux composed of a borax base, with some quantity of potassium fluoride or sodium fluoride and cryolite, might be successful because the borax would furnish the low temperature fluidity for earlier cover while the fluorides would furnish the "solvent" action for aluminum oxide removal.

The conclusions, based upon experience in both the experimental and production furnaces, indicated that any and all types of fluxes tried to date were unsatisfactory. In general, fluxes appeared to perform in a satisfactory manner during the initial stages of melting, but then appeared to "dry out" and contribute to the general drossing condition in the furnace. This materially increased the meltdown time (by insulating the metal), decreased the yield and attacked the furnace walls.

The use of fluoride-type fluxes was found somewhat beneficial when added to the ladle during tap out. However, the benefits were counteracted by the time required to remove the fluid slag, the resultant loss in temperature and the annoyance and health hazard of pungent fumes from the fluorides.

GAS ABSORPTION

Nickel aluminum bronze, unlike manganese bronze, is susceptible to gas absorption during melting. Theoretically, this should only be true when the melting atmosphere is on the reducing side. As stated before, however, large furnace atmospheres are extremely difficult to control as combustion conditions are under constant fluctuation. In actual practice gas is picked up by the metal during a reducing period or from contaminants on the metal charge. It is widely accepted that the subject gas is hydrogen and, therefore, no attempt will be made in this discussion to develop the theory of gas absorption which is adequately covered by many texts.[5]

The average foundryman or metallurgist is usually not aware of the presence of gas in molten metal unless the metal spews or puffs in the riser during solidification. It is most common, however, to gas during melting and, depending upon the amount of gas and the size of the casting which in turn governs the rate of solidification, the gas may or may not be detrimental to the casting.

The question of how or why it is known that a heat of metal has been gassed is obviously in order. The laboratory has tested a large number of heats (over 150) melted in crucibles, experimental reverberatory, three production reverberatories, a rotary and a line frequency electric furnace and, without exception, all heats were gassed to some extent. It is true that the best furnace practices can minimize this condition, but in the author's experience it

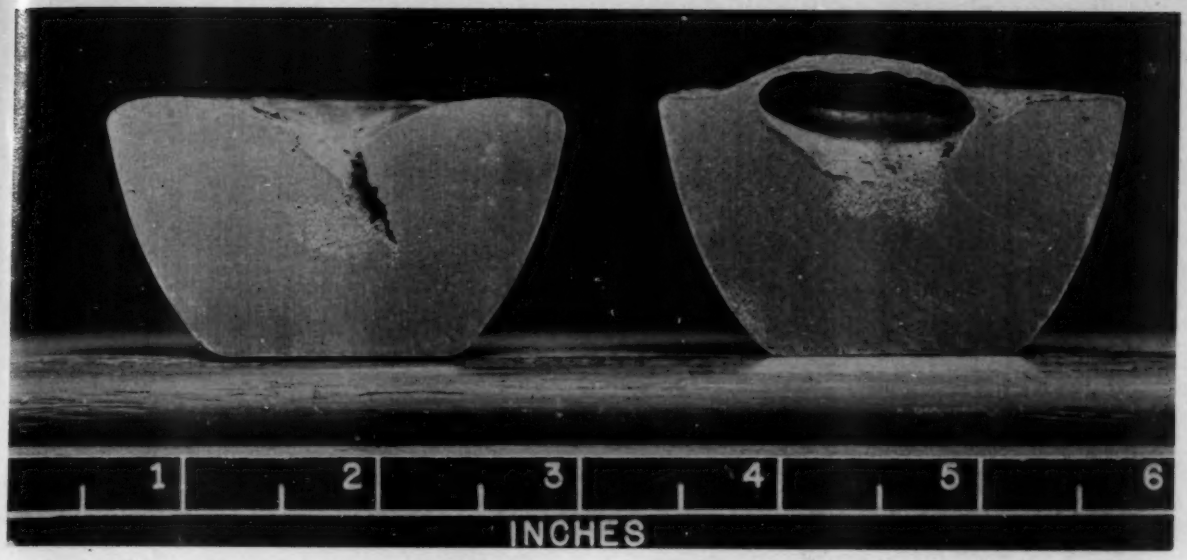


Fig. 2 — Two specimens from the same reverberatory furnace heat of nickel aluminum bronze. The specimen on the left was allowed to solidify under atmospheric pressure, while the specimen on the right solidified under vacuum.

cannot be completely eliminated unless melting is done under a vacuum. At the present time, it is economically infeasible to build vacuum melting equipment to melt large nickel aluminum bronze propeller heats.

The presence of gas in metal can be determined by a piece of equipment known as a vacuum tester, which has been described in detail by Walther, Adams and Taylor.[6] Although not commonly employed in brass and bronze foundries, it is becoming widely used in the aluminum casting field. It consists mainly of a vacuum pump and accessible vacuum chamber. The procedure for testing metal is to pour a sample either from the furnace or ladle into a dried and coated crucible, place it in the transparent vacuum chamber (usually a sealed bell jar), start the vacuum and observe the behavior of the specimen during solidification.

If gas is present the removal of atmospheric pressure down to 25-30 in. of mercury will allow the gas to come out of solution. The more gas present, the more violent the reaction, and the smaller the amount of vacuum required to effect the reaction. A control specimen from the same sample of molten metal handled in an identical manner, and allowed to solidify under atmospheric pressure unless heavily gassed, will generally show normal shrinkage. This is shown by Fig. 2.

The next question arises as to the influence of gas upon mechanical properties and casting quality. To answer this question a method had to be developed to degas the metal. It is known from experience that some castings have been poured from gassed metal wherein the gas had no noticeable influence upon casting quality or mechanical properties. It is assumed in this case that the gas was not present in sufficient quantity to visibly harm the casting. There are other cases, however, where the casting poured from gassed metal spewed out of the riser. The resultant casting contained large numbers of holes and mechanical properties were inferior to anticipated values. The problem was then to determine when a melt required degassing and how to remove the gas if removal were required.

To date no quantitative test for gas has been developed. If such a test should be developed later it will then be necessary to determine the tolerable gas content where a degassing procedure will not be necessary.

A common method of degassing adopted by the foundry involves the use of an inert gas to flush the dissolved gas by mechanical agitation.[7] The effectiveness of this treatment is shown by Fig. 3. This photograph shows two double horizontal full web-type test bars which have been split through the center. The specimen to the left represents metal from a ladle after tap. During solidification and as shrinkage occurred, the dissolved gas was present in such a large quantity that it resulted in spewing through the riser. The specimen to the right represents the same metal after the degassing treatment. Although gas was still present as evidenced by the vacuum test, it was reduced to a concentration low enough to remain in solution in this size casting. Mechanical properties of both specimens (Table 4) indicate the influence of gas upon this factor. The results of this test have been substantiated several times.

It must be appreciated that the larger the section size of the casting, the greater the influence of hydrogen. For example, a small casting may solidify

TABLE 4 – MECHANICAL PROPERTIES OF WEB-TYPE TEST BARS

Specimen	Yield Strength	Tensile Strength	% Elong. in 2 in.
A	46,180	86,550	12.5%
A_1	47,300	88,200	12.5%
B	47,550	97,000	25.0%
B_1	46,300	95,900	25.5%



Fig. 3 — Cross-sections of nickel aluminum bronze full web-type test bars. Both specimens were from the ladle after tap-off and solidified under atmospheric pressure. The specimen on the left is from before degassing treatment, and the one on the right from after degassing treatment.

so rapidly that a given percentage of gas is retained in solution without obvious harmful effects upon the casting. As the section size of the casting increases with the resultant increase in solidification time, the intensity of the problem increases. This has been demonstrated as follows:—

a) From the same ladle of untreated gassed metal (as indicated by the vacuum tester), a small cup specimen (1 lb) and the comparatively larger web-type test coupon is poured (18 lb). In the case of the cup specimen normal shrinkage occurs, while in the case of the test coupon, sufficient gas is rejected during solidification to build up pressure under the shrinking riser causing the riser metal to spew out over the top. Figure 4 is a photographic demonstration of this principle.

b) The above experiment is carried one step further, but this time with metal that has been given a degassing treatment. The cup specimen and the test coupon both shrink normally under atmospheric pressure. The gas is not sufficiently high this time to concentrate in any large volume in the test coupon. However, the next casting poured from this same metal is a propeller casting, let us say, in the neighborhood of 10,000 lb. When the mold is filled with metal, solidification begins at the mold surface and develops most rapidly in the thinner sections. As solidification progresses, the gas is rejected in larger volume and tries to escape through the liquid metal.

If the time from the liquid to solid state is appreciable, the riser metal and the heavy hub section

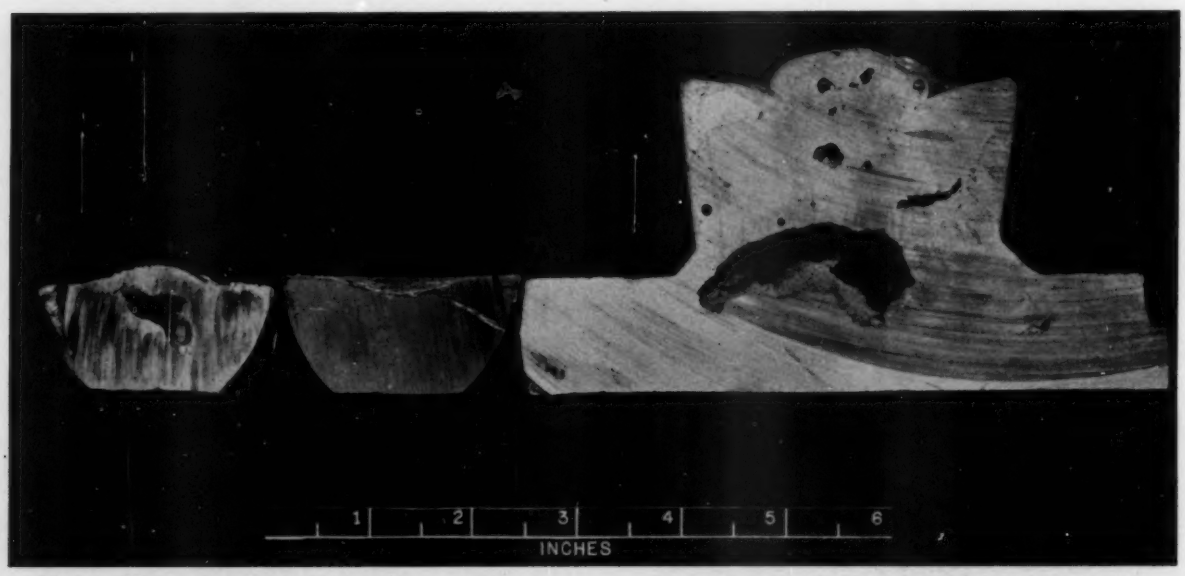


Fig. 4 — Three specimens from the same reverberatory furnace heat of nickel aluminum bronze. The specimen on the left was solidified under vacuum and shows a gassed metal condition. The center specimen was solidified under atmospheric pressure and the gas remained in solution. The right specimen was solidified under atmospheric pressure and the gas came out of solution.

become colder and less fluid and lack the ability to allow the gas to escape and the riser to perform its function. The result is that the gas is entrapped in the casting. The broken up areas of concentrated gas which are entrapped as the metal solidifies, result in many holes which are eventually diagnosed as shrinkage cavities. Actually, of course, the gas has prevented feeding which results in shrinkage defects.

Melting nickel aluminum bronze cannot be accomplished without some degree of gas absorption. However, with carefully controlled rapid melting, the use of hot dry ladles and an effective method of degassing, it is possible to deliver to the mold a metal containing a tolerable gas content.

MOLDING MATERIALS

There appears to be a wide variation in the type of molding materials used by propeller foundries throughout the world. For many years, the author's company's foundry has used a natural bond baked gravel sand for propeller molds. During the nickel aluminum bronze casting development period, experiments were made to determine the possible benefits of casting in a cement bonded sand which had been given wide publicity in the foundry industry.[8] Cement bonded sands were compared to the standard gravel sand.

It was concluded that the disadvantages of the cement sand outweighed its advantages for use in the shops and under the conditions of operation.

Advantages of the cement process were found to be as follows:

1) Molds are air dried and require no costly drying equipment.
2) No highly skilled personnel are required for molding.
3) Molding material has greater strength than sand and thereby the casting is less subject to distortion.

Claimed advantages which we were not able to substantiate were:

1) Propellers can be cast closer to size with concomittent savings in machining costs.
2) Better quality casting can be made in cement.
3) Foundry labor costs can be reduced.

Disadvantages of the cement sand process from the author's company's standpoint were found to be:

1) A well heated shop is required to prevent freezing of molds in winter.
2) Molding materials are more costly.
3) The molding material must be used within a few hours after mixing.
4) Mold shakeout time is greatly increased due to higher mold strength.
5) Cement cannot be reused and presents a disposal problem.
6) The cement molding process lacks the flexibility of the sand molding process for work speedup.

RAW MATERIALS

Scrap propellers (worn out or badly damaged) are commonly used in making up part of the furnace charge when casting new manganese bronze propellers. Such a practice aids the manufacturer in reducing cost and, if carefully controlled, does not detract from the quality of the casting produced.

It will be many years before a sufficient quantity of nickel aluminum castings will be available for scrap purposes. Therefore, from the outset rigid raw material requirements were set up for the ingot metal. For example, the use of scrap material for manufacturing nickel aluminum bronze ingots was prohibited. Only in this way could the presence of contaminating elements which could affect mechanical properties, casting quality and weldability be prevented. Ingot metal comprises all of the heat except for overmelt, gates and risers.

It is not advisable to use thin skulls and flashings such as those pulled from the ladle or runner box as they readily oxidize and contribute to the difficulty in melting. For the same reason it is poor practice to use any dirty or dross-covered material, or to melt in a furnace which has not been thoroughly cleaned of other metals. Occasionally when a charge is found lower in aluminum than desired, a master alloy containing the required additions may be added with the charge. When it is found necessary to adjust the composition of charged gates and risers which have lost aluminum during a previous meltdown, a master alloy containing the required additions may be added with the charge.

FEEDING OF PRODUCTION CASTINGS

Up to this point the discussion has revolved around melting, materials and metal handling. In continuing, the feeding operation will be considered, as it is perhaps the most difficult one of all in the casting of large nickel aluminum bronze propellers.

It became quite obvious during early investigations that feeding practices employed for small castings were not applicable to large nickel aluminum bronze propellers, nor was it possible to use manganese bronze practices for nickel aluminum bronze. To fully understand the situation, it is necessary to familiarize the reader with the general method of casting a manganese bronze propeller. This pertains only to those castings weighing in excess of 50,000 lb. The practice for smaller castings is generally less complex and cumbersome.

Metal from the ladle is poured into the runner basin containing a flow control valve. Larger castings require two ladles (Fig. 5). When the basin is full, the valve is opened and metal flows through the down and ingate to the base of the hub. Pouring is continued until the hub is filled and metal reaches between 6 and 12 in. into the riser. This distance is dependent upon the size of the casting and the dimensions of the riser. Metal approximately 100 F-150 F (38 C-65 C) hotter than the original pour is then added to the riser through a separate riser gate. The riser is insulated with a sleeve and cover of permeable casting plaster.[9]



Fig. 5 — Pouring a nickel aluminum bronze propeller casting from two ladles, necessary because of the casting size. When the basin is full the valve is opened and metal flows through the down and ingate to the base of the hub.

The insulating riser sleeve is a relatively new development, and has done an effective job in satisfying the conditions for adequate feeding. By preventing rapid heat loss, a liquid reservoir of molten metal is retained above the casting for a considerable period of time. Although other materials are available for this purpose, plaster has the advantages of being comparatively low in cost and simple to fabricate.

Directional Solidification

The above method of bottom pouring a casting would at first appear to be contrary to the fundamental principle of directional solidification. However, this is not the case. Most propeller designs are conducive to good directional solidification as the hub section is tapered and cast with the large end up. The massive surface area of the blades (the developed surface area of a 32,000 ton tanker propeller is approximately 27,000 sq in.) tends to dissipate heat rapidly and reduce time required for complete solidification of the casting. This in turn materially reduces the burden on the riser.

The large hot metal addition which is made to the insulated riser aids in setting up a temperature differential necessary for proper feeding. With all this, however, we must still contend with the fact that after the mold is filled the hottest metal is at the bottom, exclusive of course of hot metal which may have been added to the riser. This condition must reverse itself before solidification is complete otherwise the casting would contain shrinkage defects.

An experiment was made to determine the nature of the temperature reversal by placing thermocouples in the hub and riser of a manganese bronze Liberty ship propeller casting (rough cast weight of 27,000 lb) and plotting a time-temperature curve after the mold was filled. Figures 6 and 7 show the time-temperature curve and thermocouple positions for this propeller, and are a demonstration of the behavior, temperature-wise of manganese bronze when it is poured into a dry sand mold. Under parallel conditions all bottom poured castings of any material must have a similar cooling pattern.

The casting of a nickel aluminum bronze propeller is essentially the same as for the manganese bronze described above. As for the melting operation, there are also certain factors which tend to complicate the casting operation. First of all, unlike manganese bronze, the nickel aluminum alloy absorbs gas during melting. It has previously been shown by experimental work that it is possible to produce sound castings having acceptable mechanical properties from gassed metal, providing the gas content is not above a tolerable level and the rate of solidification is rapid enough to keep the gas in solution.

It can be appreciated that this latter condition is most difficult to control in the size castings under discussion. The emphasis must, therefore, be on the best possible melting practice to minimize gas absorption, and on an adequate degassing technique prior to pouring which will bring to the mold a metal of relatively low gas content. This is most important as excessively gassed metal will not feed, especially when the volume of metal involved is large and the solidification rate is spread over a long period of time.

Exothermic Reaction

Secondly, insulating plaster sleeves which had been successfully employed for manganese bronze, cannot be used for nickel aluminum bronze, as the higher pouring temperature of 2080 F-2130 F (1138 C-1166 C) causes an exothermic reaction with the plaster resulting in a violent boiling action which gases the riser metal. During the experimental program this fact was not obvious because the small casting cooled too rapidly for this reaction to take place.

Riser sleeves made from perlite, diatomaceous earth or other insulating materials, have been used with

some degree of success as these materials do not react at high temperatures. Lower insulating values are obtained from dry sand and cement sleeves, and make these materials less desirable.

Exothermic sleeves or pads are only partially effective in increasing feeding efficiency. When large castings are poured, the exothermic reaction takes place during the first few minutes after pour and the riser metal temperature increase at this time is marked from 50 F-150 F (10 C-65 C). This heat is rapidly dissipated, and by the time the casting requires liquid metal from the riser, the influence of the exothermic reaction is not available.

Some manufacturers of exothermic compounds supply their material in a number of grades of varying sensitivity. None, however, supplies grades with sufficient time delay for this application. Propellers such as the super-tanker class (28,000-100,000 ton tankers) require from 8-18 hr for complete solidification to take place. The only benefit such castings can obtain from these materials is from the insulation given by the burned out sleeve. Equally effective results can be obtained from much less costly materials. However, smaller castings which will completely solidify within a few hours can be materially aided by exothermic compounds. (Fig. 8).

Drossing and Nuclei Effect. Because of the drossing tendency of the subject alloy, it is most important that a gating system be designed to minimize the amount of turbulence developed during the pouring operation. Unsuccessful attempts were made to inert the mold with carbon dioxide to prevent the formation of dross during the pour. When the

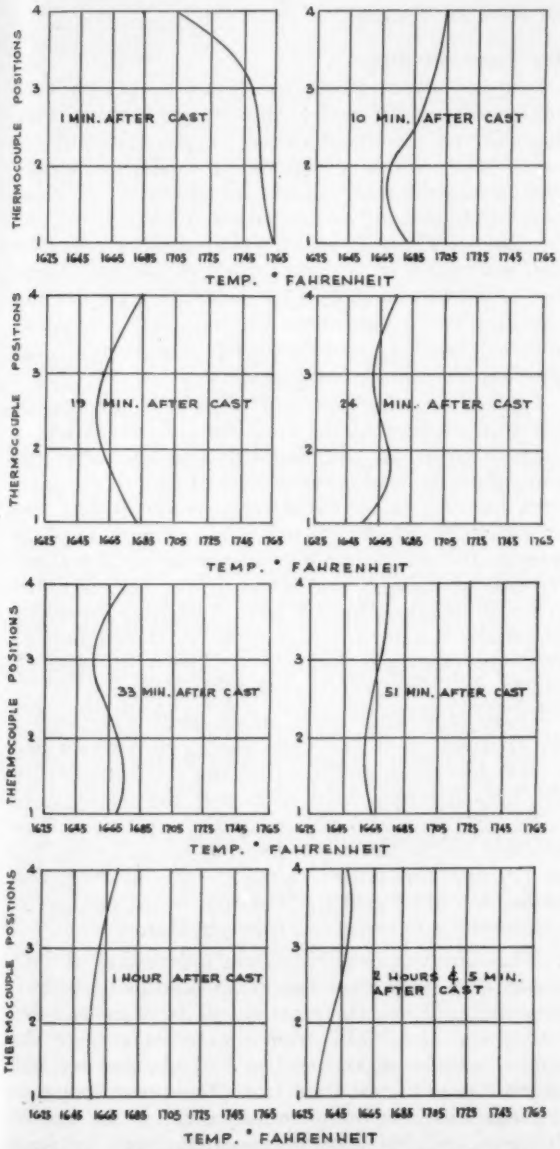


Fig. 6 — Temperature gradients in EC-2 (new design) propeller casting, cast with a bottom gate system and riser top filled with hot metal. Figure 7 shows the positions of the four thermocouples.

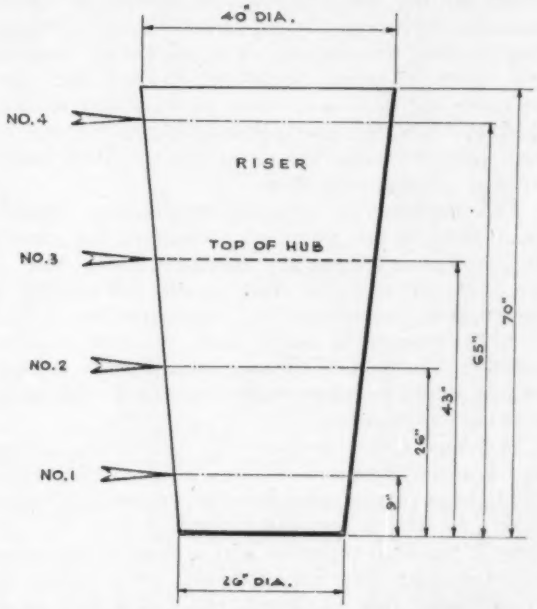
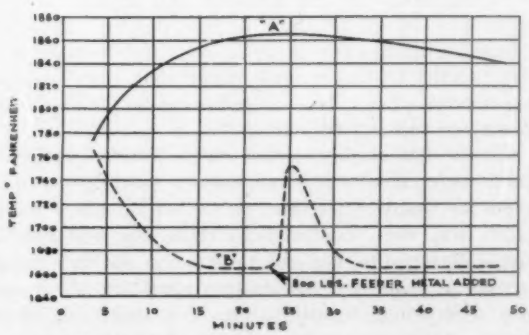


Fig. 7 — Details of positions of thermocouples in EC-2 propeller casting (Fig. 6). Thermocouples inserted into hub 5 in. from outside surface.



A — Exothermic sleeve and cover.
B — Dry gravel; plaster cover.

Fig. 8 — Time-temperature curves showing the rate of cooling in the risers of identical small manganese bronze propeller castings.

molten metal entered the base of the hub the heat caused the rapid elimination of the carbon dioxide, and the mold became for all practical purposes instantly filled with air. It was concluded that unless the mold could be made from an impermeable material and sealed inerting was a practical impossibility.

When the pour is complete and the metal has reached the top of the riser it is advisable to remove this dross to prevent its acting as a nucleus for further solidification. The problem is not a serious one for manganese bronze but from observing many nickel aluminum bronze castings, solidification progresses rapidly around this material. Unless it is removed the top surface is likely to freeze over and retard feeding. The job of removing this scum is difficult and unpleasant, and danger is always present that more will be formed if the work is not done quickly.

Pouring Temperature. Pouring temperatures which are not overly critical for manganese bronze must be closely regulated for nickel aluminum bronze. Higher metal temperatures aggravate the gas problem and result in greater shrinkage. With higher temperatures the mold loses some of its ability to chill the metal. On the other hand, temperatures must be high enough so that the metal fills the blade cavity without causing cold shuts.

The extremes of pouring temperature requirements must be met when it is considered that casting dimensions vary from approximately 1 in. thick at the blade tip to 15 in. thick in the hub section. A compromise temperature of approximately 2100 F (1150 C) appears to satisfy both of these requirements reasonably well, although shrinkage in the hub section would be more easily controlled with lower pouring temperatures.

Shrinkage Characteristics. The volumetric shrinkage of nickel aluminum bronze does not differ appreciably from that of manganese bronze, both being reported by various sources at anywhere from 6-10 per cent. It has been suggested that a peculiar formation of the primary dendrites which first freeze inhibit the gravitational flow of liquid metal. The previously mentioned possibility of gas coming out of solution would have the same effect.

Little or no solidification range exists for either of these materials as the liquidus and solidus are not separated by more than a few degrees. However, nickel aluminum bronze appears to have considerably less fluidity in this narrow range than does manganese bronze, obviously a contributing factor to the difficulty of feeding this alloy.

Chills. Although external chills judiciously placed in a mold have been successfully used by foundrymen for many years, the use of air or water cooled chills has been confined to relatively few applications. Because of the mechanics and obvious danger involved, other less satisfactory methods of promoting directional solidification have been employed. Large marine propeller castings, however, require a special application. A chill for castings of this size would have to be prohibitively large to be effective or it would require internal cooling so that it could maintain its effectiveness as a chill.

It is appreciated that the use of any kind of chill no matter how effective would have a questionable influence upon a propeller hub of the dimensions discussed here. For example, once the initial layer of metal solidifies and contracts from the chill could any further chilling action take place? It is conceivable that such a condition would render the chill completely ineffective due to the existence of an insulating air space between the casting and the chill.

However, it was decided to make a direct comparison of the influence of such a chill upon the rate of solidification of a large casting by pouring two identical propellers under closely controlled conditions and plotting a time-temperature curve on each. Because the practice for manganese bronze propellers was considerably more standardized and involved fewer variables, it was decided to run the comparison on this material.

Dry Sand Molding

Two five-bladed 32,000 ton tanker propellers were production molded in dry sand. During molding four thermocouples were inserted 5 in. into the hub. It would have been more desirable to insert the thermocouples into the mid-wall but this could not be done without subsequent major repairs to the casting. At the base of the hub, mold A contained a cast iron chill having a 1 in. diameter spiral copper coil. Crow foot connectors on each of the protruding ends of the pipe allowed for water inlet and outlet lines. Mold B, intended for the control casting, was identical except that no chill was employed.

Furnace capacity does not allow the pouring of two such castings on the same day, but the foundry was successful in maintaining approximately the same pouring temperatures and chemical compositions for each on different days. Water at full line pressure (75 - 85 lb/sq in.) was allowed to flow through the chill in mold A as soon as metal reached the base of the hub, and temperatures were recorded at 5 min intervals for 400 min. The same recording procedure was used for mold B. A comparison of each thermocouple location is shown in Figs. 9 and 10.

These time-temperature curves aptly demonstrate the effectiveness of the water cooled chill which had a most immediate influence on the lower portion of the propeller hub. Within 15 min the temperature began dropping at a comparatively rapid rate. This was not felt at the 21 in. location for 104 min, but at 170 min the chilled casting at this location had cooled to 1515 F (824 C). The nonchilled casting required 400 min to cool to this temperature.

With directional solidification progressing at such a rate, it becomes most important that the chill show some influence at the junction of the riser and the casting (located 48 in. from the bottom). Here the temperature break occurred at 252 min and reached solidification a short time later. The nonchilled casting was still in a liquid, mushy stage at the end of 400 min, and still depended upon the riser for feeding. Temperature measurements taken in the riser show the short-lived temperature influence of relatively small feeder additions.

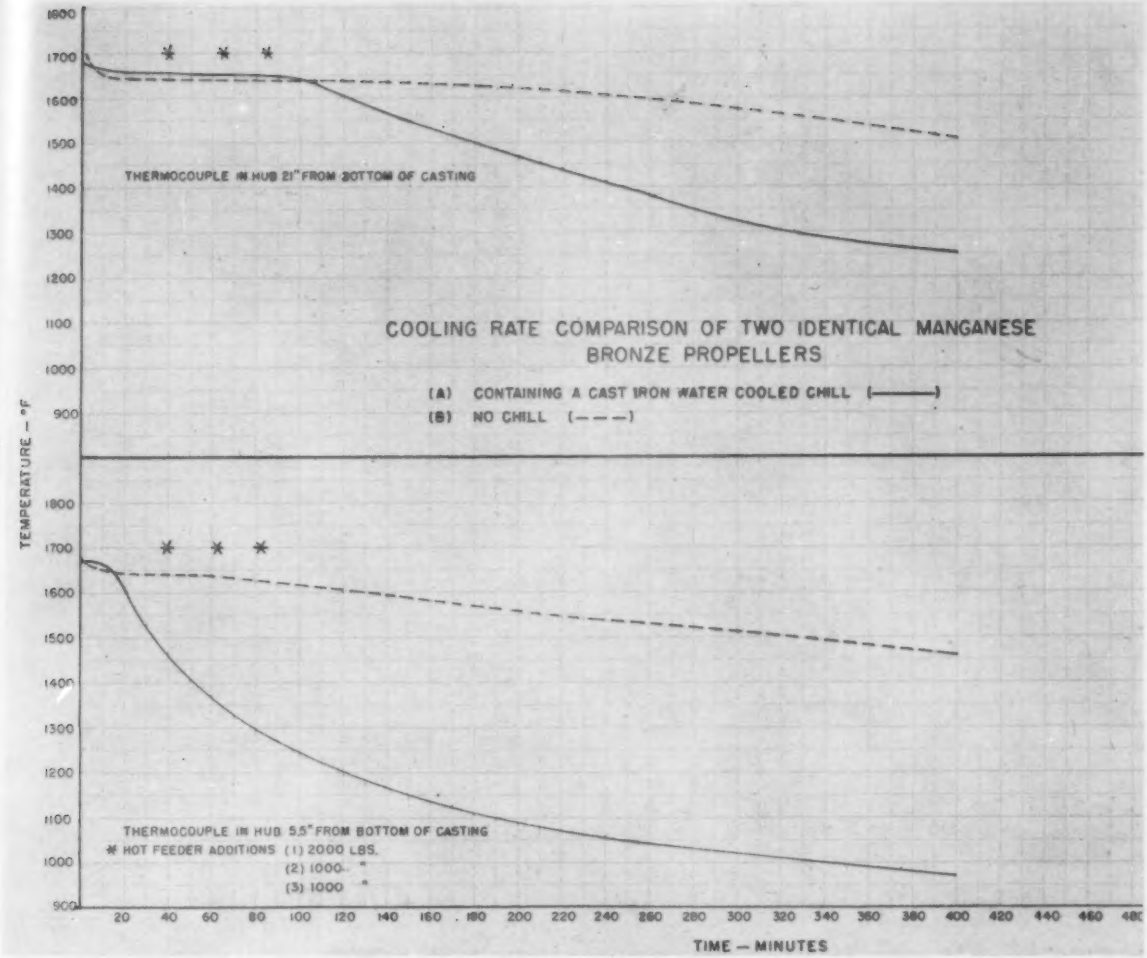


Fig. 9 — Cooling rate comparison of two identical manganese bronze propellers.

Although the above direct comparison was not repeated for nickel aluminum bronze propellers, the water cooled chill was successfully used for this material. Temperatures were taken of the chill during the casting period for both nickel aluminum bronze and manganese bronze and the time-temperature curves are shown in Fig. 11. The higher temperature reached by the chill for the nickel aluminum bronze casting is explained by the higher pouring temperature. The rather sudden temperature drop of this chill at 190 min is due to an increase of water pressure at this point.

The results of the above tests show that water cooled chills can be used effectively to decrease the solidification time for a large casting.

Although the chill is not necessary to produce sound manganese bronze castings it will enable the foundryman to increase his yield by reducing the riser volume, and improve his casting quality by promoting greater density in the heavy sections. Its advantages are more apparent, however, when applied to nickel aluminum bronze. By promoting more rapid solidification there is a greater likelihood of the ever present gas being retained in solution. If gas does come out of solution, as it is more likely to, the metal above will be more liquid (due to the shorter time it remains in the liquid state) and the gas can more readily pass through to the riser. The burden on the riser is materially reduced as it is not required to remain fluid for an unusually long period.

The effectiveness of a chill is in its ability to stay well below the metal temperature so that it is capable of rapid heat transfer. Gas, steam or air could also be used to cool the chill but there would be a marked decrease in efficiency. The latter has been employed at the foundry but results were less satisfactory than with water.

Undoubtedly it is possible to design a better chill than that described here. For example, high conductivity copper with an irregular surface would be more effective than gray iron with a flat surface such as that employed in the subject chill.

Some may question the value of a chill which reduces the solidification time to "x" numbers of hours, and admittedly this is still a long period of time for metal to remain in the liquid or mush state. It cannot be disputed, however, that halving the solidification time normally required is an appreciable speeding up of the process. It also follows that directional solidification is more certain, and the benefits derived

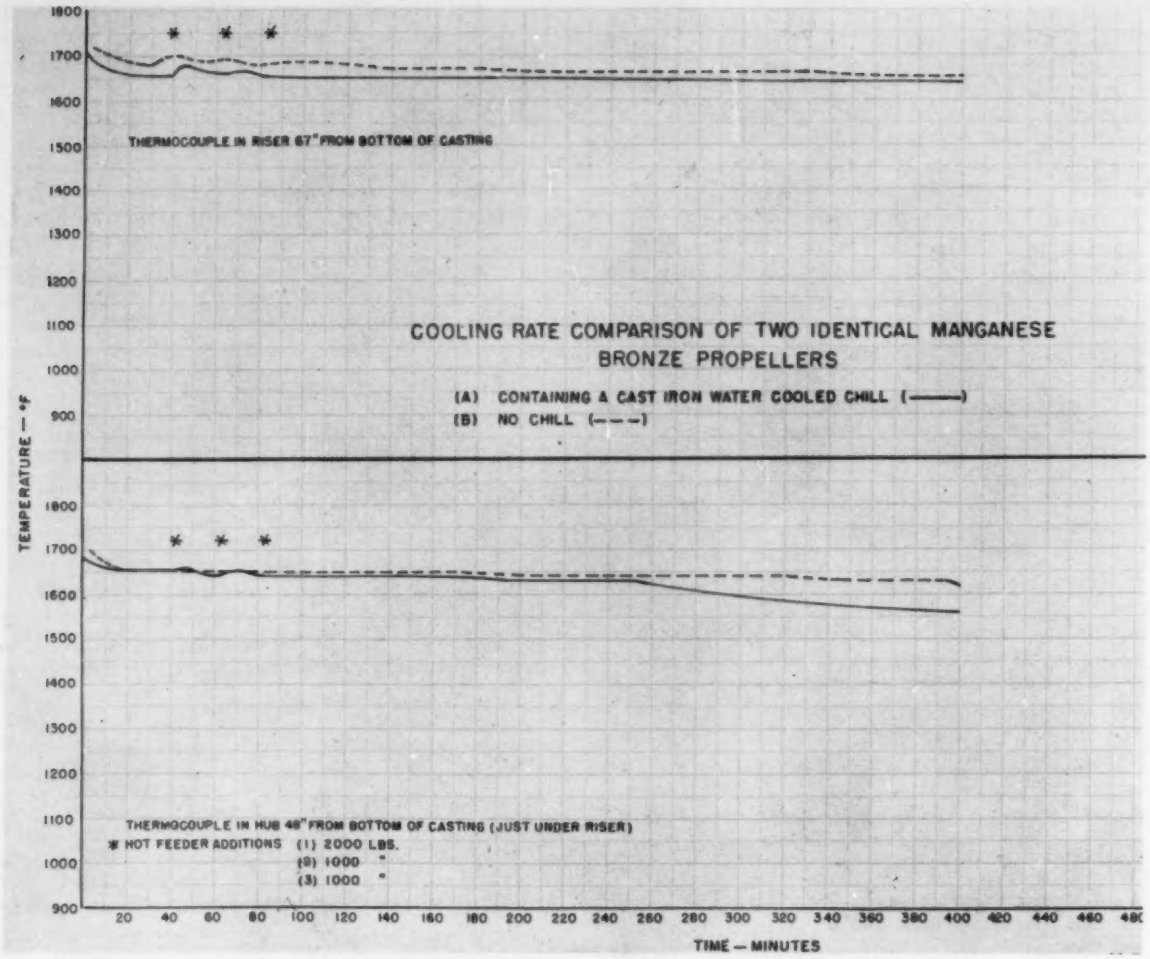


Fig. 10 — Cooling rate comparison of two identical manganese bronze propellers.

by the casting are quite positive. It is concluded that water cooled chills do work for the application noted.

There are admittedly certain disadvantages attached to the use of the water cooled chill which make it somewhat unattractive to the specialty jobbing shop. A variety of chill sizes is required to effectively cover the range of castings produced by a given foundry. They can be made interchangeable up to a point, but for full effectiveness the chill must be in complete contact with the base of the casting. Then there is the ever present danger of water under a mold full of liquid metal. The successful application of the water cooled chill does not allow for the human error.

A COMPARISON OF NICKEL ALUMINUM BRONZE AND MANGANESE BRONZE

Although this paper is primarily intended to present the general foundry problems involved in a new alloy development, it seems fitting to present a comparison of the old and new alloys with regard to service data and engineering properties obtained within the past few years.

Comparative data on chemical composition, mechanical properties and erosion-corrosion resistance are shown in Table 1. Further erosion-corrosion test data are shown in Figs. 12a and 12b.

Uniformity. Mechanical properties of a nickel aluminum bronze propeller casting are best represented by values of separately cast test bars. Tests made on this material show that only minor variations in properties exist from light to heavy sections of the casting. Manganese bronze lighter sections almost always exhibit properties superior to those of the heavy sections.

The composition of nickel aluminum bronze is responsible for its greater resistance to fouling by marine growth.

Comparative Service Data. The superiority of nickel aluminum bronze has been convincingly demonstrated by an in-service test. One propeller of each material was installed on a twin screw 6,000 ton ore carrier which subjects its wheels to unusually severe service by operating on the Orinoco River in Venezuela. The sand bars and sand in suspension in this river are responsible for damage to the ship's propellers, necessitating frequent reconditioning or replacement. For the past three years, at 6 month intervals these propellers have been examined in drydock. The nickel aluminum bronze (starboard) propeller

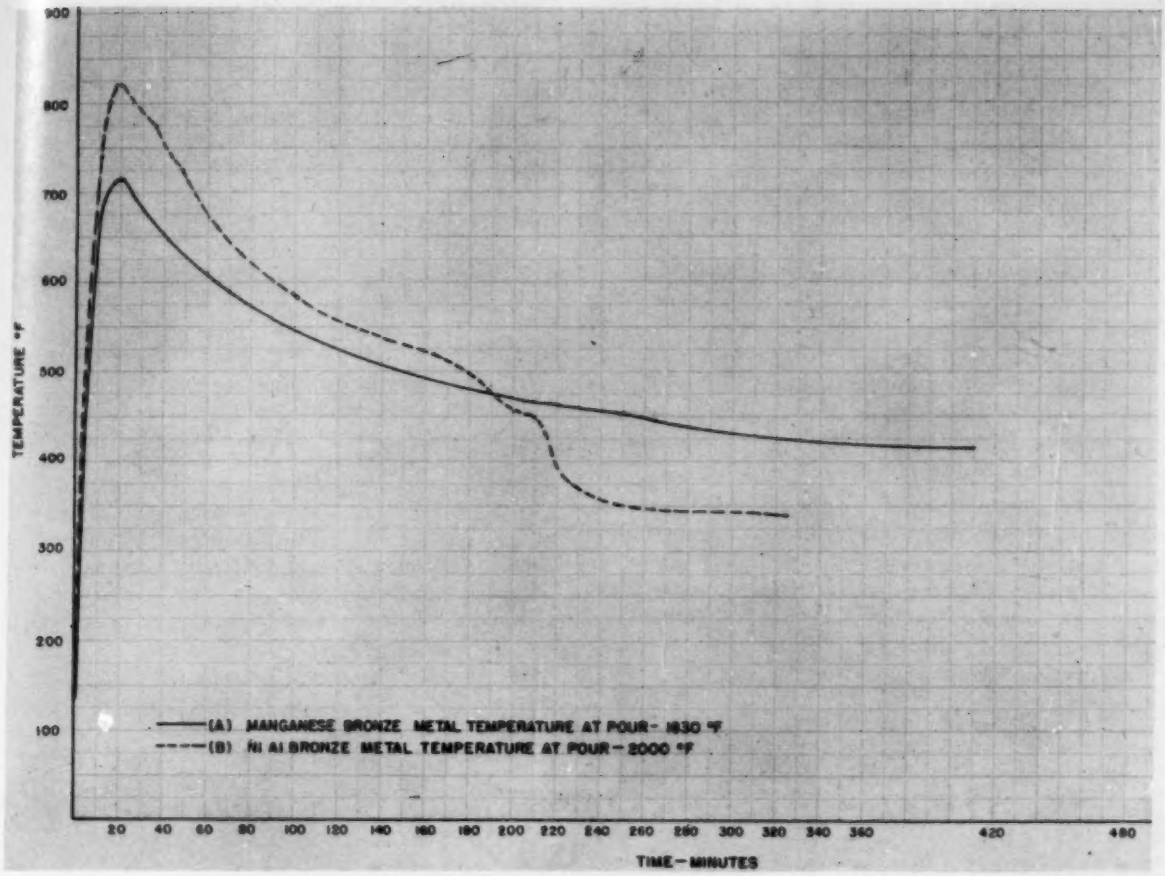


Fig. 11 — Time-temperature curve of cast iron water cooled chill in the base of propeller mold.

has suffered little mechanical damage or erosion, while the manganese bronze (port) propeller was rather severely damaged and required frequent repair. This program of comparing two different propeller materials under actual service conditions is continuing.

Two large high-speed, ocean-going luxury liners have nickel aluminum bronze propellers on one ship and manganese bronze on the other. Runs are similar in all respects. Notable fuel savings have been made on the ship with the nickel aluminum bronze propellers.

The U.S. Navy is using this material for propellers for ice-breaking service, and after a full season of operation excellent results were noted.

Other service follow-ups on nickel aluminum bronze propellers, although not on a direct comparison basis, have shown excellent wear characteristics. No service failures have been reported since these propellers were first put into service in 1954.

Propeller Efficiency. Less obvious perhaps than the data presented in Table 1, but more important from a service standpoint, is propeller efficiency. The design engineers at the author's company are most impressed by the ability of nickel aluminum bronze to retain its original smooth machined surface over a long period of time, thereby retaining its high efficiency factor. Numerically the improvement in efficiency would lie in the order of 1.5-3.0 per cent, with resultant fuel savings. The exact efficiency increase would depend upon the propeller size, design factors and length of service.

Design Benefit. As nickel aluminum bronze is in itself approximately 10 per cent lighter in weight than manganese bronze, and can be designed to thinner sections because of its higher strength, other advantages become apparent. For example, loading stresses on the tailshaft and bearings are reduced, thus permitting smaller shafts.

Resistance to Notch Sensitivity. The ability of nickel aluminum bronze to resist failure under impact when notched contributes greatly to its value as a propeller material.

A series of standard nickel aluminum bronze propeller material samples was submitted to an independent laboratory for notch sensitivity and impact tests. The general results of such tests can be summarized as follows: A brittle weld bead was used for purposes of developing a sharp crack in the test samples. Despite the presence of this crack, the material refused to break in tests at temperatures as low as −170 F (−112 C). This would indicate that the material is not susceptible to brittle fracture of the cleavage failure type.

Maintenance. Maintenance of nickel aluminum

bronze propellers under identical service conditions is greatly reduced compared to that of manganese bronze. Its superior resistance to bending, breaking and wearing, including cavitation is directly associated with the chemical and mechanical properties of the material.

Reparability. Nickel aluminum bronze is readily reparable with the inert gas process. Little or no preheat is required, and unlike the copper zinc brasses, it is not subject to stress corrosion cracking and therefore does not necessarily require a stress relief treatment. Normally repairs can be made with either the solid or stranded aluminum bronze wires. Excellent results have recently been reported with a high manganese nickel aluminum bronze wire. The straightening of propeller blades must be done at temperatures in excess of 1400 F (760 C) to avoid cracking in the brittle range.

Propeller Cost. Reduced weight of nickel aluminum bronze in conjunction with increased strength of the material allows designing the propeller approximately 18 per cent less in weight than a comparable manganese bronze propeller. Although the former is more costly by the price per pound, the long term cost reduction is appreciable.

OTHER APPLICATIONS OF NICKEL ALUMINUM BRONZE

Although this paper is primarily concerned with nickel aluminum bronze as a superior propeller alloy, its use is in no way confined to any specific engineering application. Utilization of nickel aluminum bronze has extended into almost all fields[10] where its superior strength, wear resistance and comparatively low density can be effectively realized. Industries such as aeronautical, tool and die, marine, electrical and chemical are currently employing this new material.

SUMMARY

Increased service requirements of the new higher horsepower ships of the post-war period established the need for a new propeller alloy. This paper presents a general review of some of the foundry problems involved in developing a casting practice for nickel aluminum bronze. The material was selected as the alloy having the best combination of wear resistance, mechanical properties and foundry handling qualities. However, many problems arose before an acceptable casting practice was developed. Some of the findings regarding these problems are outlined here.

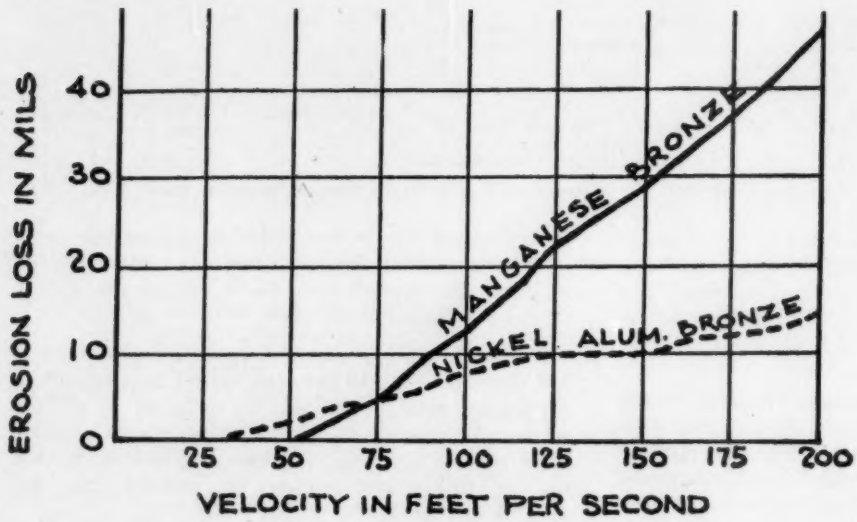


Fig. 12a — The measured erosion losses on 8 in. discs spun for 500 hr at a peripheral velocity of 200 ft per sec.

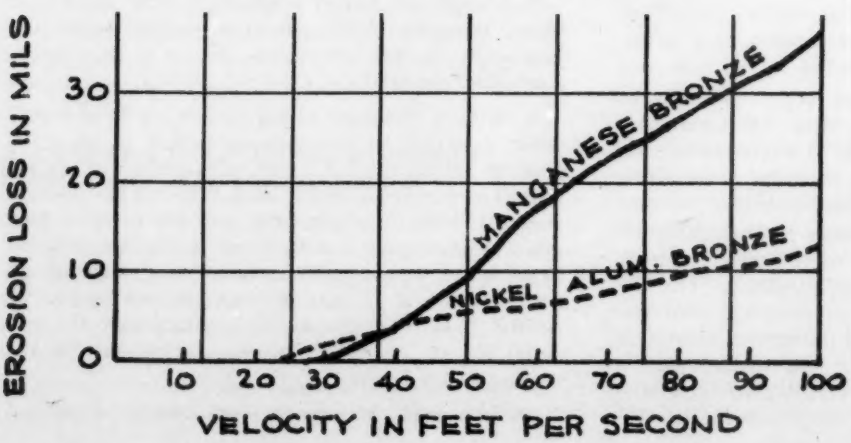


Fig. 12b — The measured erosion losses on 8 in. bronze discs spun for 500 hr at a peripheral velocity of 100 ft per sec.

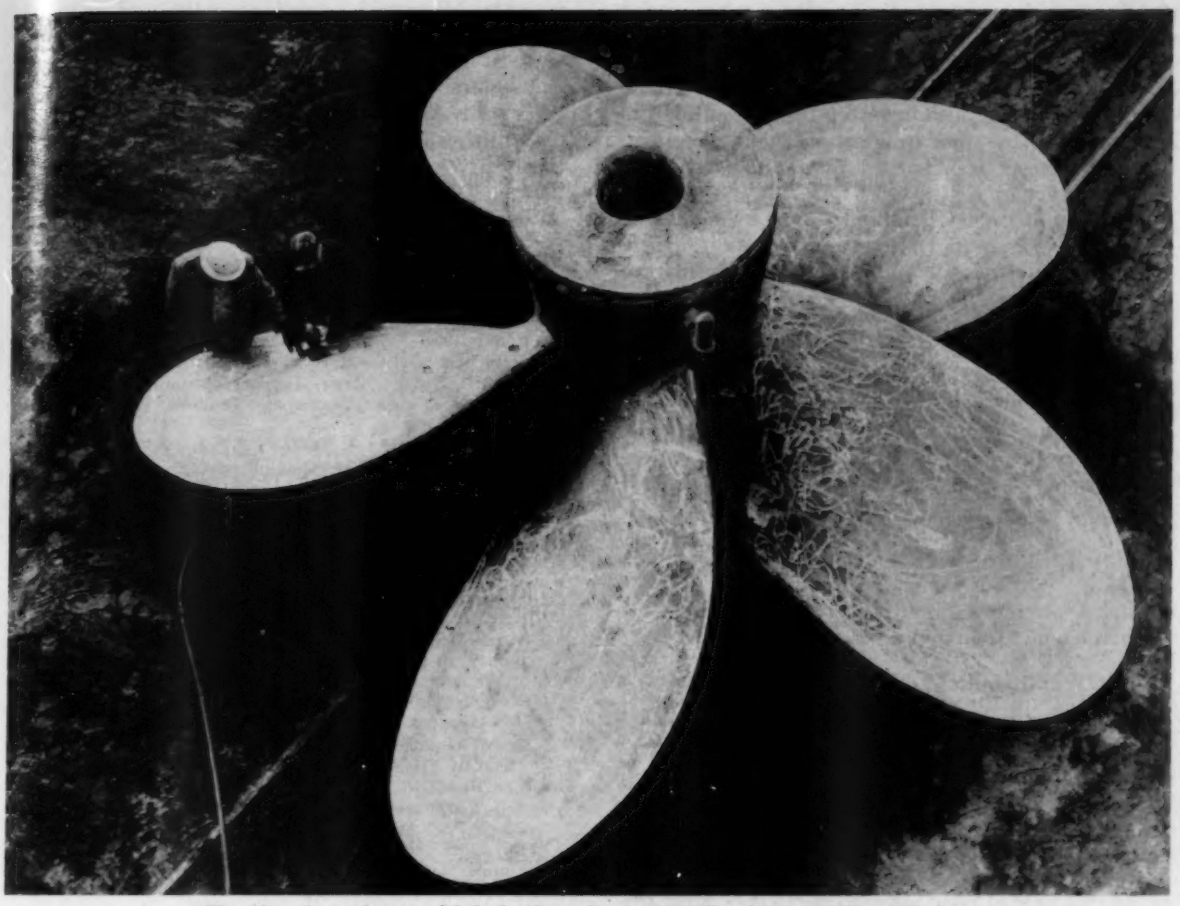


Fig. 13 — A rough cast nickel aluminum bronze propeller for a 32,650 ton tanker.

1. Melting Practice

a) It was found necessary to move the burner flame closer to the metal bath to prevent excessive oxidation of aluminum and to obtain faster meltdowns.

b) Close control of the combustion atmosphere was required to prevent excessive gas (hydrogen) absorption and oxidation.

c) It was found essential that raw materials be controlled by rigid specifications.

2. Gas Absorption

a) Gas was absorbed by metal melted in all types of melting equipment tested but was most pronounced in reverberatory furnaces.

b) The reduced pressure tester was utilized to establish the presence of gas in the molten metal.

c) Excessive gassing of the metal markedly reduces the mechanical properties and results in serious casting defects by hindering the feeding operation.

d) Gas pickup could not be avoided entirely but could be reduced to a comparatively harmless level by flushing with an inert gas.

3. Molding Methods and Materials

The conventional molding methods and materials (baked gravel sand) used at the author's company presented no particular problem in the casting of this alloy. Cement bonded sands were also investigated.

4. Refractories

There is an apparent need for high grade refractories at reasonable cost.

5. Fluxes

Fluxes were successfully employed in crucible furnaces but hindered the melting operation when used in the reverberatory furnaces.

6. Feeding

The most difficult problem in the production of nickel aluminum bronze was feeding the propeller casting. Plaster insulating riser sleeves could not be used due to the high temperatures involved. Exothermic riser materials, although satisfactory for small castings have limited use on the larger castings because of the long period required for complete solidification.

Water cooled chills were successfully used to promote directional solidification in large propeller castings and materially reduced the time required for complete solidification to take place.

7. Pouring Temperature

For nickel aluminum bronze of low gas content, pouring temperature has no appreciable influence upon mechanical properties.

8. Advantages of Nickel Aluminum Bronze

Although manganese bronze is still the most widely used propeller alloy and continues to give acceptable service for slower propulsion machinery, nickel

aluminum bronze is rapidly gaining in popularity throughout the shipbuilding industry. Comparative service tests under identical conditions of operation have placed nickel aluminum bronze in a favorable light. Some of the advantages of this new alloy are made obvious by comparing its mechanical properties with those of manganese bronze. Other less obvious advantages are:

a) Propeller efficiency is retained over a long period of time due to its resistance to surface roughening. This efficiency increase results in marked fuel savings.
b) Because of its lower density and higher strength, a nickel aluminum bronze propeller can be made approximately 18 per cent lighter in weight than a comparable manganese bronze propeller. This reduces the loading stresses on the tailshaft and bearings, thereby permitting smaller shafts.
c) The material has comparatively low notch sensitivity.
d) Although the nickel aluminum bronze alloy is more costly by the price per pound, its long term cost is lower than manganese bronze.
e) Nickel aluminum bronze propellers require less maintenance than manganese bronze propellers.

ACKNOWLEDGMENTS

The author has recorded some aspects of a development program which entailed the cooperative efforts and hard work of many people. Particular reference is made to P. Ffield, Bethlehem Steel Co., F. L. LaQue, J. S. Vanick and G. Lee of The International Nickel Co. for their early aid and encouragement of this program and for their continuing interest in this field. L. M. Mosher, Bethlehem Steel Co., Shipbuilding Div., contributed greatly to this program by his erosion-corrosion studies and in-service follow-up of propeller alloys. E. R. Miller, Foundry Supt. at Bethlehem Steel Co. Staten Island Foundry and his staff exhibited great persistence and skill in successfully carrying out this program. J. Fitzgibbon of Whitehead Metal Products Co. successfully stimulated early interest in the nickel bronzes for propellers. A. J. Krug, Met. Chemist and Asst. to the author aided in putting together this paper.

REFERENCES

1. P. Ffield, "A Shipbuilder Looks At Corrosion," Society of Naval Architects and Marine Engineers, *Transactions, vol. 62*, p. 697, 1954.
2. Bethlehem Steel Co., Inc., Shipbuilding Div., Central Technical Div., Quincy, Mass., "Erosion Resistance of A Series of Proposed Propeller Brasses and Bronzes," Research Report No. 84.
3. American Society for Testing Materials, "A.S.T.M. Standards — Part 2, Non-Ferrous Metals," pp. 346, 348, Dec. 1955.
4. Copper Development Assoc., *Aluminum Bronze*, C. D. A. Publication No. 31, p. 88, Copper Development Assoc., Kendals Hall, Radlett, Herts, England, Fifth Edition, 1951.
5. L. W. Eastwood, "Gases in Non-Ferrous Metals and Alloys," *Gases in Metals* by D. P. Smith, D. J. Carney, L. W. Eastwood and C. E. Sims, 1953, pp. 23-68, The Haddon Craftsmen, Inc., Scranton, Pa.
6. W. D. Walther, C. M. Adams, Jr., and H. F. Taylor, "Vacuum Tests Show Up Gassy Metal," AMERICAN FOUNDRYMAN, March 1955, pp. 62-64, from "Techniques for Improving Strength and Ductility of Aluminum Alloy Castings," AFS TRANSACTIONS, pp. 219-230, *vol. 62*, 1954.
7. P. M. Hulme, "Removal of Dissolved Gases from Molten Metal by Flushing," Reprinted from *Steel* by Air Reduction Co., New York.
8. F. J. Tector and J. Martland, "The Casting of Marine Bronze Propellers," *Foundry Trade Journal*, July 22 and 29, 1948, pp. 73-80 and 99-104.
9. United States Gypsum, Technical Bulletin No. 1GL-35.
10. J. S. Vanick, "Nickel Aluminum Bronze," *Foundry, vol. 79*, No. 8, August 1951, pp. 98-102 and *vol. 79*, No. 9, September 1951, pp. 108-111, 192 and 195.

PRACTICAL PLANNING FOR NON-DESTRUCTIVE TESTING BY USE OF RADIATION

By F. S. Sutherland

ABSTRACT

The general problems concerning non-destructive testing by radiation are considered. The problems discussed are:

1) Choice of source or unit.

2) Exposure room design.

3) Protective materials.

4) Fixed and portable sources.

5) Employee considerations.

The three testing units described are the 250 kv x-ray, the 1000 or 2000 kv x-ray and the betatron.

When a company decides to accept orders for a product which will require x-raying, what procedure should be followed? If the company is located in a large urban area where industrial x-ray facilities are available, it may choose to use these facilities, until the extensiveness of this program has been decided as well as its future possibilities.

If the company is not located in a community where the necessary facilities are available, or believes it would be more advantageous to install its own x-ray department, then there are several problems to consider. It is the purpose of this paper to point out some of these problems.

In general, the problems fall in three main categories. 1) The selection of personnel to perform the work, 2) the source of radiation and 3) the protection of those who will work with radiation equipment for radiation is no respecter of persons.

EMPLOYEE SELECTION

The number of employees required will depend on several factors, such as:

1) Type of equipment obtained.
2) Volume of work.
3) Rapidity of exposure time.
4) Conditions under which work must be performed.

For example, a crew of three will be sufficient where exposure time is approximately 3 min and four to 10 radiographs are taken per hour. Two employees for positioning the films and source, and the third employee in the darkroom.

The employee selected to be responsible for results should be capable of exercising ingenuity as many technical problems will arise when a variety of castings are submitted. He should have some knowledge of which part of a casting is more susceptible to defects, and how to bring out the defect to best advantage on a radiograph. He should have some knowledge of darkroom technique. Good darkroom procedures make it possible to avoid a lot of retakes.

The employee responsible for the operation should be able to train those working with him. He should see that no employee in the plant is exposed to excessive radiation. If electrical equipment is used, this employee, or some maintenance employee, should become sufficiently familiar with the installation so as to maintain it in proper working order and make minor repairs.

CHOICE OF SOURCE OR UNIT

The most common equipment used in radiographing metal in foundries are:

1) Portable capsule 250 kv-x-ray.
2) 1000 or 2000 kv-x-ray.
3) Betatron.

These are commonly referred to as a pill – a 250 (1), million volt or 2 million volt (2) and 24 million betatron (3) (Figs. 1,2,3).

One or more of the above units will generally accomplish the work encountered in foundries. In order to acquaint those who have not had experience



Fig. 1— A 250 kv x-ray unit, showing braces used for support.

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in x-ray with the problems involved, a brief general discussion is given.

The choice of unit will depend on the kind of metal, the metal thickness and the amount of work to be performed in a given period of time. For example, a small unit may be sufficient for the metal thickness involved, but the exposure time will be too long for the number of pieces per turn. A larger unit could reduce the exposure time to one-fifth that of the smaller unit, and the work all performed on one turn. Also, continued use of a unit for long exposures frequently increases the maintenance costs. A small unit may also be idle time for the x-ray crew unless an extra setup room is provided, as exposure time mounts rapidly on a small unit. Figure 4 gives some idea of the time required for various metal sections with different sources of radiation.

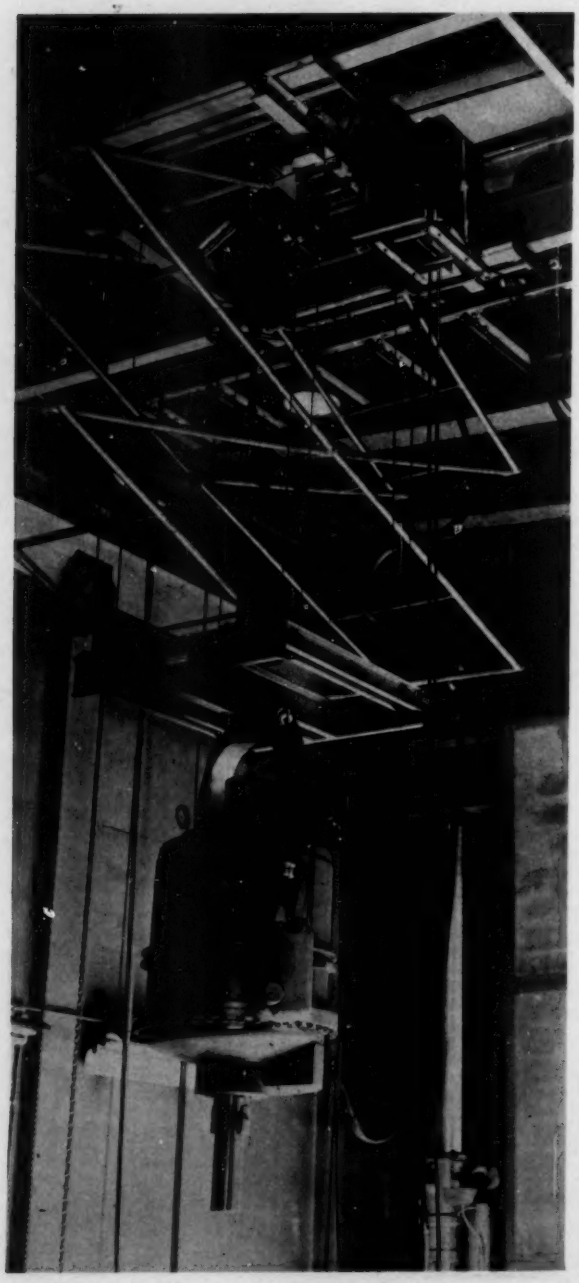


Fig. 2 — A million-volt x-ray unit, showing pantagraph supports.

LOCATION

The location of a given-size unit will determine the type and amount of protection required. If the unit is isolated, then only the operating personnel and film storage need be protected. If the unit is located in a building where other employees are working, then protection must be adequate for all possible exposures. Remember that radiation can radiate in all directions, and should employees be working above or below this room they must be protected. The craneman who may have to pass overhead cannot be overlooked. The more powerful the unit, the more protection that is necessary. The suppliers of the unit can give a great deal of advice and information concerning protection, and also check the installation for weak spots.

DESIGN

Considerable thought should be given to the design of the layout, for mistakes are expensive to rectify. A compact exposure room will be less expensive to build, but insufficient room may necessitate rebuilding at a later date. An L-shaped room can save a lot of expense on door protection. It is wise to visit some other installations and get ideas before proceeding.

The operating panel should not be built into a wall without access to both sides by maintenance

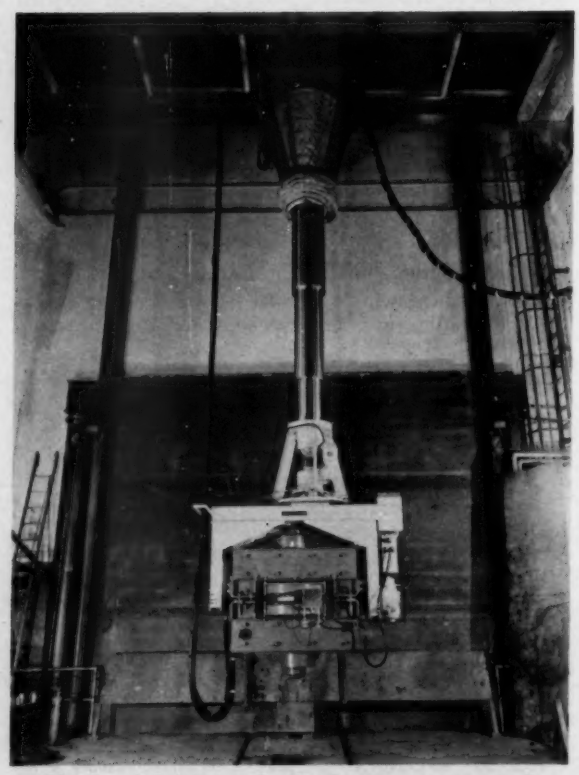


Fig. 3 — A betatron unit, showing crane arrangement.

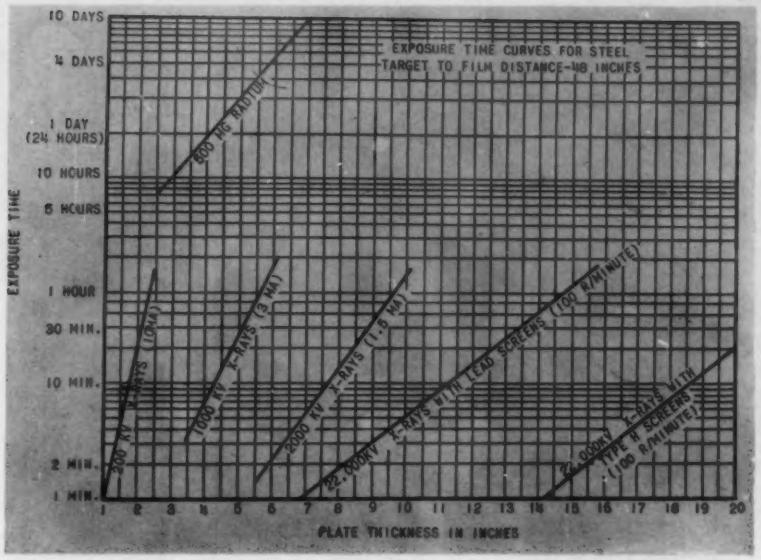


Fig. 4 — Graph showing time required for various metal sections with different sources of radiation.

men (Fig. 5). Wherever the panel is located, this dual accessibility is necessary. This is also true of the dark room developing tanks (Fig. 6). Figure 7 shows a simple design that is functional, compact and planned so that an excessive amount of walking is unnecessary. The electrical and plumbing facilities are accessible to the maintenance crew (Fig. 8).

Work may be delivered through an ordinary door or interlock door at the work entrance by truck or transfer car. Castings may be temporarily stored in this area if the workroom is congested. An overhead crane may be used to transfer the castings to the radiographic area. This crane may travel on the same track that carries the support of the x-ray equipment. The wall thickness around the x-ray room is determined by the size of the unit and construction material.

If the number of items to be x-rayed are numerous and the exposure time short, a preparation room may be advantageous, as a setup may be completed on a car while other exposures are being made. Cars can then be exchanged and the process continued. The volume of work can almost be doubled by this procedure. Also, the radiation room can be made smaller and the expense for protective material reduced.

PROTECTIVE MATERIALS

The most common protective materials for radiation room are lead, steel and concrete. One in. of lead is equivalent to approximately 4 in. of steel or 12 in. of concrete. Combinations of these materials may be used to advantage.

As mentioned previously, if the building is isolated, preferably in a fenced enclosure, then only the operating personnel and film storage need protection. The roof and the three side walls may need little, if any, protection. If the unit is located in a one-story building where there are other employees, then protection for the 4 walls must be supplied. If there are workers above or below the radiation room, then additional protection is necessary. Again, the

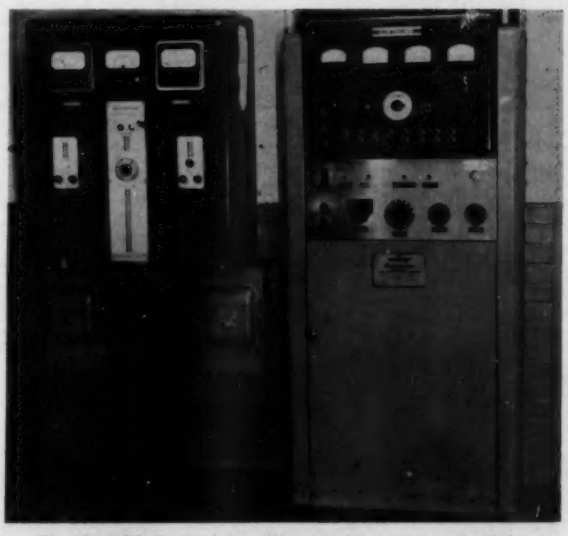


Fig. 5 — Method of installing an instrument panel for x-ray and betatron units, built into the wall in such a way that the panel can be repaired from both sides by maintenance men.

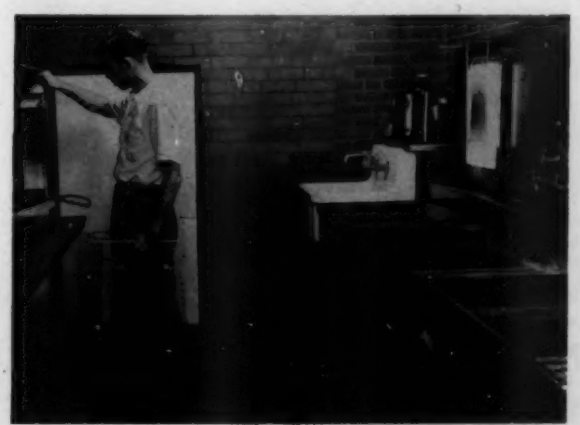


Fig. 6 — One type of darkroom arrangement. The drying cabinet opens out into the viewing room.

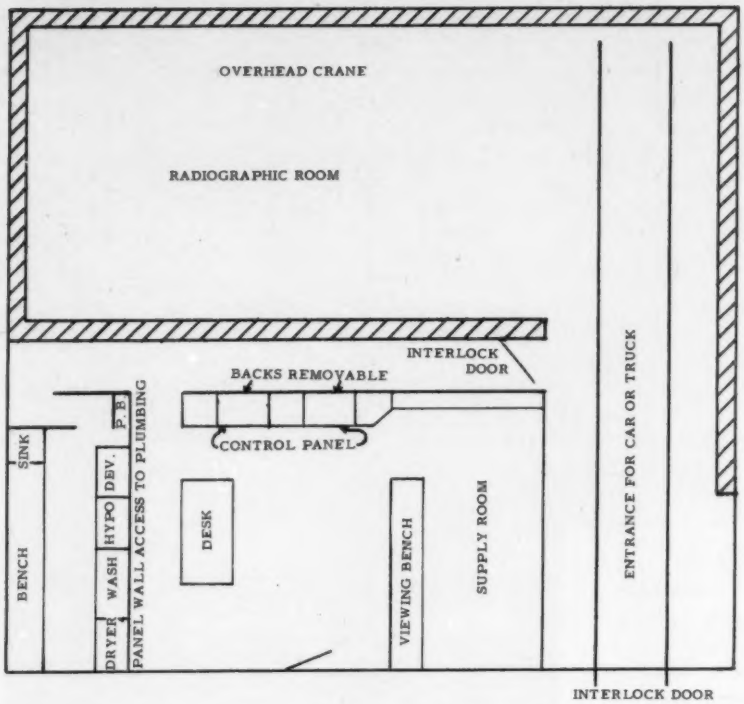


Fig. 7 — A simple, functional design for an x-ray building that is compact, and planned so that an excessive amount of walking is unnecessary.

supplier of the unit can be of assistance as to amount of protection required for a particular unit.

It may be mentioned here that some state laws endeavor to control this protection, and a company must comply to the satisfaction of an inspector.

Frequently, the building design includes a protective passageway between the radiation room and the control room.

A right-angle exit to the radiation room will serve as protection to a point close to the exit. Both walls, however, should have sufficient protection. Tests should be made in the passageway for scatter, with the unit in operation in various parts of the radiation room. X-ray units will give a great deal more scatter than a betatron due to the character of the beam.

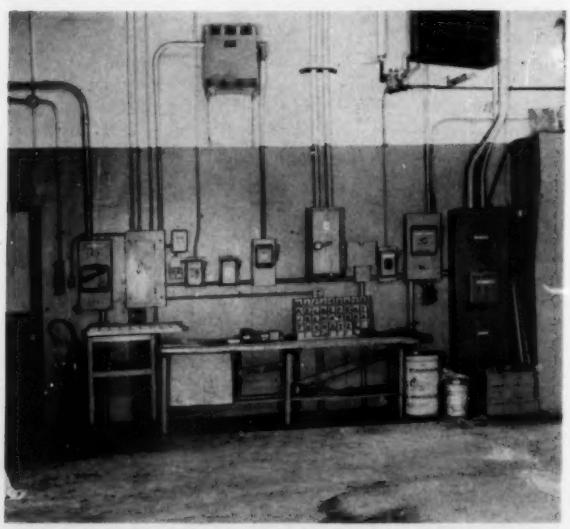


Fig. 8 — Example of installation of electric switches located near a door so they are handy in case of electrical fire. This is an important consideration in designing an x-ray building.

FIXED SOURCES

Fixed energy sources usually refer to x-ray (Fig. 9) and betatron installations (Fig. 10) to which the material to be tested must be brought. These units are generally mounted on a crane, and can be positioned up and down, forward and reverse and in a revolving motion. The unit, however, is confined to one building. An auxiliary crane is generally provided for handling the castings, as these units are for heavy metal sections.

Included in this class are the million and two million volts x-ray and the betatron. For protection here, 18 in.-72 in. concrete walls are used on all parts where employees may be adjacent. By the use of steel or lead sheets as inserts or linings, the wall thickness may be reduced. Consult your supplier of equipment to be certain your protection is adequate. The installation should be thoroughly checked with the units in various positions. On the smaller x-ray machines, the protection can be modified considerably and the expense reduced. Since this equipment is much lighter, it may be hung from an arm or other light crane. If a great number of small castings are to be handled, some kind of positioner will expedite the amount of work which can be performed, and space saved, all of which cut protection costs.

PORTABLE SOURCES

Generally, portable sources include x-ray units up to 250 kv, and various types of radium and cobalt capsules. For these units, there are numerous variations in protection. If the capsule is removed from a

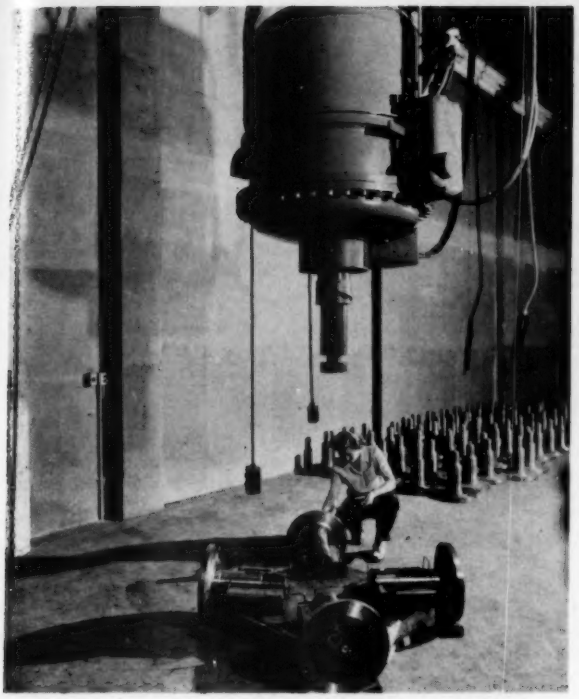


Fig. 9 — A million-volt x-ray unit, showing a four shot arrangement. Four castings are arranged for radiographing, using only one exposure. The controls for positioning the unit are seen in the foreground.

safety container, precautions must be taken by all personnel. Usually, the occasion is out in the plant some place, and distance is about the only safety factor that can be employed. The capsule can be handled by a long pole such as a fishing rod. This permits pulling the capsule up to end of the rod and then removing the rod during exposure. Roping off the area and placing warning signs on all sides should provide protection for other employees. This procedure permits the accomplishment of several exposures at one time by setting castings in a circle around the capsule (Fig. 11).

Where individual shots are required, there are various types of equipment which are available to provide protection. Capsules are enclosed in a container, with an opening for exposures. Portable x-ray units of various sizes also are available, both mounted and unmounted. In these, protection is provided for the employee performing the work. However, the problem of protection for other employees is still necessary. Here, again, distance is often the only protection available.

At this point, it should be mentioned that intensity of the radiation is inversely proportional to the square of the distance. It is therefore apparent that reasonable protection can be attained by making a given area out of bounds to all personnel.

Where a direct beam is employed, a steel or lead screen can also be employed for additional protection when the operation is confined to a limited space. It should be remembered that radiation, however weak, is accumulative, and continued exposure should be prevented. Individual checks are always



Fig. 10 — A large steam turbin casting, metal 16 in. thick, set up for radiographing by betatron.

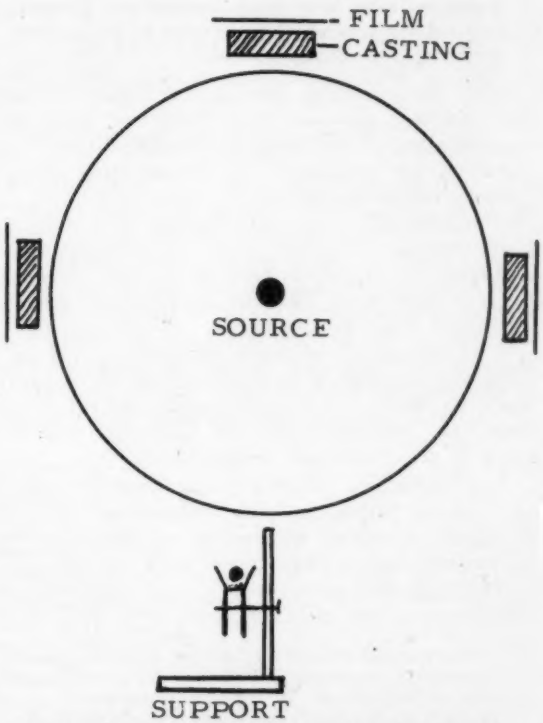


Fig. 11 — Method of setting castings in a circle around the capsule so several exposures can be made at one time.

advisable. This may be accomplished by a pencil-shaped dosimeter made for this purpose and worn by the employees. Also, the badge system can be used for weekly checks. This badge contains a sensitive film which is developed by the supplier and read for intensity of accumulation, and serves as a record if a question of liability arises.

Also, periodic blood or urine analysis should be employed as a record of employee's physical condition. These records are valuable in determining any general trend in blood condition. The application of these tests will be determined by the plant doctor.

MODERN QUENCHING OILS AND TECHNIQUES FOR FOUNDRY HEAT TREATING

By N. F. Squire

ABSTRACT

The increasing importance of, and variables affecting, the quenching phase of heat-treating steels is discussed. The conventional quench, marquench and modified marquench are dealt with, giving pertinent data for each. Case histories showing results of using an additive-type quenching oil medium are presented.

With the increased demand for heat-treated castings having physical properties not obtainable by annealing, normalizing or homogenizing, many foundries are installing new heat-treating facilities utilizing liquid quenching.

Foundries not previously engaged in this type of heat treating will find that the quality control standards for heat treatment of wrought steel parts have become more rigid in recent years, and will be equally rigid for castings.

It is no longer sufficient for parts simply to be hardened. Because of high processing costs, castings must be clean, have good machinability, a minimum of distortion and, in addition, maximum physical properties and fatigue life.

The quenching phase of heat treating plays an increasingly important part in meeting these rigid standards, especially in leveling out the many variables with which the heat treater must contend.

The American Society of Metals states this very adequately in the *Metals Handbook*, p. 615:

> "The purpose of quenching steel generally is to produce a completely martensitic structure on the surface of the piece to be hardened (a minimum requirement) or in the center of the piece (a maximum requirement). Hardening implies the formation of martensite, which, before tempering, provides the highest hardness obtainable in a steel of low alloy content. This hardness is directly related to carbon content.
>
> "If the hardening is incomplete, that is, if some pearlite or ferrite is present, usually the mechanical properties will be inferior to those of a completely martensitic steel, even though both are tempered at the same hardness. Accordingly, when highest mechanical properties are required, quenching must be controlled with precision.
>
> "Whether or not satisfactory hardening is achieved, depends upon the composition and structure of the steel, the shape of the piece, and relation between size and section, as well as on the composition of the quenching medium. Because of the complex relations among all of these variables, few of which can be expressed quantitatively, hardening remains largely an art. However, failures during hardening can be minimized by recognition of a few well-established relations among hardening variables."

N. F. SQUIRE is Vice Pres., Aldridge Industrial Oils, Inc., Cleveland.

VARIABLES

Time-Temperature Factor

The variables mentioned apply to castings as well as to conventional parts made from bar stock or forgings. One of the most critical variables in determining the quenching media required is the time-temperature factor of both composition and structure.

The difference between shallow hardening and deep hardening of steel is the cooling rate required through a critical temperature zone of approximately 1200 F - 800 F (649 C - 427 C) so a completely martensitic structure can be obtained and undesirable transformations avoided.

An available publication shows the time-temperature-transformation curves of many casting compositions as well as wrought steels. In this publication if the curves for any particular casting composition are not shown, the curves for similar wrought steel compositions can be used as a guide as their quenching reaction will be the same.

Variations in the T-T-T factor are shown in Figs. 1 and 2.

Figure 1 is for a Society of Automotive Engineers 1080 steel. These curves are typical of the straight carbon steels. A fast cooling rate is required down to a temperature of approximately 600 F (316 C) to avoid upper transformations to undesirable ferrite and bainite.

The T T-T curves in Fig. 2 are for a S.A.E. 4340 steel. They illustrate the effect of alloy composition in increasing the allowable cooling time to obtain a full transformation to martensite.

Transformation curves for the majority of production-type steels will fall somewhere between the curves in Figs. 1 and 2.

Section Thickness

Another variable affecting transformation is section thickness. As the thickness increases the cooling time increases. Also, density of the load in a batch-type furnace has a similar effect unless the velocity of the quenching media (agitation) is sufficiently high to offset the retarding effect of the sectional thickness.

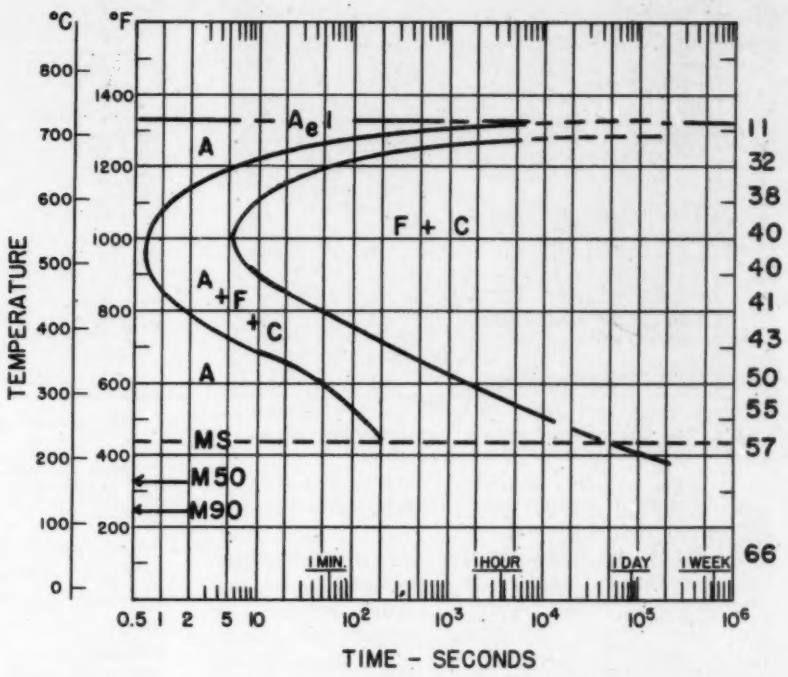


Fig. 1 — Transformation curves for a S.AE. 1080 steel, with alloy elements — C-0.79 per cent and Mn-0.76 per cent, austenitized at 1550 F (843 C). Grain size — 6.

To meet the problems imposed by steel composition variables – grain size, geometry and section – three types of quenching are used. They are generally known as the conventional quench (Fig. 3), the martemper quench (Fig. 4) and the modified marquench (Fig. 5).

CONVENTIONAL QUENCH

This quench (Fig. 3) is the most commonly used, and may employ water, brine, caustic or oil as the quenching medium. The quenching medium temperature ranges from 70 F - 100 F (21 C - 38 C) for the water-type media, and from 125 F - 170 F (52 C - 77 C) for oil. The selection of one of these quenching media requires careful consideration. In the majority of applications where it is necessary to process a variety of steel compositions, oil is the more generally used quenching medium. Oil is more suitable for a wide range of steel compositions as it causes less distortion and cracking. Oil can be

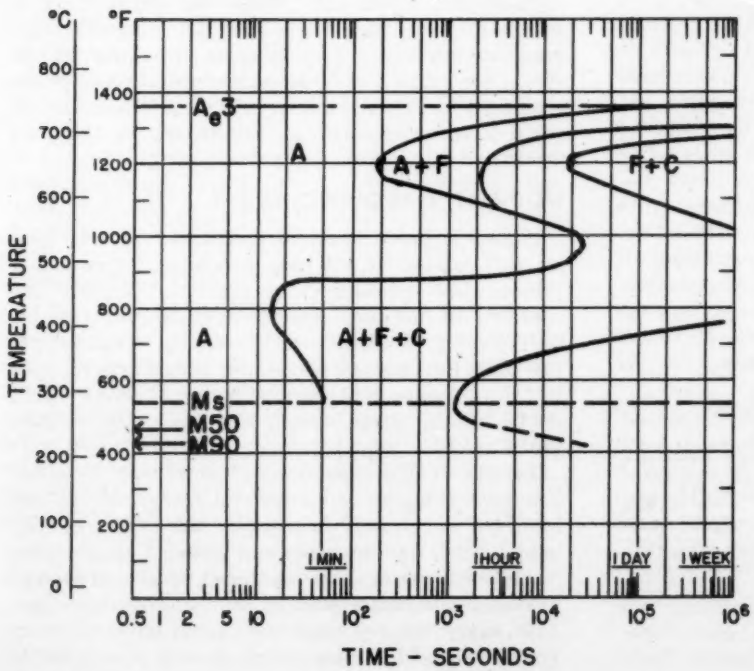


Fig. 2 — Transformation curves for a S.A.E. 4340 steel, with alloy elements — C-0.42 per cent, Mo-0.33 per cent, Ni-1.79 per cent, Mn-0.78 per cent and Cr-0.80 per cent, austenitized at 1550 F (843 C). Grain size — 7-8.

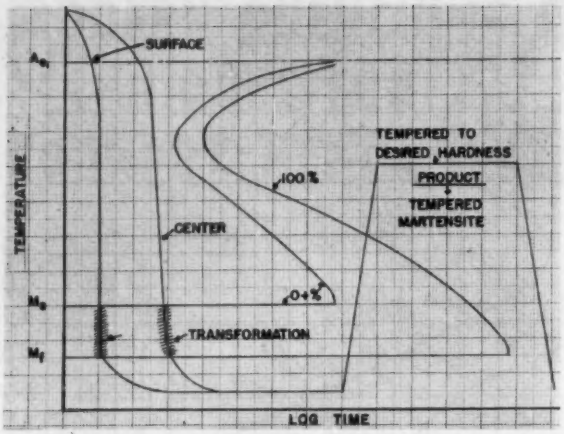


Fig. 3 — Conventional quench and tempering curves used to meet problems of steel variables.

used at higher temperatures thereby reducing the degree of thermal shock.

Oil being the more versatile quenching medium has encouraged considerable investigation and development in recent years to further increase its quenching power, operating temperature range and stability for long service. One additive for quenching oil with which the author has worked is a proprietary additive that, when added in proper ratio to selected quenching oils, has the ability to approximately double the cooling power of the quenching oil from 1550 to 700 F (843 to 371 C). At 700 F (371 C) the rate of cooling reduces rapidly. This is a desirable feature since it permits more equalization of temperature between the core and surface of the work before complete hardness is obtained.

Case History

A comparison of Brinell hardness readings before and after the use of the quenching oil additive is shown in Table 1. Increased cooling power was obtained with the additive, with a resulting increase in quenching out hardness. A wide range of diameter of bar stock was used for the three alloy compositions at this company.

The bar stock was heat-treated for machinability by a commercial heat-treating company. The "Before" hardness was obtained with the quenching oil in regular use at the company. The "After" hardness was obtained after the additive was used in the existing conventional quenching oil. No other changes were made in equipment or agitation.

TABLE 1 – BRINELL HARDNESS READING COMPARISON

	S.A.E. 4140		S.A.E. 4150			S.A.E. 4340	
Size in.	Bhn Before	Bhn After	Bhn Before	Bhn After	Size in.	Bhn Before	Bhn After
1	429	495	477	555	1½ x 3	477	534
1½	415	495	477	578	2 x 4	429	487
2½	415	477	461	534	3 x 7	387	444
3	401	444	444	495			
5	388	444	444	514			
7	388	444	415	495			
8	375	429	388	444			

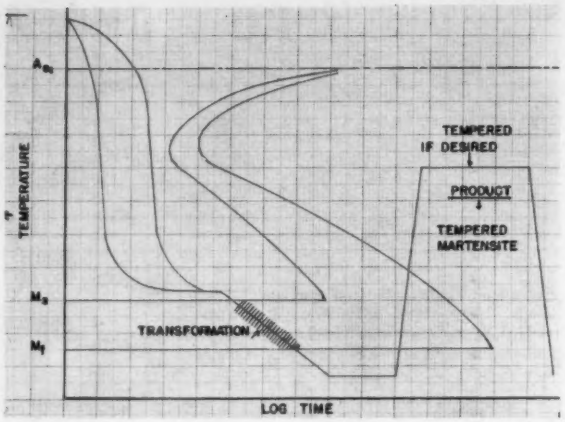


Fig. 4 — Marquenching curves for steel.

Draw temperatures had to be increased from 100 to 125 F (38 to 52 C) over those previously used in order to temper to the required hardness. This greatly improved machinability.

These comparisons (Table 1) demonstrate that although agitation is extremely important, agitation alone does not compensate for the difference in cooling power of quenching oils. The table also shows the effect of section thickness on the obtainable quenched-out hardness.

MARQUENCHING

Figure 4 shows marquenching which is used primarily when control of distortion is of major importance. Because of the high temperature of the quenching medium required in this process, salt is usually used. The cooling power of fused salt is somewhat less than oil, therefore, marquenching in salt may be limited by the composition of the steel and the size of the part. Where the temperature of the bath must be 400 F (204 C) or above, salt must be used as a quenching medium due to the short life of oil at this temperature. Where the temperature of the bath is 350 F (177 C) or lower, oil is a better quenching medium due to the poor fluidity of salt in this temperature range.

MODIFIED MARQUENCHING

Figure 5 shows modified marquenching, the most recently adopted quenching procedure. This is one process that promises to be the most versatile and effective in meeting today's rigid heat-treating requirements. Modified marquenching employs the use of a fast quench oil as the quenching medium in a temperature range between 200 F - 300 F (93 C - 149 C), but usually between 225 F - 260 F (107 C - 127 C).

In this temperature range, control of distortion has been comparable to regular marquenching and hardness is greatly improved. Many of the late model batch-type furnaces and some of the continuous-type furnaces are equipped with immersion heaters for accurate control of the oil temperature. Also, many existing quenching tanks have been converted to modified marquenching at low cost by the use of immersion heaters, steam coils or radiant tubes.

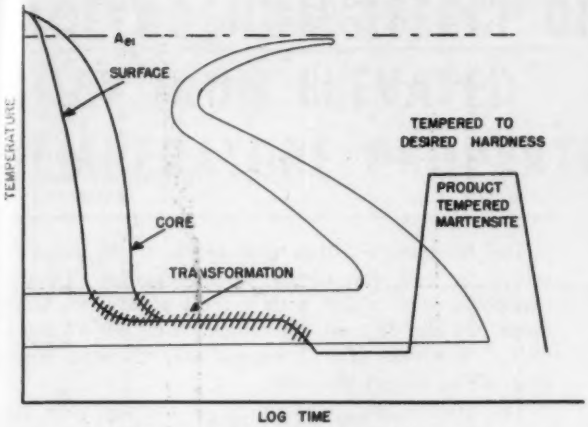


Fig. 5 — Modified marquenching curves for steel.

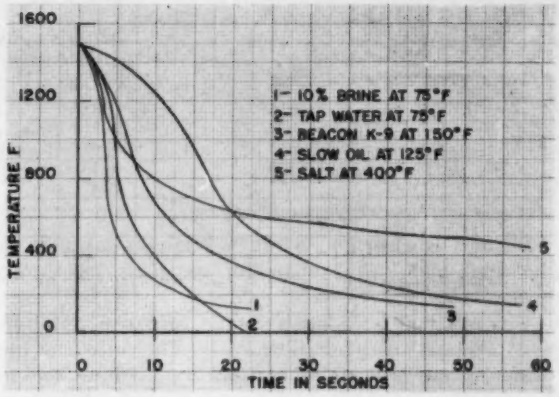


Fig. 6 — Cooling curves for a metal immersed in different quench media.

Control of oil temperature is always important for the best quenching results in either the conventional quench or modified marquench.

What, then, is the most ideal quench to offset the many variables with which the heat treater must contend, and yet produce maximum physical properties and dimensional control? It has been said, and correctly so, that the ideal quench is one which has high cooling power in the initial or early portion of the quench and a slow cooling rate in the lower temperature range.

Figure 6 shows the cooling curves of metal immersed in all the quenching media previously mentioned. The water and brine media have the highest initial cooling speed, but continue to cool rapidly throughout lower temperature ranges where distortion and cracking usually occur. This is why they are generally used for shallow hardening steels, simple geometries or large, heavy cross-sections.

Conventional quenching oils have a much slower cooling rate both in the initial and intermediate range, but the cooling power is not adequate to obtain maximum hardness in the straight carbon steels and some of the lean alloy steels, except with light sections. However, the slower cooling in the lower temperature range will permit a more gradual transition, and have less tendency to cause distortion and cracking than water or brine.

Additive-Type Oils

The additive-type quenching oils for the conventional quench, or for the modified marquench have high cooling power in the initial and intermediate range sufficient to miss the nose of the T-T-T curves on most compositions of steel. In the lower temperature range, the cooling rate is sufficiently slow to permit gradual transition to martensite, thereby producing maximum hardness with a minimum of distortion.

The most common misconception of fast quenching oils has been that they will cause more distortion and cracking than a slow quenching oil, the reasoning being that since we get more distortion and cracking with water, any quenching media which make steel harder will produce similar results.

This is not a fact, because a fast quenching oil having better wetting action reduces the vapor-phase portion of the quench. This causes more even cooling in the initial portion of the quench and reduces stresses that are caused by vapor pockets.

In the lower temperature range where water or any water composition quenching media continues to cool rapidly, fast quenching oils cool more slowly and permit better equalization of temperature between light and heavy sections before transformation is completed.

Agitation of a quenching medium is important regardless of which medium is being used. Violent agitation produced either by high capacity pumps or propellers, if properly directed, can greatly increase the cooling power of any medium. This is especially important when using fast quenching oils. Because of the ability of fast quenching oils to extract heat more rapidly, the hot oil must be moved away from the surface of the work and distributed throughout the quenching bath. A practical study of proper agitation that can be adapted to almost any quenching facility can be found in the booklet *Improved Quenching of Steel by Propeller Agitation,* published by the United States Steel Corp.

A fast, additive-type quenching oil, operating in an oil temperature range of 200 F - 300 F (93 C - 149 C) with adequate agitation, has proved to be a combination which has offset the majority of the variables with which the heat treater has to contend. While it obviously does not eliminate all of the problems, it has improved the quality control beyond any point that has previously been attained.

A second case history shows how many variables have been offset by this method. This company is a large automobile parts plant making steering gears. A steering gear shaft of S.A.E. 6120 steel carburized was quenching out with a spotty hardness and excessive distortion.

The automotive gears were on shafts of from 1¼ - 1⅜ in. diameter by 5 - 7 in. long. The gears were carburized for 12 hr in a surface combustion furnace, loaded vertically in trays and quenched directly through a protective atmosphere at 100 F (38 C) in their regular quenching oil.

TABLE 2 – TYPICAL ROCKWELL HARDNESS READINGS

Gear No.	Rockwell C Hardness
1	51-54-49-54
2	41-48-48-46
3	49-51-61-54
4	42-43-58-49
5	40-45-60-48
6	51-41-58-45

Typical Rockwell C hardness readings taken from the shaft 90 F (32 C) apart for six gears are shown in Table 2.

Changes had been made in the furnaces, agitation and processes used for these gears. The only way the company was able to get satisfactory hardness was to lay the parts on their sides, but this caused distortion of the gears which exceeded the passable limits.

The plant metallurgist decided that the vapors created when the hot trays and parts hit the quenching oil at 1650 F (899 C) were the cause of the nonuniform hardness readings. To offset this, the proprietary additive concentrate previously mentioned was added to the oil already in the furnace. The effects of the vapor phase of quenching was reduced, and the Rockwell C hardness readings were increased, as shown in Table 3.

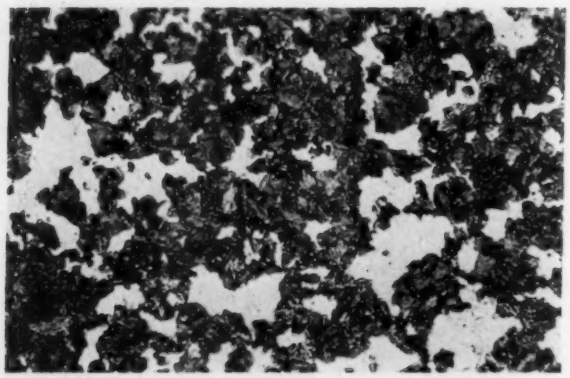


Fig. 7 — Bainitic microstructure (structure at surface is untempered martensite with nodular troostite) of automobile steering gear shaft made of a S.A.E. 6120 steel carburized and conventional quenched. Hardness ranged from 40-61 Rockwell C. 500 ×.

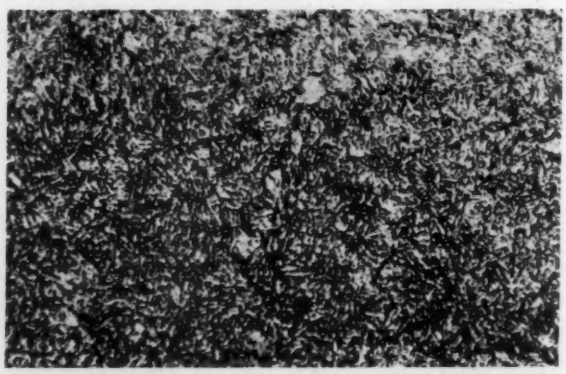


Fig. 8 — Martensitic (100 per cent) microstructure of automobile steering gear shaft made of a S.A.E. 6120 steel carburized and quenched with additive in quenching oil. Hardness ranged from 63-66 Rockwell C. 500 ×.

TABLE 3 – ROCKWELL C HARDNESS AFTER ADDITIVE USE

Gear No.	Rockwell C Hardness
1	64-65-66-67
2	67-66-62-66
3	65-66-66-64
4	66-66-66-66
5	64-64-65-66
6	65-64-64-64

The hardness readings were taken at the same intervals as with the regular quenching oil. Twenty-two parts were tested, with a total of 88 tests being made. Of the 88 tests, 83 ranged from 64 - 67 Rockwell C hardness. The hardness did not drop lower than 62 in any of the tests.

The engineering department of the company has now set 58 Rockwell C hardness as the minimum acceptable to improve wearing surfaces of gears.

Figure 7 shows the type of microstructure (bainitic) which was being obtained with this company's conventional quench. The previously mentioned additive was put in the existing quenching oil with no other changes in procedure. Figure 8 shows the type of microstructure (martensitic) obtained with quenched-out hardness after using the additive.

With the uniformity of quenched-out hardness, inspection was reduced from a 100 per cent check to a spot check, and straightening costs in man-hours reduced by 75 per cent. Such savings are obviously far greater than the total cost of a quenching oil, the loss usually amounting to between 2 and 4 quarts drag out per ton of work heat treated.

SUMMARY

The benefits to be obtained by good quenching techniques may be summarized:

1) Economical service life of the quenching medium.
2) Uniform hardness pattern.
3) Elimination of undesirable transformation structures.
4) Development of optimum quenching conditions for steels of small or varying grain sizes, different hardenabilities and variable cross-sections.
5) Minimum dimensional changes.
6) Elimination of quenching cracks.

Contacts with heat-treating departments throughout the country show that there are many of them not taking advantage of improvements in equipment and quenching oils now available. In most cases, personnel responsible for heat treating recognize the benefits which can be obtained, but other branches of management that control appropriations are not familiar with heat treating or aware that heat treating has become a precision operation.

There are many case histories where improvements in equipment, or the adoption of a more efficient quenching oil, has eliminated re-runs and scrap in heat treating and reduced distortion with subsequent straightening costs, or permitted closer preliminary machining which reduced final grinding time thereby reducing processing costs in other operations.

MOLYBDENUM EFFECT ON GRAY IRON ELEVATED TEMPERATURE PROPERTIES

By G. K. Turnbull and J. F. Wallace

ABSTRACT

The effect of up to 2.0 per cent molybdenum on the short-time tensile and stress-rupture properties of a 3.8 per cent carbon equivalent gray iron has been determined by various additions of that alloy to two base irons: one with only residual alloys; and another containing approximately 0.60 per cent chromium. The gray iron was melted in a cupola, transferred to an electric arc furnace for holding and preheating, and alloyed as required by additions to the pouring ladles. The tests were performed on 0.505 in. diameter tensile specimens that were machined from 1.2 in. diameter bars cast in baked core sand molds. Short-time tensile tests were conducted at room temperature, 800, 1000 and 1200 F; the stress-rupture strength was measured for fracture times up to 100 hr at 800, 1000 and 1200 F.

Increasing molybdenum additions up to 2.0 per cent generally resulted in greatly improved high-temperature properties at all testing temperatures. Some decrease in short-time tensile strength occurred with the higher molybdenum irons, particularly those containing chromium, because of the presence of considerable massive cementite. The stress-rupture properties of the irons containing chromium were superior to the straight molybdenum irons for all compositions at 1200 F and for those irons with less than about 1.5 per cent molybdenum at 800 and 1000 F. The rupture strength of the higher molybdenum-chromium irons at 800 and 1000 F was reduced below the values for the comparable molybdenum irons, however, by the presence of massive cementite. Metallographic examination of the irons before and after the 100 hr stress-rupture testing indicated: some bainites were slightly decomposed at 800 F; all bainites were decomposed at 1000 F; and all bainites and pearlites were agglomerated and spheroidized at 1200 F.

INTRODUCTION

Current A.S.M.E. Codes limit the use of gray iron in unfired pressure vessels to 650 F for stressed applications. The problem of raising the allowable temperature limit for this material for load bearing parts has been of concern to the foundry industry.[1] More data on the high temperature properties of gray iron are required to provide the evidence necessary to encourage application of gray iron castings at higher temperatures.

G. K. TURNBULL is Rsch. Asst. and J. F. WALLACE is Assoc. Prof., Dept. of Met. Eng., Case Institute of Technology, Cleveland.

Available data for unalloyed gray iron show that the short-time high-temperature strength is reduced at temperatures in excess of 700 F.[2] However the decrease in short-time tensile strength does not appear rapid until the temperature exceeds 850 F.[3-7] In addition to the decrease in strength with temperatures, it has been shown that the modulus of elasticity of gray iron is decreased to some extent and the ductility increased slightly.[5-6]

The short-time high-temperature strength of gray irons is considerably improved by addition of alloys. Molybdenum and chromium are particularly effective in this respect.[4,6,7] In addition to increasing the room temperature strength,[2,8,9,10] these elements have a stabilizing effect on the combined carbon in the matrix and retard the rates of softening and of secondary graphitization.[2,7,10,11,12] In this manner, these elements also reduce growth of the cast iron at elevated temperatures.[2,12-16]

Creep and stress-rupture information are of considerably more importance than short-time tensile tests when applications involving extended elevated temperature exposure are considered. The creep rate of unalloyed gray iron at temperatures over 800 F is fairly high[2,7] compared to some other ferrous materials for use at high temperatures. The stress-rupture strength of unalloyed gray iron is also appreciably reduced by temperatures in excess of 800 F. Both creep and stress-rupture properties can be markedly improved by the addition of alloying elements, particularly molybdenum. It has been shown that molybdenum was the most effective alloying element for improving these high temperature properties of gray irons in a recent series of tests at 800 and 1000 F.[7] The beneficial influence of molybdenum on the stress-rupture properties of several gray irons has been shown for other work conducted at 750 F[6] and 800, 1000 and 1200 F.[9,17] Some of the results of this latter, heretofore unpublished work, are reported in this paper.

It has been shown in these references that increasing additions of molybdenum resulted in a steady improvement in high-temperature properties. The results were limited, however, and reported at scattered temperatures and for a wide variety of irons.

The investigation described in this report was undertaken to provide a detailed study of the effect of

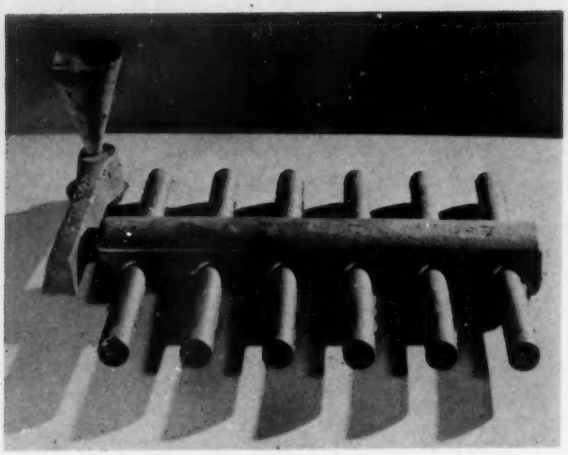


Fig. 1 — Cast coupons showing gating and risering.

molybdenum on the high temperature properties of irons of a similar type over the temperature range of greatest commercial interest, 800-1200 F. The base analyses and type of iron employed in this investigation were maintained at a reasonably constant level, and various amounts of molybdenum were added to these irons so that the effect of this element could be observed. A Class 40 gray iron was employed as the base composition. Molybdenum additions up to 2.5 per cent were made to two series of this iron: first, to the base iron; and second, to the base iron containing about 0.60 per cent chromium. The high temperature tests conducted on these irons were limited to short-time tensile tests and stress-rupture tests up to a fracture time of 100 hr.

PROCEDURE

The gray iron was melted in a cupola and then transferred to a direct electric-arc holding furnace for mixing and preheating. The molten iron was tapped from the electric furnace in weighed amounts into a ladle suspended on a scale. The additions of ferro-alloys and inoculants were made to the small ladle as the weight of molten metal required for each analysis was tapped. The temperature in the electric furnace was controlled at approximately 2800 F before tapping. The charge in the cupola consisted of steel and gray iron scrap, pig iron and dolomite, with small additions of ferrosilicon and ferromanganese. The constituents of the charge were selected to avoid alloy residuals. No additions were made to the electric furnace. Because of the large amount of metal required and limited capacity of the electric furnace (4000 lb), continual additions of molten metal were made to the furnace from the cupola holding ladle as required.

All compositions were inoculated during teeming from the electric furnace by additions to the stream of 0.30 per cent silicon as 85 per cent regular ferrosilicon (containing 1.5 per cent Al). Approximately 0.03 per cent carbon as electrolytic graphite was added to the stream on the composition containing over 1.5 per cent molybdenum. Some additions of mexican graphite were also made to the ladle spout on the compositions containing over 0.56 per cent molybdenum.

After preheating, alloying and inoculating, the molten iron was poured into horizontal molds produced from baked oil-bonded sand. The pouring temperature was approximately 2600 F. The test bars employed in this investigation were cast as 1.2 in. diameter, 6 in. long bars from a central runner that also served as a riser. The arrangement of 12 bars about this central runner-riser and the special gating system employed are shown in Fig. 1. The test bars were permitted to remain in the molds overnight (14 hr) before shakeout.

A considerable number of coupons were poured for each composition. It was desired to vary the molybdenum content in steps of 0, 0.25, 0.50, 0.75, 1.00, 1.50, 2.00 and 2.50 per cent in the base iron for one series and in the base iron containing 0.60 per cent chromium for the second series.

After shakeout, the coupons were broken from the riser and machined to standard 0.505 in. diameter threaded tensile bars with 2 in. gage lengths. The diameter of the gage section was slightly and gradually reduced at the center of the gage length to facilitate fracturing at the position where the lowest thermal gradient occurred. Bars from the three higher molybdenum analyses of the chromium series and the two higher molybdenum analyses of the plain series were annealed at 1400 F for 1½ hr before machining in order to soften the matrix to some extent. One bar from each series was employed for chemical analysis and another was split longitudinally through the center and etched for a determination of soundness. A hardness survey was taken along the length of this bar.

Short-time tensile tests were conducted on a 60,000 lb tensile machine. A three-coil electric resistance furnace was used in conjunction with an automatic controller to provide temperature control. Thermal gradients from the ends of the specimen to the center and thermal fluctuations during testing were measured with three thermocouples and held to ±5 F throughout the tests. After reaching the testing temperature, a 30 min stabilization period was permitted before loading was initiated.

Stress-rupture tests were conducted on two lever-arm machines of the dead-weight type, and a screw-type creep-rupture machine with a calibrated stiff spring operating a screw mechanism. All three machines were of the constant load type and varied only in their method of load application. Thermocouples were placed at the bottom, center and top of the gage length section of each specimen before heating. Temperature fluctuations and gradients were held to ±2 F. Before testing, a 30 min stabilization period was provided to permit the specimen to reach thermal equilibrium. The load was then applied and the time to fracture recorded.

Metallographic examination was employed to examine the "as cast" or annealed microstructure of each composition, and to observe the effect of holding under load for approximately 100 hr at each temperature. Tests from the cast and annealed specimens were removed from the center of the coupon. The metallographic specimens from the bars which ruptured after 100 hr were taken from a section perpen-

TABLE 1 – CHEMICAL ANALYSIS

% C	% Si	% Mn	% Mo	% Cr
	Straight Molybdenum Irons*			
3.22	1.88	0.78	0.05	0.16
3.32	1.78		0.22	
3.26	1.73		0.50	
3.28	1.84		0.76	
3.28	1.68		1.00	
3.09	1.82		1.54	
3.01	1.82		2.02	
3.03	1.78		2.25	
	Chromium-Molybdenum Irons*			
3.31	1.76	0.73	0.12	0.64
3.37	1.76		0.28	0.64
3.23	1.88		0.47	0.61
3.21	2.06		0.77	0.66
3.23	2.03		0.96	0.68
3.15	1.95		1.42	0.53
3.13	1.81		1.95	0.64
3.16	1.82		2.19	0.55

* S – 0.072%, P – 0.046%, Ni – 0.17%, Cu – 0.07%

dicular to the fractured surface of the bar at the center line of the tensile specimen.

RESULTS AND DISCUSSION

Chemical Analysis

The chemical analyses of the different compositions of gray iron investigated are listed in Table 1. These compositions were generally close to those selected and present a good coverage of the range of molybdenum contents from 0-2.5 per cent. The only serious variations in compositions were:

1) a lower molybdenum content of the highest addition, 2.25 and 2.19 per cent reported in each series when 2.50 per cent was desired,
2) a higher chromium residual of 0.15-0.16 per cent in the first series, and a variation in chromium from 0.53-0.68 per cent in the second,
3) a residual molybdenum content of from 0.05-0.12 per cent for the compositions that were to represent 0 per cent molybdenum,
4) variations in carbon from 3.01-3.37 per cent and in silicon from 1.73-2.06 per cent. While these variations in carbon and silicon undoubtedly had some effect on the strength, they generally occurred gradually as the analysis changed, and none of the compositions was far from a carbon equivalent value of 3.8 per cent.

Soundness and Hardness Survey

The results of the longitudinal sectioning and hardness survey of the bars from each composition indicated that no shrinkage, and little variation in hardness, occurred in the bars with any composition. It was believed that the riser was entirely adequate to avoid any shrinkage, including microporosity, particularly at the low phosphorus level of 0.044-0.047 per cent.[18,19] Even after grinding and etching, the bars showed no shrinkage during visual and microscopic examination. As was anticipated, a slight effect of annealing can be observed, however, in the results of the hardness surveys of the as-cast bars presented in Table 2. The only appreciable change in hardness occurred from the faster cooling rate of the extreme

TABLE 2 – HARDNESS SURVEY OF AS-CAST TEST BARS

Chemical Analysis		Brinell Hardness Number		
Mo	Cr	Ave.	Gate End	Chill End
0.05	0.16	229	241, 228, 228,	223, 228
0.22		243	235, 235, 241,	248, 255
0.50		248	241, 241, 248,	248, 262
0.76		257	248, 255, 255,	269, 269
1.00		251	255, 241, 241,	255, 262
1.54		306	302, 302, 302,	321, 302
2.02		304	293, 293, 302,	311, 321
2.25		372	302, 365, 351,	365, 375, 402, 444
0.12	0.64	258	255, 255, 255,	262, 262
0.28	0.64	277	269, 262, 269,	286, 302
0.47	0.61	273	255, 255, 269,	286, 302
0.77	0.66	289	286, 277, 293,	286, 302
0.96	0.68	300	286, 286, 286,	311, 332
1.42	0.53	330	332, 302, 311,	332, 375
1.95	0.64	371	351, 365, 332,	340, 351, 418, 444
2.19	0.55	382	375, 365, 365,	365, 375, 387, 444

TABLE 3 – ANNEALING TREATMENT EFFECT ON HARDNESS

Iron Composition		Average Brinell Hardness Number	
% Cr	% Mo	As-Cast	After Anneal
0.53	1.42	330	268
0.64	1.95	371	279
0.55	2.19	382	271
0.16	2.02	304	261
0.16	2.25	372	263

end of the bars located farthest from the gate. The 5½-in. long tensile specimens were located toward the gate end of the 6 in. long coupon so that this hard end was not within the specimens.

The gray irons with higher molybdenum contents in both series were annealed at 1400 F for 1½ hr before machining the tensile bars. This annealing treatment was conducted to soften the matrix structure so that the machining would be facilitated. It did not appear to decompose the massive carbides but tempered the matrix. The effect of this treatment on the hardness is shown in Table 3.

Short-Time Tensile Tests

The short-time tensile tests indicated that increasing molybdenum content increased the strength of the plain and 0.60 per cent chromium gray irons to a maximum value at 1.4-1.9 per cent molybdenum. These results are shown in Fig. 2 for the iron without a chromium addition, and in Fig. 3 for the iron containing 0.60 per cent chromium. The room temperature strength of the plain irons increased steadily at nearly 2000 psi for each 0.1 per cent molybdenum added up to 1.7 per cent molybdenum. This rate of increase is similar to some irons and slightly greater than others reported.[20] The rate of increase of the room temperature strength of the iron containing chromium was approximately the same up to 0.6 per cent molybdenum additions, but leveled off to only a slight increase in strength for larger additions.

The increase in short-time tensile strength was approximately the same for equivalent molybdenum additions to both irons at 800 and 1000 F testing temperatures. The actual level of strength, particularly

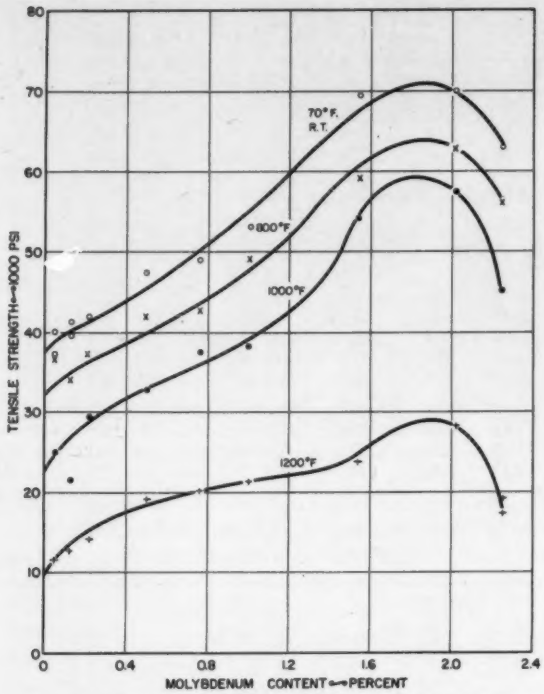


Fig. 2 — Tensile strength as a function of molybdenum content for gray iron at different temperatures.

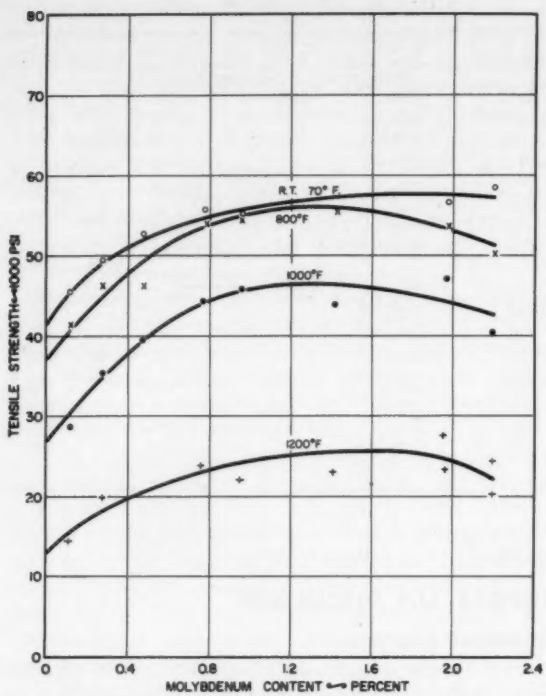


Fig. 3 — Tensile strength as a function of molybdenum content for gray iron with 0.6 per cent chromium addition at different temperatures.

at 1000 F, was somewhat lower than at room temperature. The increased softening effect at higher temperatures is shown by the short-time tensile test results of those specimens fractured at 1200 F. Not only is the strength level considerably lower, but the rate of increase of strength with increasing molybdenum content is less for both irons at 1200 F.

The decrease in strength of both irons at the highest molybdenum contents was believed to be the result of the embrittling effect of considerable quantities of the massive cementite obtained upon solidification of these irons, particularly in the series containing chromium. The presence of this cementite in the microstructure was confirmed by microexamination as discussed in a subsequent section of this paper.

Stress-Rupture Tests

The results of the stress rupture tests conducted on the plain iron with various molybdenum additions are shown in Figs. 4, 5 and 6 for testing temperatures of 800, 1000 and 1200 F, respectively. The stress-rupture curves for the irons containing about 0.60 per cent chromium are illustrated in Figs. 7, 8 and 9 for the testing temperatures of 800, 1000 and 1200 F, respectively. All data are presented on plots of logarithm of initial stress vs. the logarithm of the time to fracture. From six to nine specimens were employed to establish the curve for each composition at each temperature. The data are consistent over the entire time and temperature range. To improve the clarity of the figures, individual points were not plotted; only the straight lines through these points are shown. The time to failure was extended to 100 hr for each curve by extrapolation or interpolation.

The stress-rupture curves of all irons did not show noticeable deviation from linearity within 100 hr at any of the temperatures investigated. Although decomposition of the matrix of some irons was detected at some of the temperatures, it was believed that the change was gradual and, therefore, did not cause a sudden discontinuity in the curve. Others investigating the stress-rupture characteristics at longer times and similar temperatures observed definite changes in slope of some of the curves beyond 100 hr.[7] This was attributed to decomposition of some structure phases.

The decrease in stress-rupture strength with increasing time to fracture was similar for the irons with and without chromium at the different molybdenum contents when tested at 1000 and 1200 F. The comparative performance of the irons at 800 F, however, varied somewhat with the different amounts of molybdenum. When the molybdenum content was between 0.50 and 1.5 per cent, the irons containing chromium did show a slightly smaller decrease in stress with testing times compared to the irons with only a residual chromium content. With a molybdenum addition of about 2.0 per cent, the strength of the iron without chromium decreased at a slower rate at the longer times of stress than the comparable iron with chromium.

This behavior was probably caused by the embrittling effect of the large amount of massive cementite in the high molybdenum iron with a chromium addition. As the testing temperature increased to 1000 and then 1200 F, the stress rupture strength of all irons was reduced considerably and the decrease in strength with increasing time of testing was more marked.

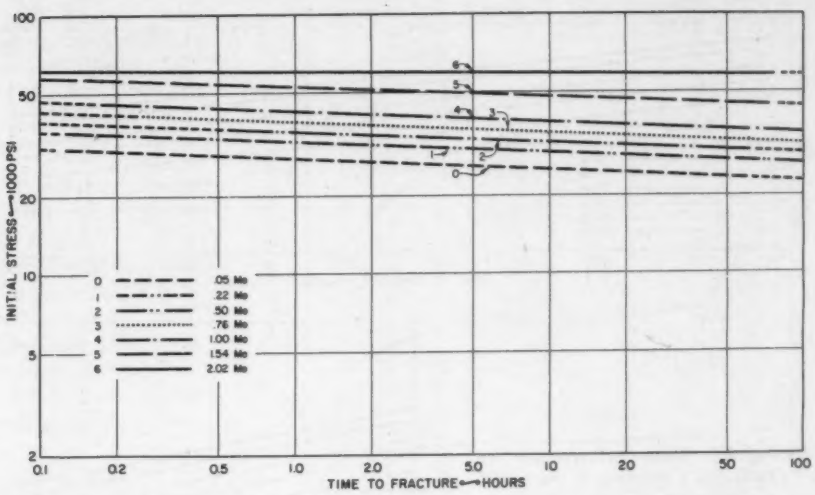


Fig. 4 — Time to fracture at 800 F as a function of the initial applied stress for gray iron with different molybdenum contents.

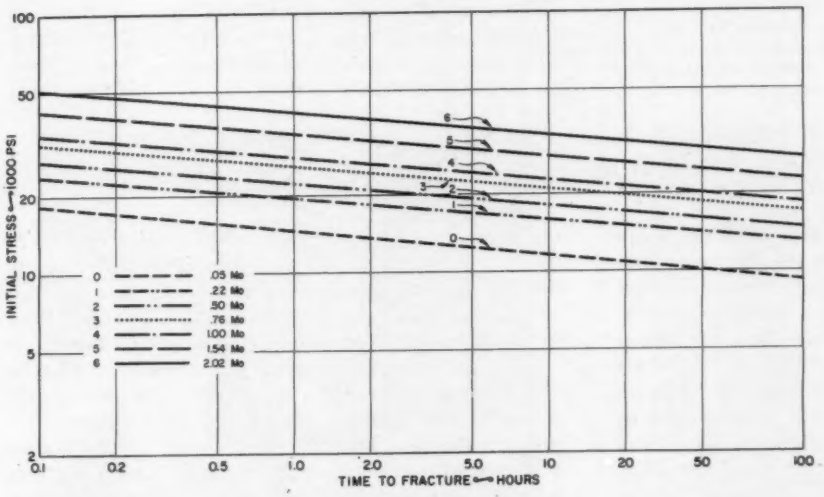


Fig. 5 — Time to fracture at 1000 F as a function of the initial applied stress for gray iron with different molybdenum contents.

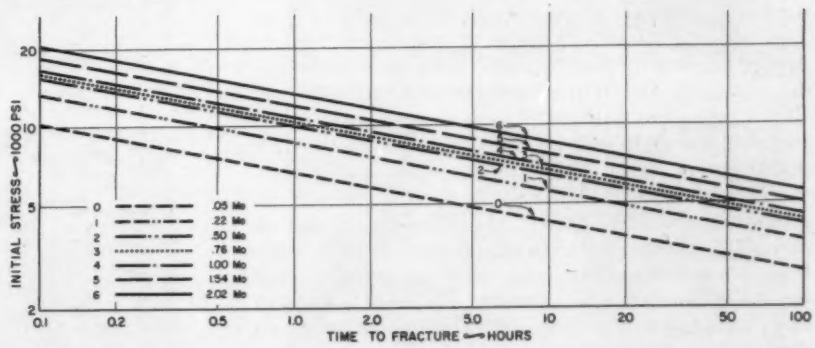


Fig. 6 — Time to fracture at 1200 F as a function of the initial applied stress for gray iron with different molybdenum contents.

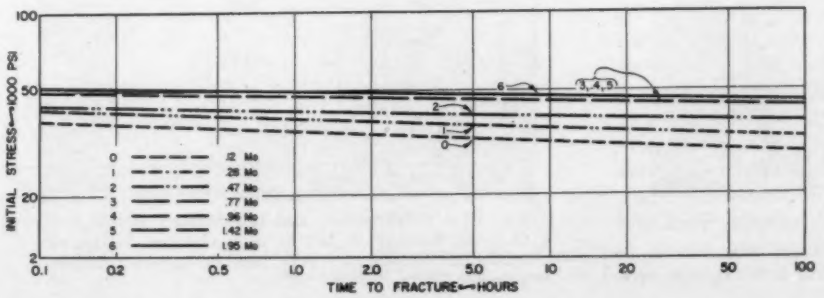


Fig. 7 — Time to fracture at 800 F as a function of the initial applied stress for gray iron with 0.6 per cent chromium and different molybdenum contents.

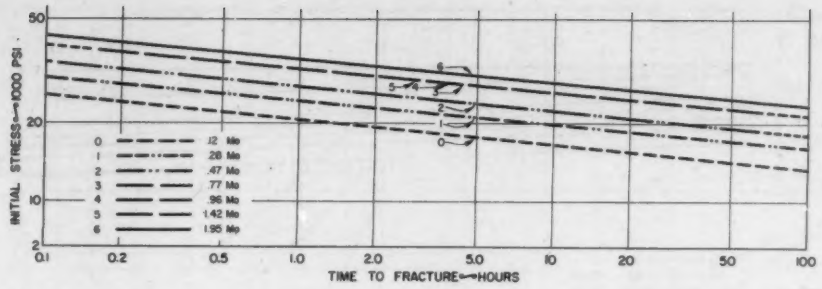


Fig. 8 — Time to fracture at 1000 F as a function of the initial applied stress for gray iron with 0.6 per cent chromium and different molybdenum contents.

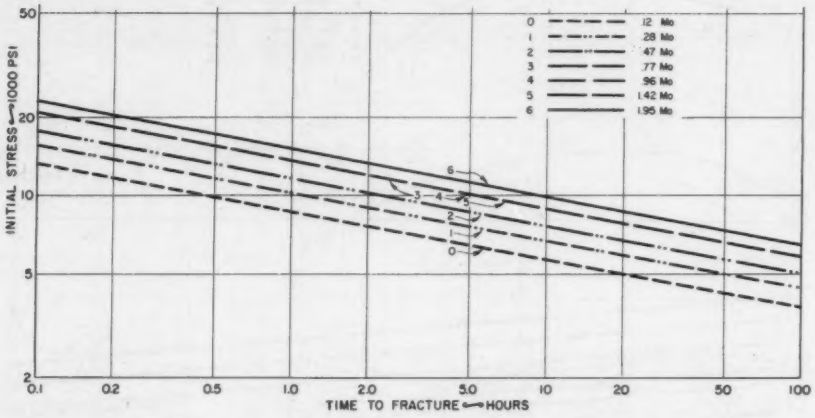


Fig. 9 — Time to fracture at 1200 F as a function of the initial applied stress for gray iron with 0.6 per cent chromium and different molybdenum contents.

The addition of 0.50 per cent molybdenum is apparently more beneficial in improving the stress-rupture properties at all temperatures than a similar amount of chromium. The curve for the iron containing 0.50 per cent molybdenum and 0.16 per cent chromium was consistently higher than the curve for 0.05 per cent molybdenum and 0.64 per cent chromium in Figs. 4-9. The molybdenum was believed to have strengthened the pearlitic matrix, whereas the chromium caused the formation of massive carbides.

While molybdenum consistently improved stress-rupture strength throughout the series of irons in Figs. 4-9, the relative behavior of the higher molybdenum irons with and without chromium additions depended on the testing temperature. At 800 and 1000 F, the irons without chromium and with a 1.54 and 2.02 per cent molybdenum addition exhibited a higher stress level in the stress-rupture curves than the corresponding irons in the series containing chromium. However, at 1200 F, the stabilizing effect of chromium on the matrix was apparent and the stress level of the chromium containing irons was higher for the irons with approximately 1.5 and 2.0 per cent molybdenum than for the comparable molybdenum content irons in the series without a chromium addition.

The effect of molybdenum additions on the initial stress to cause fracture at 100 hr during stress-rupture testing is shown in Fig. 10 for the irons without a chromium addition, and in Fig. 11 for the 0.60 per cent chromium irons. These curves show the beneficial effects of molybdenum and molybdenum with chromium on the stress-rupture properties. Molybdenum additions to the straight molybdenum irons and the series of irons containing chromium result in better higher temperature performance. As the testing temperature increases from 800 to 1000 to 1200 F, the improvement in strength obtained with molybdenum additions decreases steadily. This was believed to be the result of the structural alterations that occurred in the matrix at higher temperatures.

A comparison of Figs. 10 and 11 illustrates the relative effect of these alloys on the 100 hr stress-rupture strength. The series of irons with 0.60 per cent chromium exhibited better stress-rupture strength than the irons with 0.16 per cent chromium from 0 to about 1.5 per cent molybdenum at 800 and 1000 F. At higher

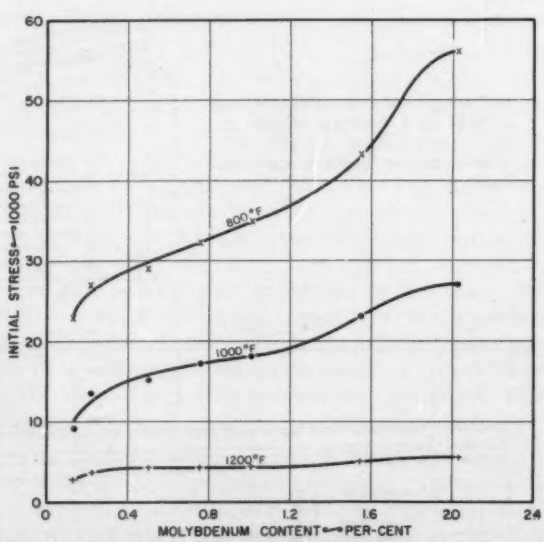


Fig. 10 — Extrapolated and interpolated initial stress to cause fracture at 100 hr as a function of the molybdenum content for gray iron at different temperatures.

molybdenum contents, the irons with residual chromium were superior at these temperatures. At 1200 F, however, the irons with chromium additions were slightly superior to those with residual chromium for all molybdenum contents.

Microstructure

The microstructures of the coupons demonstrate many of the factors that produce the variations in short-time tensile and stress-rupture strength of these irons. The as-cast or annealed microstructures of the plain iron series contained considerably less free or massive cementite than the series containing about 0.60 per cent chromium. The plain base iron with the residual amount of 0.05 per cent molybdenum and 0.16 per cent residual chromium exhibited a microstructure of a pearlitic matrix and graphite flakes, as shown in Fig. 12a.

This structure was not changed appreciably as the molybdenum content was increased to 0.22, 0.50, 0.76 and 1.00 per cent, except for an increase in the fineness of the pearlite. The microstructure obtained with a 1.00 per cent molybdenum content is illustrated in Fig. 12b. When the molybdenum was in-

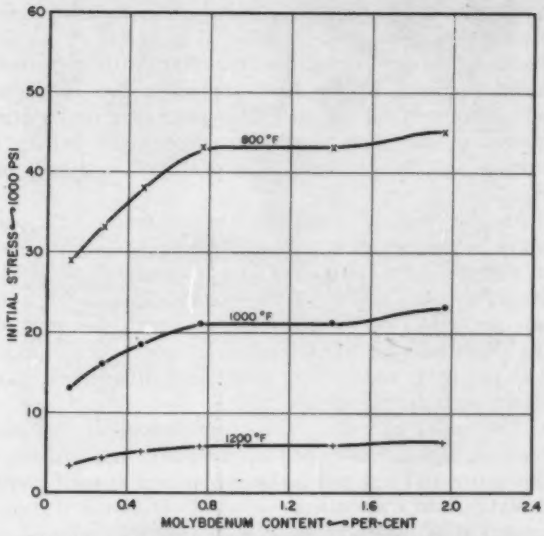


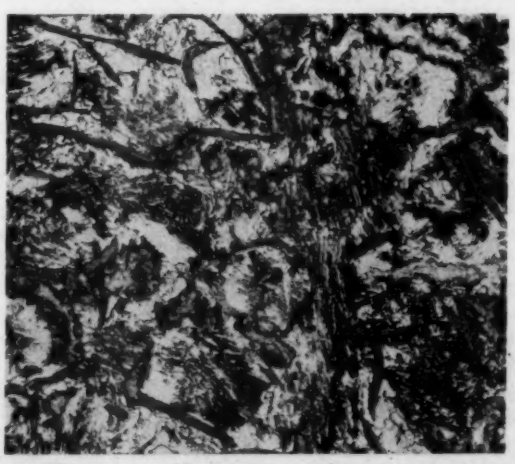
Fig. 11 — Extrapolated and interpolated initial stress to cause fracture at 100 hr as a function of the molybdenum content for gray iron with 0.6 per cent chromium at different temperatures.



a) 0.05 per cent Mo, as-cast



b) 1.00 per cent Mo, as-cast



c) 1.54 per cent Mo, as-cast



d) 2.02 per cent Mo, annealed at 1400 F

Fig. 12 — Microstructures of gray iron at various molybdenum contents with 0.16 per cent chromium, before testing. Picral etch. 250 ×.

creased to 1.54 per cent, the matrix became a mixture of fine pearlite and bainite (Fig. 12c). A small amount of fine cementite together with graphite flakes occurred in the structure when the molybdenum content was increased to 2.02 per cent. The matrix in this iron was approximately 90 per cent bainite and 10 per cent fine pearlite, as shown in Fig. 12d.

At the highest molybdenum content (2.25 per cent) in this series of irons, the structure contained a slightly larger amount of "skeletal" cementite, graphite flakes and a bainite matrix. This latter iron was not subjected to stress-rupture testing because of the slight increase in molybdenum content from 2.02 - 2.25 per cent, and greater machining difficulties associated with the higher hardness.

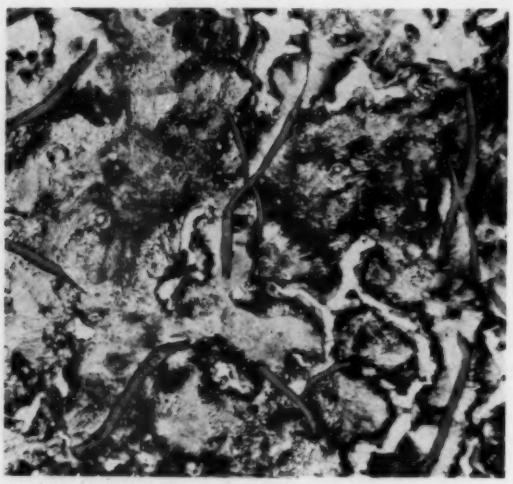
The series of irons with approximately 0.60 per cent chromium contained appreciably more massive cementite and less bainite in the matrix at equivalent molybdenum contents. Prior work[8,10,12] has demonstrated that molybdenum has a greater tendency to change the matrix from pearlite to bainite, and that chromium additions were more prone than molybdenum to produce massive cementite in the structure. In this series, the base iron with 0.12 per cent residual molybdenum and 0.64 per cent chromium contained an appreciable quantity of free cementite together with graphite flakes in a pearlitic matrix, as shown in Fig. 13a.

The size and quantity of the free cementite increased slowly but steadily in this 1.2 in. round section as the molybdenum content was raised to 0.28, 0.47, 0.77 and 0.96 per cent. The microstructure at 0.96 per cent molybdenum is shown in Fig. 13b. In addition to the somewhat larger size and increased amount of massive cementite, the pearlitic matrix was considerably finer than with lower molybdenum percentages. When the molybdenum content was increased to 1.42 per cent in this series the amount of massive cementite increased, but the matrix was still primarily fine pearlite with only small amounts of acicular bainite (Fig. 13c).

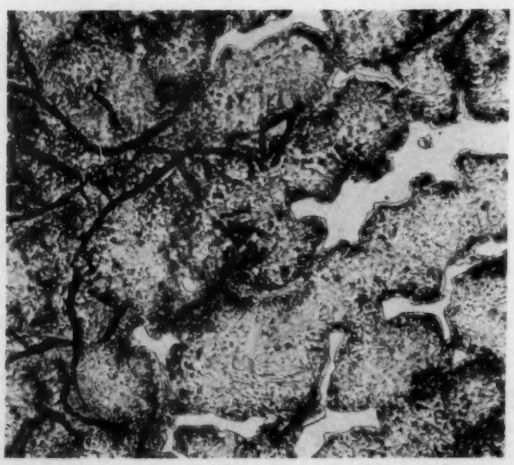
The massive cementite also increased in quantity at 1.95 per cent molybdenum and the matrix became a mixture of 30 per cent fine pearlite and 70 per cent bainite, as illustrated in Fig. 13d. The highest



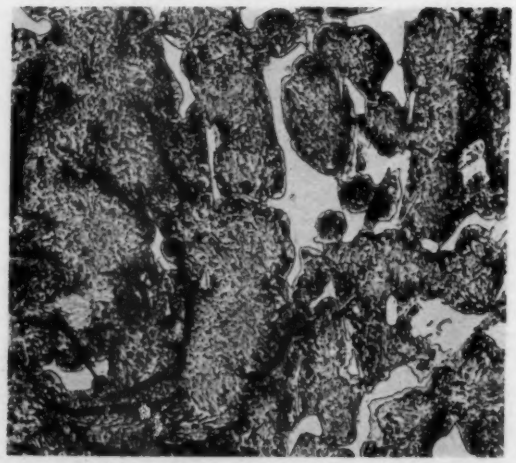
a) 0.12 per cent Mo, as-cast



b) 0.96 per cent Mo, as-cast



c) 1.42 per cent Mo, annealed at 1400 F



d) 1.95 per cent Mo, annealed at 1400 F

Fig. 13 — Microstructures of gray irons at various molybdenum contents with 0.60 per cent chromium, before testing. Picral etch. 250 ×.



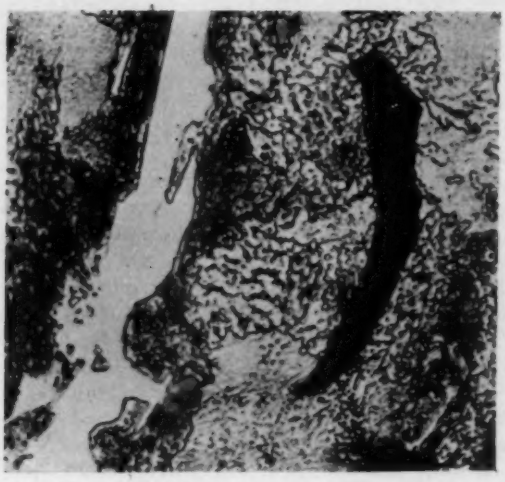
a) 0.05 per cent Mo, 0.16 per cent Cr



b) 0.12 per cent Mo, 0.64 per cent Cr



c) 2.02 per cent Mo, 0.16 per cent Cr



d) 1.95 per cent Mo, 0.64 per cent Cr

Fig. 14 — Microstructures of gray irons of various molybdenum and chromium contents after 100 hr under stress at 1000 F. Picral etch. 1000 ×.

molybdenum iron in this series (2.19 per cent) contained considerable free cementite with graphite flakes in an acicular matrix. Stress-rupture tests were not conducted with this composition because of the high hardness, difficult machining and small increase in molybdenum over the 1.95 per cent molybdenum iron in this series.

Structure Effects on Properties

The influence of structure on short-time tensile strength can be obtained from a concurrent consideration of Figs. 2, 3, 12 and 13. The strength of the series of irons containing chromium is somewhat better than the other base iron at low molybdenum contents for all testing temperatures. As the molybdenum content increases, this difference between the strengths of the irons at equivalent molybdenum contents becomes less. The strength of the two series became equal for the room temperature tests at approximately 1.0 per cent molybdenum, for the 800 and 1000 F tests at 1.4 per cent molybdenum and for the 1200 F tests at 1.6 per cent molybdenum.

At still higher contents than those listed the irons without a chromium addition were generally superior in strength. The strengthening influence of both chromium and molybdenum were apparent at lower molybdenum contents. However, with the larger additions, the embrittling effect of the greater quantities of massive cementite reduced the strength of the series containing chromium. The higher molybdenum irons without chromium additions were strengthened by the greater amounts of bainite in the matrix. The stabilizing effect of chromium on the matrix structure accounted for the better comparative performance at the higher testing temperatures of the series containing this alloy.

The results of microstructural examination of the stress-rupture bars after loading for 100 hr at various temperatures indicate the effect of these conditions on the structures of the different irons. An analysis of these structures and the changes occurring during the tests explain many of the variations in stress-rupture strength shown in Figs. 4-11. Stress-rupture testing at 800 F did not cause any significant changes in the

microstructure of any of the irons containing less than 1.5 per cent molybdenum. A slight amount of decomposition of the bainitic matrices of the higher molybdenum irons was observed. The stress-rupture tests at 1000 F, however, exhibited considerable decomposition of the bainitic matrix in those irons containing bainite.

The pearlitic matrix of the unalloyed and lower alloy irons, however, was not changed significantly by the 100 hr stress-rupture testing at 1000 F. The unaltered pearlitic matrix of the iron containing residual molybdenum and chromium, and the iron with residual molybdenum and 0.64 per cent chromium, are shown in Figs. 14a and b. The agglomeration and spheroidization of the carbides in the bainitic matrices of the irons with 2.02 per cent molybdenum, 0.16 per cent chromium and 1.95 per cent molybdenum, 0.64 per cent chromium after stressing at 100 hr at 1000 F, are apparent in Figs. 14c and d.

The stress-rupture tests to 100 hr at 1200 F indicated considerable decomposition of both the baintic and pearlitic matrices of the series of iron without a chromium addition. The spheroidization of the pearlite is shown in Fig. 15a for the iron with 0.05 per cent molybdenum, 0.16 per cent chromium, and in Fig. 15b for the iron with 1.00 per cent molybdenum, 0.16 per cent chromium. The decomposition of the bainitic matrix for the iron containing 2.02 per cent molybdenum, 0.16 per cent chromium, is readily apparent in Fig. 15c.

The pearlitic matrices of the irons containing chromium were relatively stable during stress-rupture testing 100 hr at 1200 F. Figures 15d and 15e show the stability of microstructure of the irons containing 0.12 per cent molybdenum, 0.64 per cent chromium and 0.96 per cent molybdenum, 0.68 per cent chromium. The higher molybdenum compositions of the chromium irons with bainitic matrices were decomposed to some extent, however, by this 1200 F stress-rupture test for 100 hr, as exhibited by the photomicrograph of the 1.95 per cent molybdenum, 0.64 per cent chromium iron shown in Fig. 15f.

Many features of the stress-rupture results can be explained directly from the influence of the temperature on the microstructures. The more rapid loss of stress-rupture strength with increasing time of testing at the higher temperatures was the result of the decomposition of the matrix. The greater stability of the matrix at 800 F compared with 1000 F is shown by the more rapid increase in initial stress to fracture with increasing molybdenum content at this lower temperature. This effect can be observed for both series of irons in Figs. 10 and 11. The marked decomposition of some of the irons at 1200 F explains the slight improvement in stress at 100 hr testing with increasing molybdenum content, also shown in these figures. The greater resistance of the chromium irons to this decomposition accounted for the superior stress-rupture properties of these irons compared with the straight molybdenum irons at 1200 F.

Comparison with Previous Unreported Data

The information on the effect of molybdenum on the stress-rupture properties obtained previously[9,17] but not published heretofore can be compared with the data in this investigation. This prior work consisted of stress-rupture curves up to 100 hr duration conducted at 800, 1000 and 1200 F on a chromium-containing base iron to which additions of 0, 0.33, 0.60 and 0.81 per cent molybdenum were made. The base iron contained 3.12 per cent carbon, 1.72 per cent silicon and 0.52 per cent chromium. This composition is only slightly lower in carbon equivalent and chromium content than that obtained in this investigation. The irons were cupola melted and sand cast in 2 in. diameter bars.

The results of this previous stress-rupture investigation are reported in Fig. 16 for the three test temperatures. When these data are compared with the stress-rupture curves for the chromium-containing iron obtained in this present work, it is noted that the effect of molybdenum is similar in both series of irons. The stress-rupture curves exhibit the same slope for both investigations at the three temperatures.

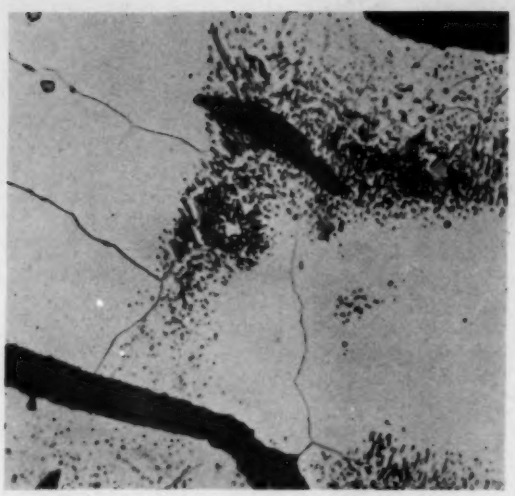
For comparable molybdenum contents, the previous data in references[9,17] are just slightly lower in strength at each fracture time. The effect of molybdenum in increasing the strength at the three temperatures is similar for both types of iron. The stress to cause fracture after 100 hr is about 5000 psi lower at 800 F, 3500 psi lower at 1000 F and only 1000 psi lower at 1200 F for the irons employed in these references.[9,17] The slightly lower results obtained in the earlier work may be attributed to the larger cast section size (2 in. vs. 1.2 in. diameter).

CONCLUSIONS

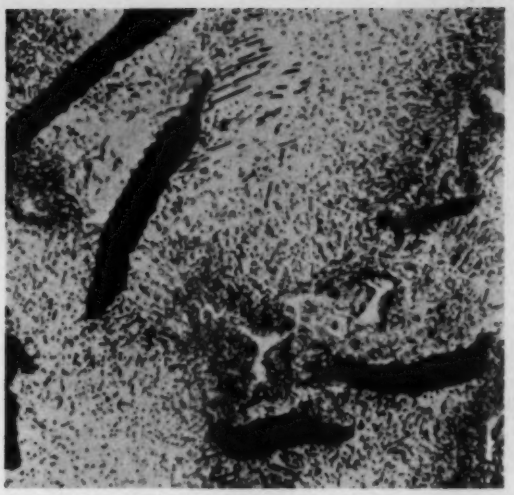
Molybdenum additions up to approximately 1.5 per cent produce a considerable and consistent improvement in the short-time tensile strength at room temperature, 800, 1000 and 1200 F of a gray iron with a 3.8 per cent carbon equivalent. The strength is only slightly lower at 800 F than at room temperature. Increasing the temperature of testing to 1000 and then to 1200 F results in considerably lower strength levels.

The stress-rupture properties of this gray iron to a fracture time of 100 hr at 800, 1000 and 1200 F are appreciably improved by molybdenum additions up to 2.0 per cent. The incremental increase in strength for each increment of molybdenum decreases as the test temperature is raised. This reduction in the effectiveness of molybdenum additions at higher test temperatures is associated with decomposition of the matrix.

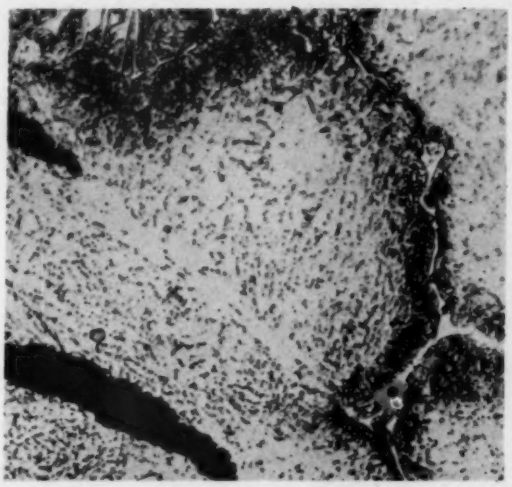
When similar amounts of molybdenum are added to this 3.8 per cent carbon equivalent iron containing approximately 0.60 per cent chromium, the room and elevated temperature properties are further improved for some conditions, but reduced in other cases. The stabilizing influence of chromium on the matrix improves the stress-rupture strength of these irons for all molybdenum contents at 1200 F. At 800 and 1000 F, superior short-time and stress-rupture strengths are obtained with this chromium addition for molybdenum contents up to 1.4 per cent. At higher molybdenum additions, however, the straight molybdenum irons exhibit higher strengths because of the embrittling influence of the massive cementite produced by chromium.



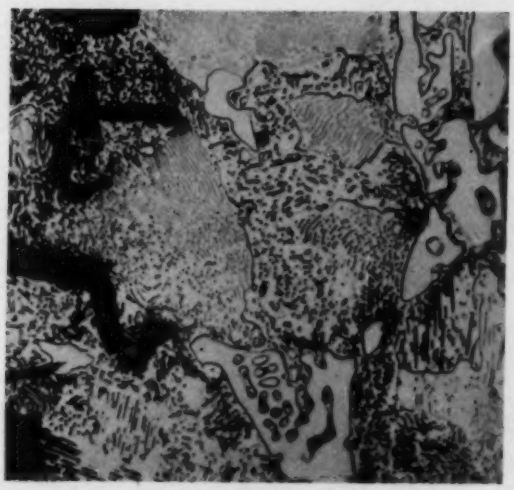
a) 0.05 per cent Mo, 0.16 per cent Cr



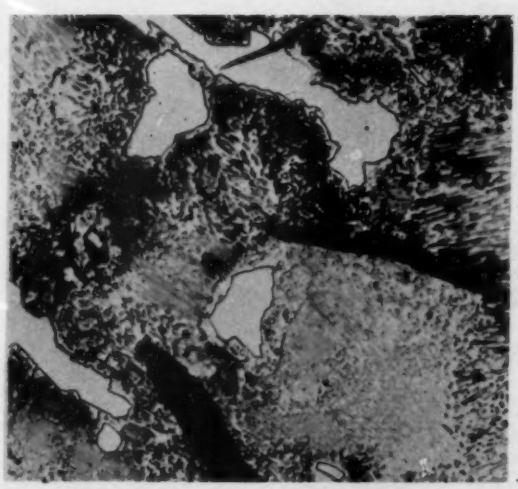
b) 1.00 per cent Mo, 0.16 per cent Cr



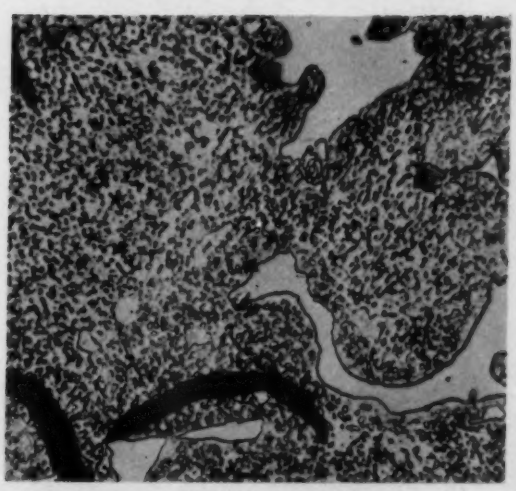
c) 2.02 per cent Mo, 0.16 per cent Cr



d) 0.12 per cent Mo, 0.64 per cent Cr



e) 0.96 per cent Mo, 0.68 per cent Cr



f) 0.95 per cent Mo, 0.64 per cent Cr

Fig. 15 — Microstructures of gray irons of various molybdenum and chromium contents after 100 hr under stress at 1200 F. Picral etch. 1000 ×.

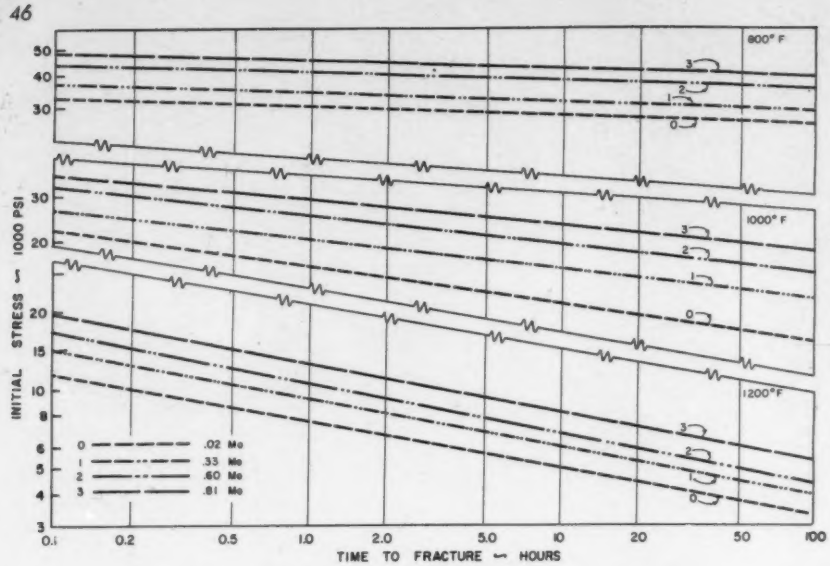


Fig. 16 — Time to fracture at 800 F, 1000 F and 1200 F as a function of the initial applied stress for gray iron with 3.12 per cent carbon, 1.72 per cent silicon, 0.52 per cent chromium and different molybdenum contents. ***(Courtesy Arthur G. McKee Co.)***

The stress-rupture strength decreases uniformly with increasing time of testing to 100 hr for all irons. A more rapid loss in strength with increasing time is evident at the higher testing temperatures. Some of this loss of strength, particularly at the higher temperatures, is attributed to changes in the microstructure. Microscopic examination revealed a slight decomposition of the bainitic matrix at 800 F, further decomposition at 1000 F, and considerable spheroidization of this bainite at 1200 F. The pearlitic matrices showed evidence of decomposition only at 1200 F.

ACKNOWLEDGMENT

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REFERENCES

1. C. O. Burgess and T. E. Barlow, "Increased Use of Gray Cast Iron in High-Temperature Operations", AFS TRANSACTIONS, *vol. 63*, (1954), pp. 283-291.
2. *Gray Iron Castings Handbook*, Gray Iron Founders' Society, 1958.
3. W. L. Collins and J. O. Smith, "Fatigue and Static Load Tests of a High Strength Cast Iron at Elevated Temperatures", A.S.T.M. *Proceedings, vol. 41*, (1941), pp. 797-810.
4. R. S. MacPherran and R. Krueger, "Effect on Cast Iron of Prolonged Heating at 800-1100 F", AFA TRANSACTIONS, *vol. 38*, (1930), pp. 826-864.
5. J. W. Grant, "A Few Short-Time Growth and Creep Tests on an Unalloyed Pearlitic Grey Iron", *Journal of Research*, British Cast Iron Research Association, *vol. 3*, June, (1950), pp. 441-445.
6. C. R. Tottle, "Deformation Characteristics of Five Gray Cast Irons at 400 and 500 C", *Foundry Trade Journal, vol. 89*, (1950), pp. 273-281.
7. J. R. Kattus, "Properties of Cast Iron at Elevated Temperatures, 1000 F Creep-Rupture Properties", Progress Report to Joint A.S.M.E.-A.S.T.M. Committee on Effect of Temperature on the Properties of Metals, Southern Research Institute, February 1, 1955.
8. C. H. Lorig and F. B. Dahle, "Mechanical and Creep Properties of Molybdenum Cast Iron", *Metals and Alloys, vol. 2*, (October 1931), pp. 229-235.
9. L. J. Klingler and L. J. Ebert, "The Stress Rupture Characteristics of Several Alloy Cast Irons", Report to A. G. McKee Co., December 20, 1951.
10. V. A. Crosby, "Microstructure and Physical Properties of Alloy Cast Irons", AFS TRANSACTIONS, *vol. 45*, (1937), pp. 626-650.
11. W. G. Scholz, D. V. Doane, and G. A. Timmons, "Effects of Molybdenum on Stability and High Temperature Properties of Pearlitic Malleable Iron", AFS TRANSACTIONS, *vol. 63*, (1955), pp. 251-260.
12. T. D. Parker, "Molybdenum: Versatile Alloy for Cast Iron", *Pig Iron Rough Notes, no. 118*, (Autumn, 1956), pp. 2-9.
13. C. O. Burgess, "Influence of Chromium on Oxidation Resistance of Cast Iron", A.S.T.M. *Proceedings, vol. 39*, (1939), pp. 604-625.
14. H. J. Tapsell, M. L. Becker, and C. B. Conway, "The Behavior of Five Cast Irons in Relation to Creep and Growth at Elevated Temperatures", *Journal of the Iron and Steel Institute, vol. 133*, (1936), pp. 303-329.
15. R. R. Kennedy and G. J. Oswald, "The Effect of Various Alloys on the Growth of Gray Iron Under Repeated Heatings", AFS TRANSACTIONS, *vol. 34*, (1926), pp. 871-880.
16. H. Timmerbeil, G. Clas, and O. Mattern, "The Heat Resistance of Cast Iron and the Effect of Different Alloying Elements", *Giesserei, vol. 38*, (September 20, 1951), pp. 476-482; (October 4, 1951), pp. 523-526, (in German).
17. D. J. Garibotti and L. J. Ebert, "Evaluation of the Effect of Molybdenum Additions of the Stress-Rupture Characteristics of Cast Iron", Report to A. G. McKee Co., March 1953.
18. F. B. Rote, and W. P. Wood, "Segregation of Molybdenum in Phosphorus-Bearing Alloyed Gray Cast Iron", *A.S.M. Transactions, vol. 35*, (1945), pp. 402-425.
19. J. C. Hamaker, Jr., W. P. Wood, and F. B. Rote, "Internal Porosity in Gray Iron Castings", AFS TRANSACTIONS, *vol. 60*, (1952), pp. 401-427.
20. R. S. Archer, J. Z. Briggs, and C. M. Loeb, *Molybdenum*, Climax Molybdenum Co. (1948).

SAND MOVEMENT AND COMPACTION IN GREEN SAND MOLDING

By T. J. Bosworth, R. W. Heine, J. J. Parker, E. H. King and J. S. Schumacher

ABSTRACT

A description of the principles controlling sand movement during squeezing, jolting and contour squeezing, and combinations of these, is given. The compaction process, and development of mechanical properties with increasing density, is also described. Rigid frame design rules for contour squeezing are given, with methods of using the contour frame to achieve molds of high and uniform density and hardness.

INTRODUCTION

When molding is done with green sand, the sand is moved and compacted simultaneously. Forces applied for molding cause the sand to move down into the flask. As it moves, it is compacted, i.e., the bulk density is increased. When the sand has been compacted, and is supported by the flask and pattern, the molding forces are balanced and sand movement ceases. Thus, a study of sand movement in green sand molding must be concerned with compaction and molding forces as well as sand movement.

BULK DENSITY

To describe the limits of compaction or bulk density consider the data in the table. The table shows that the bulk density of a bentonite bonded sand may change from as low as 50 lb/cu ft to as high as 105 lb/cu ft from compaction during the molding process. In a fire-clay bonded sand, the density may vary from about 60 to as high as 115 lb/cu ft during compaction. The higher density level of the latter sand is due to the packing of fire clay in the voids of the sand grain mass.

Even when compacted, the molding sand mass is composed of approximately 60 per cent solids and 40 per cent voids (table).[1,2,3] Regardless of type of bonding clay in the sand, there is a progressive increase in sand bulk density as the amount of work done in compacting the sand increases. The limiting bulk density is achieved when compaction has occurred to the point of sand grain to sand grain contact throughout the mass. This will occur at a density of about 100-115 lb/cu ft. Effort at further compaction then causes sand grain fracture. Application of higher forces will not cause more compaction once the maximum bulk density has been reached.

As compaction of the sand occurs, the mechanical properties of the sand change in the manner shown in Fig. 1. The sand mixture represented in Fig. 1 is composed of 6 per cent western bentonite, 2 per cent D-4 seacoal, 1 per cent carbonized cellulose, balance 85 AFS-4 screen sand. The mixture normally was used at 3.2-3.4 per cent H_2O, although Fig. 1 includes higher moisture contents. As the sand density increases, its green compressive, green shear strength and mold hardness increases. This means that to raise the bulk density from one level to a higher level, sufficient force must be applied to overcome the strength of the sand and cause the increased

TABLE – DENSITIES OF SOME TYPICAL MOLDING SANDS

Material	Bulk Density, lb/cu ft	% Solids*	% Voids**
1. Silica sand (clay free)	100-115	60.4-69.5	39.6-30.5
2. Dry sands plus southern or western bentonite and carbonaceous additives having 9-11% AFS clay, 4-8% total combustible, 60-75 AFS fineness	90-105	54.4-63.5	45.6-36.5
3. Molding sand in No. 2 tempered with water to molding consistency			
a. Riddled	50- 65	30.2-39.2	69.8-60.8
b. Compacted to 70-85 mold hardness	65- 85	39.2-51.3	60.8-48.7
c. Compacted to 90-95 mold hardness	90-105	54.4-63.5	45.6-36.5
4. Fire-clay bonded molding sand; 12-15% AFS clay, 4-8% total combustible, 60-75 fineness; tempered with water			
a. Riddled	60- 75	36.2-45.3	63.8-54.7
b. Compacted to 60-85 mold hardness	75- 95	45.3-57.3	54.7-42.7
c. Compacted to 90-95 mold hardness	105-115	63.5-69.5	36.5-30.5

*True density of solid silica = 165.4 lb/cu ft

$$\text{Per cent solids} = \frac{\text{Bulk Density}}{\text{True Density}} \times 100;$$

$$\text{Example: } \frac{100}{165.4} \times 100 = 60.4$$

**% Voids = 100% − % Solids

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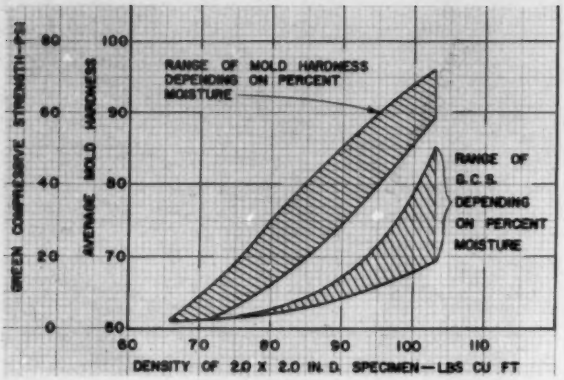


Fig. 1 — Graph showing how mechanical properties change as compaction of sand occurs, for sand mixture: 6 per cent western bentonite, 2 per cent D-4 seacoal, 1 per cent carbonized cellulose, remainder 85 AFS-4 screen sand.

density. Thus, in green sand molding there are interacting processes at work, as:

1. gross movement of the sand must occur as sand flows above and adjacent to the pattern and is compacted.
2. fitting together or packing of sand grains and expression of the gross voids as the maximum bulk density is approached by compaction.
3. increase of mechanical strength of the sand aggregate as gross movement and packing of the sand occurs.

SAND MOVEMENT

Gross movement of sand during green sand molding was studied by means of a section of a mold, flask and pattern. The section (Fig. 2) consists of a rectangular slab 12 in. high, 9 in. within the flask sides and 1 in. thick. The slab was closed on each face by a metal plate. Distance between pattern and flask is 3 in. on each side. The pattern was made to be either 4 in. high or 6 in. high as desired. Sand was put into the section in a grid pattern of units 2 in. x 1 in. x 1 in. (Fig. 2), in order to observe the movement caused by the molding action. The molding sand mixture was composed of 6 per cent western bentonite, 1 per cent carbonized cellulose, balance 4 screen – 85 AFS sand used at a moisture content of 3.20-3.60 per cent. Seacoal 2 per cent of D-4 grind, was added to a part of the sand to obtain a black color for contrast in the grid of Fig. 2.

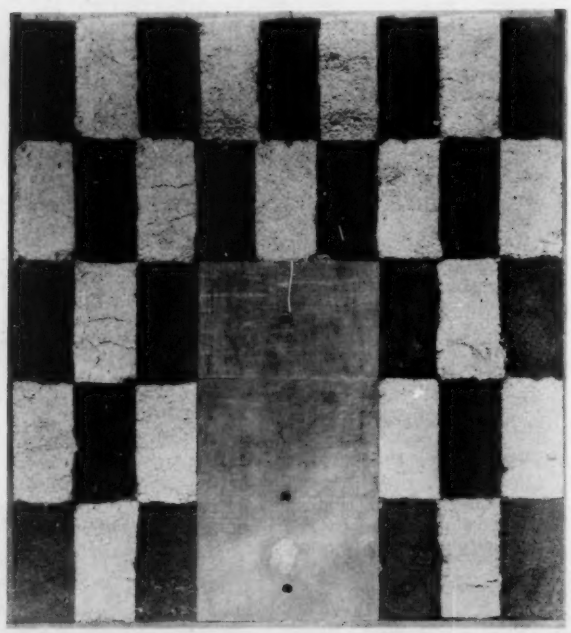


Fig. 2 — Mold, flask and pattern section used to study sand movement. Scale of grid units is 1 in. wide x 2 in. high.

The initial bulk density of the sand was controlled at 50 lb/cu ft by weighing the sand put into the grid. After filling in the grid, the face of the section was covered with a metal plate to confine the sand. Molding effort was then applied in the form of squeezing, jolting or jolting with a follow weight. Sand movement was observed by removing the metal plate and noting deformation of the sand grid. Also, mold hardness measurements were made on the exposed surface and on the surfaces adjacent to the bottom parting, vertical and horizontal walls of the pattern and surfaces in contact with the squeeze plate.

Average final density was also determined by noting the final volume and sand weight. The density at any particular location on the grid can be determined by noting the volume occupied by the original 2 x 1 x 1 in. units. For example, if the height of a unit is decreased from 2.0 in. to 1.0 in., then the density of that unit has doubled, i.e., compacted from 50 to 100 lb/cu ft. Thus, measurement of the grid units shows both sand movement and amount of compaction.

Sand Movement From Squeezing

The gross movement of sand from squeezing alone is illustrated in Fig. 3a. The flask and part of the pattern have been removed in the figure. Mold hardnessess of the grid are shown in Fig. 3b. Figures 3a and 3b reveal a bulged column of sand of maximum mold hardness, 93-98 above the pattern. The height of the grid units above the pattern has been reduced to less than 1 in. from the initial height of 2 in. This means that the density in this location has doubled, i.e., compaction caused bulk density to change from 50 lb/cu ft to about 100 lb/cu ft. Average density of the sand in the mold is only 65 lb/cu ft because of low density near the parting.

At the higher density (100), this sand would have a green compressive strength of more than 50 psi (Fig. 1). Squeezing to this density and strength could be achieved only by applying 50 psi squeeze pressure above the pattern. Note this is pressure above, not adjacent to, the pattern. This is the squeeze pressure required, and it agrees well with earlier work reported.[4] The column of high strength sand over the pattern supports the major portion of the applied squeezing load and transmits it to the pattern.

Figures 3a and 3b also show that a compacted

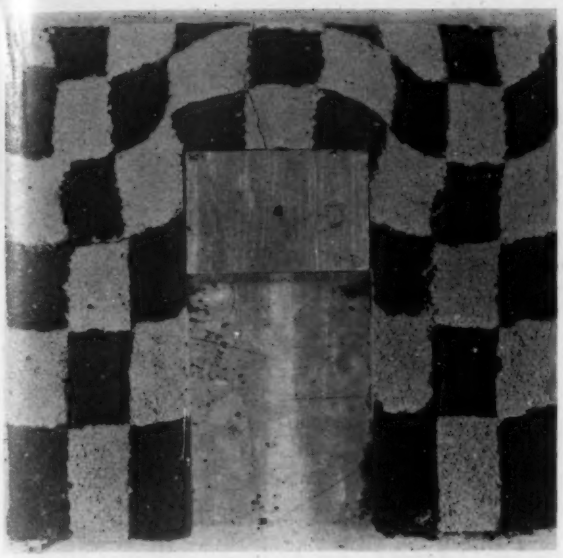


Fig. 3a — Deformation of sand grid caused by squeezing with a flat squeezing plate. Pattern height is 6 in., initial sand height was 12 in., and final sand height is 8 in.

MOLD HARDNESS SURVEY

TOP		91	92	97	98	98	95	93	
	81	86	86	94	94	95	89	89	85
	82	78	87	95	93	98	90	82	87
	68	68	53				63	73	70
	60	61	51				57	60	56
	48	46	45				54	55	60
	36	33	32				48	52	48
	30	25	28				34	40	37
	20	24	8				25	40	29

TOP OF PATTERN

PARTING				89	92	95			

Specimen number 53

Flat squeeze X	Contour squeeze		Jolt
Pattern 3 in. x 6 in.			
Initial sand height 12 in.		Density, lb/cu ft	50
Final sand height 8 in.		Density, lb/cu ft	65
Remarks: Soft spot at pattern corners. Density nonuniform			

Fig. 3b — Hardness survey of mold section shown in Fig. 3a.

layer of sand builds up in advance of the moving squeeze plate. This compacted layer sends out a compaction wave of partially compacted sand in advance of its own movement. This wave extends downward alongside the pattern, and shows gradually diminishing compaction at greater distances from the parting. The compaction wave does not extend to the pattern plate in this case, because of the resistance to further movement of the squeeze plate offered by the column of sand over the pattern. Consequently, little compaction occurs in the grid units adjacent to the parting, and mold hardness therefore remains low (Fig. 3b).

Rupture of continuity of the sand at pattern corners also may be noted in Fig. 3a. This is the result of bulging of the sand column over the pattern, and attendant side-wise thrust on the sand pushing it away from the corner. Figure 3a also reveals a frictional effect or drag between sand and vertical pattern walls as the sand between pattern and flask moves downward.

Finally, Fig. 3a demonstrates that sand does not flow around corners from above a pattern downward around its sides when squeeze loading is applied. The sand moves by compaction in horizontal layers and by vertical shear. The combination of horizontal compaction and vertical shear produces the compaction waves, illustrated in Fig. 3a.

Sand Movement in Jolting

The gross movement of sand from jolting alone is illustrated in Fig. 4a and 4b. The pattern and flask have been removed in Fig. 4a. Twenty jolts of 3 in. length were applied to obtain the results shown. Figure 4a shows again the build-up of a column of sand over the pattern. However, this sand column does not reach the high density achieved with squeezing. In this case, density is maximum at the parting and just above the pattern, as shown by hardness measurements and by the grid deformation. A shear plane separates the column of sand over the pattern from the column that moves downward between the pattern and flask walls.

A mold hardness of 82-88 at the parting corresponds to a density of 73-80 lb/cu ft. Average density for this section is 65 lb/cu ft. The average is lower than that at the parting, because density decreases rapidly away from the parting. This is indicated by the grid units at the top of the mold maintaining almost the 2 in. height which they had initially. Again, in jolting, it is seen that virtually no sand moves from above the pattern downward along the sides, i.e., sand does not flow around corners. Discontinuities in compaction of the sand at corners develops because of this lack of fluid flow behavior.

Compaction by jolting occurs at the immovable horizontal surfaces of the pattern and parting. The momentum imparted to the sand during jolting is dissipated in compaction when the sand strikes the immovable surface. Thus, a horizontal compaction wave is set up which produces maximum density at the horizontal surfaces, and decreasing density at increasing distance from the horizontal surfaces. Increasing the number of jolts utilized simply increases the thickness of the compaction wave above the horizontal surfaces. To cause the compaction wave to progress upward along the vertical pattern wall to the top of the pattern would require an impractically large number of jolts.[4,5] Increasing moisture content of the sand greatly increases the number of

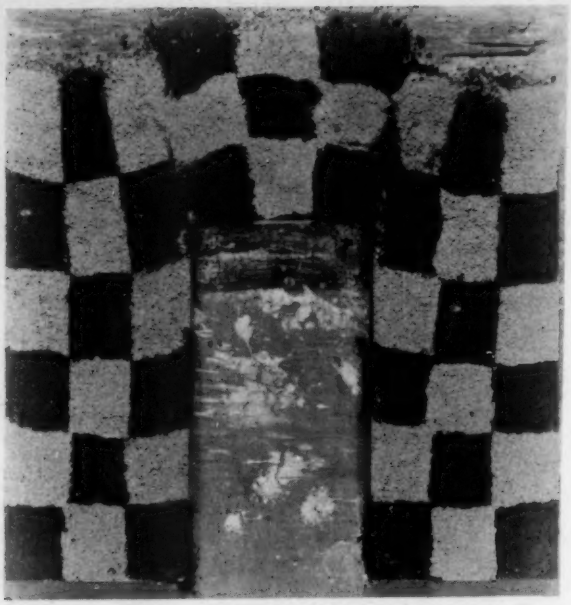


Fig. 4a — Deformation of sand grid caused by jolting the mold section 20 times with a 3.0 in. jolt. Pattern height is 6 in., initial height of sand was 12 in., and final sand height is about 9 in. Pattern and flask have been removed from mold.

MOLD HARDNESS SURVEY

TOP									
				26	38	35			
	32	33	34	57	58	54	35	38	40
	40	45	37	65	68	64	40	36	40
	47	48	46				41	47	47
	51	52	51	←67			47	53	52
	58	59	58	←72			50	54	55
	64	60	58	←81		78→	53	56	61
	65	65	63	←81		75→	61	61	65
	62	63	63	←82		78→	66	64	65
				TOP OF PATTERN					
PARTING	85	88	84	86	82	83	83	82	80

Specimen number 61

Flat squeeze	Contour squeeze		Jolt 20-3.0 in.
Pattern 3 in. x 6 in.			
Initial sand height 12 in.		Density, lb/cu ft 50	
Final sand height 8½-9½ in.		Density, lb/cu ft 65	
Remarks: Shear planes emanating from corners of pattern.			
Density nonuniform			

Fig. 4b — Hardness survey of mold section in Fig. 4a.

jolts required to cause the compaction wave to move up along the pattern.[4,5]

Sand Movement in Squeezing and Jolting

Compaction waves at the top of the mold from squeezing and compaction waves at the parting from jolting could, theoretically, be caused to meet and produce uniform high density molds by combining squeezing and jolting. Figures 5a and 5b show sand movement from jolting, leveling the sand and then squeezing. Maximum compaction occurs in a column of sand over the pattern. Compaction at the parting is caused practically by jolting alone. Minimum compaction and mold hardness occur about half way between the top and bottom, because the two compaction waves have not been brought together. In fact, it is not possible to bring the squeeze compaction wave down to meet the jolt compaction wave, because the sand column over the pattern prevents further downward movement of the flat squeeze plate. Thus, a low density zone will always exist between the two high density zones by this method of molding, unless the flask is shallow; for example, 4 in. deep or less. Because of the low density middle zone, average density was 85 lb/cu ft, although 100 lb/cu ft was approached in the densest layers.

Sand Movement in Contour Squeezing

The limiting effects of columns of sand over the pattern can theoretically be offset by the use of a contoured squeeze plate, rather than by a flat squeeze plate, as suggested in Fig. 6.[4] The design of the squeeze plate is predicted[4] by the formula $U = P/H \times M$.

U = height of the upset or contour block added to the squeeze plate.
P = pattern height.
M = total movement after contact of the contour plate against the sand surface.

The formula is set up graphically in Fig. 6.

As an example of its use, consider the mold section of this study with a pattern height of 6 in., an initial sand height (H) of 12 in., and a movement (M) of 8 in. From the graph in Fig. 6, for $P/H = 0.50$, $M = 8$, $U = 4$ in. The use of such a contoured squeeze plate is shown in Fig. 7a and 7b. In the case shown M was experimentally found to be 8.25 in. where U was 5 in. This design of contour squeeze plate accomplished compaction of sand down to the parting line by squeezing alone. The compaction wave advancing ahead of the contour block has penetrated to the parting to produce high mold hardness both at the parting and vertical side walls. Unfortunately the thin layer of sand between the contour block and pattern is impractical.

In addition, wedging of sand between flask and pattern locks in stresses having a horizontal component that bulges the flask walls. The development of such stresses by diversion of vertically applied load toward horizontal force components is well illustrated in Fig. 8. The wedging effect is analogous to attempting to stuff a large ball into a small hole.[4] This problem is encountered in all uses of contour blocks, air or hydraulic diaphrams, and also of course in using the conventional flat squeeze board.

The problem is one of designing a squeezing method, contour or otherwise, which will prevent

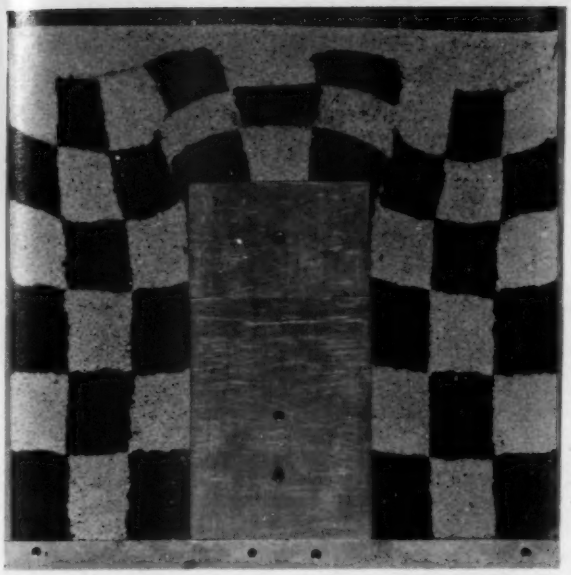


Fig. 5a — Deformation of sand grid caused by jolting 20 times with 3 in. jolt stroke, filling with sand to level and then squeezing with a flat plate.

MOLD HARDNESS SURVEY

TOP	94	95	96	93	94	93	94	95	92
	92	92	91	95	92	94	88	90	94
	91	92	92	95	97	94	92	93	91
	91	92	90	95	93	95	90	91	90
	82	82	81	←77		76→	85	84	86
	77	77	72	←71		73→	75	81	77
	71	71	66	←66		68→	70	71	71
	60	64	60	←78		75→	57	60	63
	63	66	60	←78		77→	63	67	64
	67	59	67	←82		81→	66	67	67
				TOP OF PATTERN					
PARTING	82	85	84	93	89	93	83	82	84

Specimen number 63

Flat squeeze V	Contour squeeze		Jolt 20-3.0 in.
Pattern 3 in. x 6 in.			
Initial sand height 12 in.		Density, lb/cu ft 50	
Final sand height 8½ in.		Density, lb/cu ft 60-95; 85 ave.	
Remarks: Sand was added to level the surface after jolting, compare with no. 61. Density nonuniform			

Fig. 5b — Hardness survey of mold section shown in Fig. 5a.

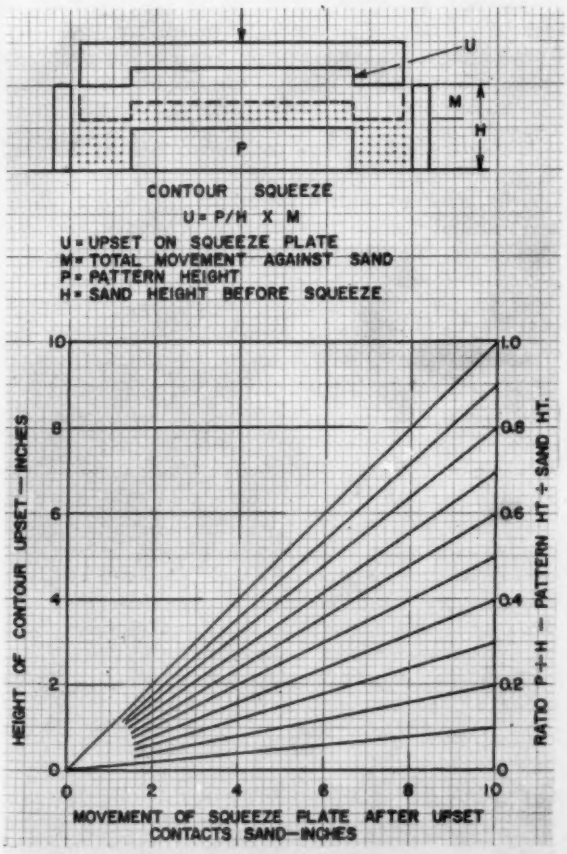


Fig. 6 — Theoretical design of upset on a contour squeeze plate, and graphical means of determining the relation of height of upset to total movement of squeeze plate against the sand.

the building up of columns of sand over the pattern, and also prevent the diversion of forces or wedging effects between pattern and flask.

Many different designs of contour squeezing frames were studied to solve this problem. Finally, contour squeezing was found to be most efficiently performed by means of a contour squeeze frame of the shape shown in Figs. 9a and 9b. Deformation of the grid (Fig. 9a) shows the sand movements caused by this method of contour squeezing. Figure 9b shows the mold hardnesses produced. They are notably high and uniform running 93-96 at the pattern surface. A final density of 95 lb/cu ft or more was achieved. Thus, this method of contour squeezing produces the ultimate in uniformity of hardness and high density.

Sand movement from squeezing by the method shown in Fig. 9 reveals several important phenomenon:

1) Maximum movement and compaction occurs immediately ahead of the tip of the contour block, and extends for a distance of 1½ in. to 2 in. from the tip.
2) When the leading edge of the fully compacted wave ahead of the tip strikes the parting, further movement of the contour frame will be prevented by the sand column between the tip and parting.
3) Horizontal movement of sand toward the pattern is caused by shearing off of compacted sand under the block, and shear movement around the curved tip toward the pattern. Horizontal movement is shown in Fig. 10.

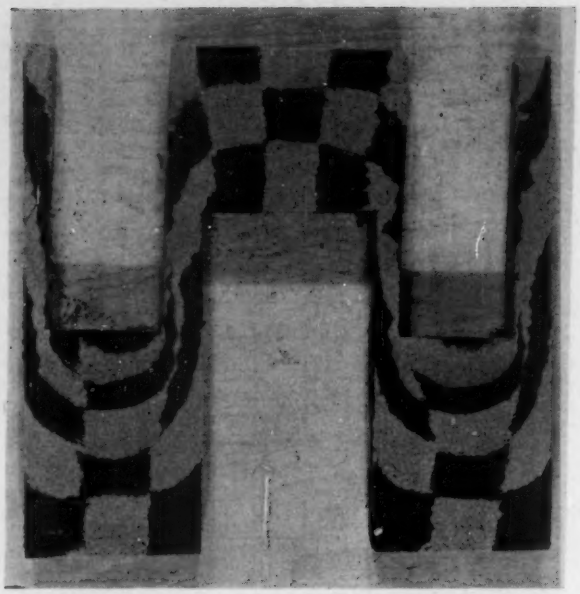


Fig. 7a — Example of use of contour blocks for compacting sand alongside the pattern, and obtaining the hardness down to the parting. Squeezed only. Pattern and flask have been removed.

4) A tear-drop shape compaction wave emanates from the tip, because of the shearing off process, described in 3 above.
5) If a sand column builds up between the contour block tip and the parting before the sand over the pattern is compacted, then the sand over the pattern will not reach maximum hardness and density. The reverse of the situation is also true. This is discussed further in a following section.

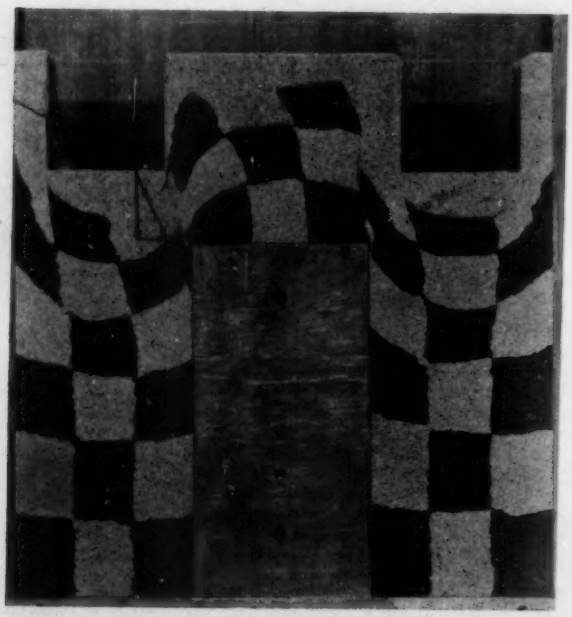


Fig. 8 — Molding section illustrating diversion of vertical squeezing force to horizontal force components. Wedging of the sand causes bulging of flask walls and locked in stresses that make patterns difficult to draw and cause mold cracks.

MOLD HARDNESS SURVEY

TOP	99	99	90	92	94	93	93	99	100
	91			90	92	92	94	94			57
	88			90	90	93	92	92			57
	77			89	92	90	93	91			65
	72			79				81			70
	70			76				76			73
	83	95	91	←97		98→	89	93	92	
	89	93	93	←98		96→	90	92	93	
	87	91	89	←97		95→	89	90	87	
	83	86	87	←97		96→	83	86	83	
					TOP OF PATTERN				
PARTING	96	94	92	98	94	97	84	90	81

Specimen number 57

Flat squeeze	3 in. Jolt	Contour squeeze	V
Pattern 3 in. x 6 in.			
Initial sand height 12 in.		Density, lb/cu ft 50	
Final sand height 8¾ in.		Density, lb/cu ft 80-95	
Remarks: Contour squeeze plate, 2 in. x 5 in. blocks. Density nonuniform, soft spots between contour block and pattern.			

Fig. 7b — Hardness survey of mold section shown in Fig. 7a.

6) Wedging or side-wise thrust against the flask can be eliminated if the contour block is properly designed.

Using this technique of contour squeezing, the goal of achieving high and uniform hardness and density can be achieved with a minimum of molding force and residual stresses. The essential principle of squeezing by this method is utilization of the compaction wave ahead of the contour block, to move high density sand vertically down to the parting and horizontally to the pattern surface. Proper use of the compaction wave eliminates wedging and residual stresses.

Many modifications of pattern and contour frames of this type were studied to further observe sand movement. An increase in pattern height may require addition of a step on the pattern contour block to compensate for the tear-drop compaction movement of the sand (Fig. 10). Figure 10 also shows how horizontal movement of the sand prevents soft, low density areas from developing at corners of the pattern. If the initial density of packing sand into the flask is raised from 50 lb/cu ft to 65-75 lb/cu ft, then the step shown on the contour block in Fig. 10 may not be needed.

Pattern Shape. The effect of inclined and rounded pattern surfaces and pockets was also studied, as shown in Figs. 11, 12, and 13. In each case, the contour squeeze frame produced uniformly high density and hardness, as in the case of Fig. 9. Horizontal sand movement serves to hold the sand against sloping or rounded pattern surfaces, as demonstrated by grid deformation in Figs. 11 and 12. In the case of the

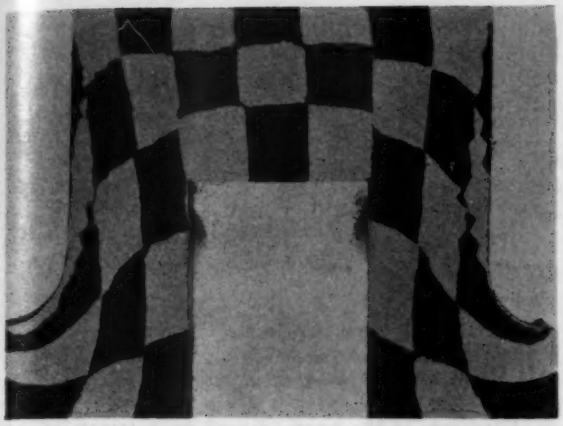


Fig. 9a — Mold section squeezed with specially designed contour frame. Note the tear-drop shape of compaction wave which extends down to the parting, and also causes horizontal movement of sand against the pattern walls.

MOLD HARDNESS SURVEY

TOP		95	98	97	98	97	98	95	
		94	97	96	98	95	97	94	
		96	94	97	95	96	94	96	
		95	94	94	95	94	93	95	
		94	94				93	92	
		98	95				92	92	
	99	96	95				95	98	98
	99	96	95				96	97	99
PARTING									

TOP OF PATTERN

Specimen number 83

Flat squeeze	Contour squeeze V	Jolt
Pattern 3 in. x 4 in.		
Initial sand height 10 in.	Density, lb/cu ft 50	
Final sand height 6¾ in.	Density, lb/cu ft 95	
Remarks: Contour block is 5 in. long x 1 in. wide with 1 in. radius on end. Density high and uniform.		

Fig. 9b — Mold hardness survey and contour frame shape of mold section shown in Fig. 9a.

pocket (Fig. 13) a separate probe must be added to the contour frame to send a compaction wave down into the pocket.

Molding Sand Amount in Flask. One important fact discovered in using the various designs of contour frames concerns the use of the proper amount of molding sand in the flask. When too little sand was used, maximum compaction occurred between the contour block and the parting, but not between the top of the pattern and the squeeze plate. Conversely, when too much sand was used, maximum compaction occurred between the top of the pattern and the squeeze plate, but not between the contour block and the parting. In other words, uniform high density and mold hardness was developed only when the proper amount of sand was used in the flask.

The requirement of using the proper amount of sand for squeezing by this method has a fundamental basis. After squeezing, the volume between the contour frame and pattern is fixed. If the volume is fixed, the weight of fully compacted sand in this volume is also fixed. Therefore, only one weight of sand is the correct one to occupy this volume. Ex-

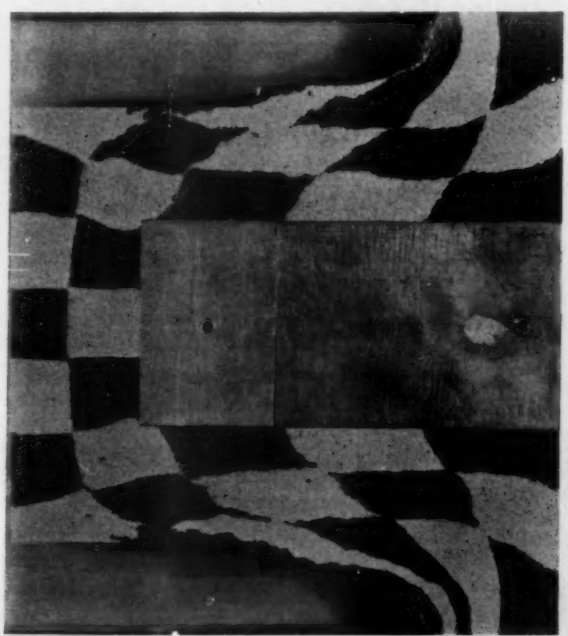


Fig. 10 — Mold section with 6 in. high pattern, and contour plate with step added to avoid low density (and hardness) area at upper corner of pattern.

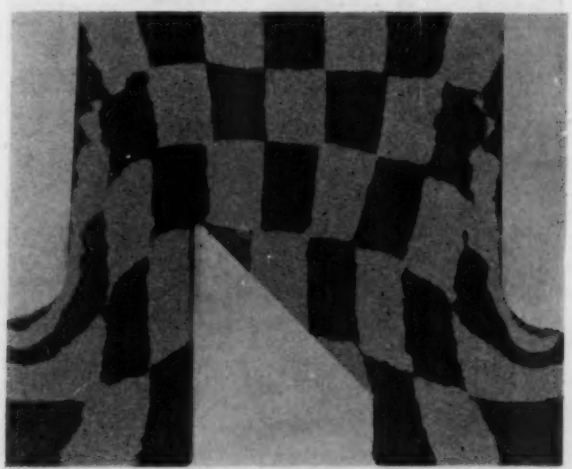


Fig. 11 — Sand movement pattern developed by squeezing against an unlined surface with the contour frame.

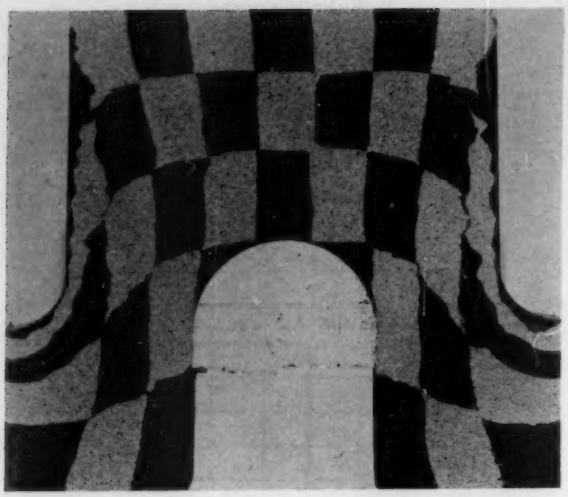


Fig. 12 — Sand movement pattern developed by squeezing against a curved surface with the contour frame.

perimentally, it was found, both in the mold section and in actual molding, that by weighing the amount of sand going into the mold the proper volume after squeezing was assured. Actually, this requires sand on the basis of 100 lb sand to occupy 1 cu ft, in the case of bentonite bonded sand.

Design Rules for Contour Squeezing

Further studies of squeezing molds with the contour frame resulted in a set of rules for designing the squeezing frame. They are:

1) The probe (contour block) height should be equal to 1 to 1.20 times the pattern height.

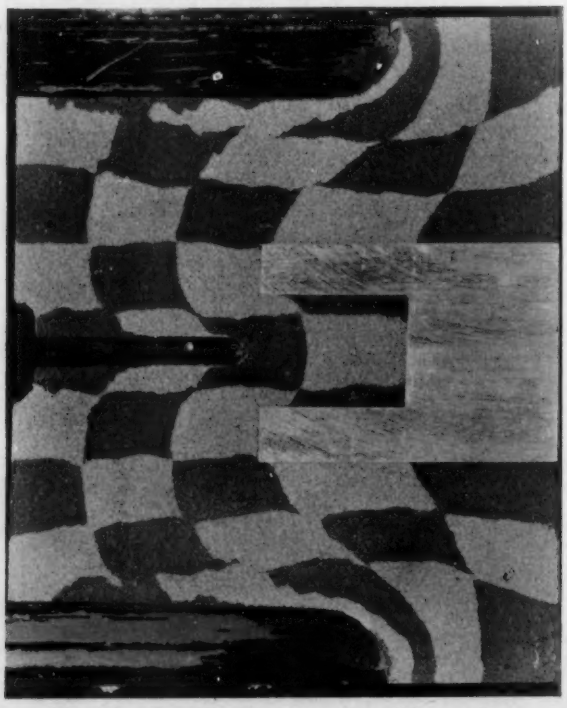


Fig. 13 — Squeezing into a pocket with a probe.

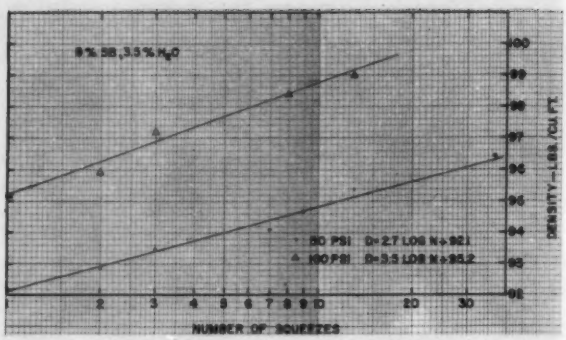


Fig. 14 — Effect of number of squeezes on density of a sand using two different squeeze pressures.

2) Thickness of the probe should be ¼ to ⅓ of the distance between the pattern and the flask.
3) The tip of the probe is a quarter circle of radius equal to probe thickness. It may also consist of 4 inclined planes of 22.5 degrees.
4) The probe sides should be vertical or have not more than one degree taper.
5) A step addition may be made to the inside of the probe, if the pattern is more than 4 in. high. The step addition is ¼ to ⅓ the width of the probe, and 2½ to 3 in. from the tip of the probe. This step addition is not needed unless soft spots are encountered on vertical walls, or initial density of sand packed in the flask is low.
6) The amount of sand put into the flask must be controlled by weighing. The sand weight should correspond to the volume of sand required between the contour frame and the pattern after compaction to a density of 95-100 lb/cu ft. An alternative to weighing the sand is to have contour blocks which move independently of the flat squeeze plate.
7) Additional probes are required to pack sand into pockets (Fig. 13). These probes are designed to be long enough to reach to within 1½ to 2.0 in. of the bottom of the pocket after squeezing. Further, these probes should have a thickness (diameter) equal to ¼ to ⅓ of the width (diameter) of the pocket.

Using these rules, the authors have designed contour frames for squeezing a number of different pattern and flask combinations in foundries.

Contour Frame and Jolting

The contour frame designed as described above can be used for several different techniques of molding. After the flask has been loaded with the proper weight of sand, conventional jolting may be applied. Jolting can then be followed by squeezing with the contour frame to complete molding.

The contour frame can also be used as a follow-weight in jolting. After filling the flask with the proper weight of sand, the frame is placed in position and jolting performed. When completely jolted, the frame will settle into the same position it would occupy from squeezing. The follow-weight must weigh about 0.75-1.0 lb per sq in. of flask area, to have sufficient ramming power if the frame is used as a

follow-weight. However, the same uniform high density and mold hardness of sand will be obtained as by squeezing.

COMPACTION

The preceding paragraphs have dealt with gross movement of sand during squeezing or jolting. Compaction, or increase in bulk density, occurs as the sand is caused to move by the forces of squeezing or jolting. This occurs in several stages:

a) Expression of gross voids from loosely packed sand.
b) Buildup of horizontal layers of higher density sand under a moving squeeze plate, or on the parting from momentum of the sand in jolting. High density layers of sand are also built up in waves on horizontal and vertical surfaces of contour squeeze frames.
c) Eliminations of low density areas and complete expression of gross voids by the impact of high density layers against each other.
d) Fitting together of agglomerates and particles as maximum bulk density is approached from the effects of items a, b and c, above.

The above changes are naturally overlapping in various parts of the mold. However, as maximum density is approached there is obviously no longer any room in the mass for gross voids or low density areas. In other words, maximum density implies uniform density and uniform packing of the sand. This is a primary advantage of molding sands rammed to high density; they cannot be nonuniform in density or properties.

Another consequence of the above modes of compaction, is the manner in which mechanical properties of the sand change. Refering again to Fig. 1, it is seen that green compressive strength and mold hardness change slowly as density increases from 60 to 80 lb/cu ft. However, as density increases from 80 lb/cu ft upward, these mechanical properties shoot up dramatically, as shown in Fig. 1. This is also true in connection with mold wall movement and consequent dimensional accuracy of castings.[4] Much greater dimensional accuracy is obtained when mold hardness exceeds 92 and density exceeds 95 lb/cu ft,[4] for bentonite bonded sands.

Considering a number of sands, it has been found that the bulk density which is at least 94-95 per cent of the maximum obtainable is desired in order to obtain a practical combination of uniformity of density, high mold strength and hardness and dimensional accuracy of castings. For bentonite bonded sands this will amount to about 90-98 lb/cu ft depending on additives, and for fire-clay bonded sands this will amount to about 108-115 lb/cu ft again depending on additives.

As pointed out in the preceding paragraph, mechanical strength of the sand increases greatly as compaction causes maximum density to be reached (Fig. 1). If simple compression only were acting when a mold is squeezed, Fig. 1 shows that squeeze pressure of 40-50 psi would be adequate to obtain full compaction. However, sand columns build up over the pattern and prevent uniform application of the squeeze pressure (Fig. 2). No mere increase in squeeze pressure will compensate for the resistance offered by such sand columns.

Pressures as high as 40,000 psi have been used experimentally, with fragmentation of the sand grains in the column as the only result. Instead, the squeeze pressure must be properly applied so that the full benefit of applied forces can be obtained. The contour squeeze frame described accomplishes this by utilizing and placing compaction waves in their proper relation, and preventing sand columns from building up until maximum compaction has occurred.

Green shear strength, as well as green compression strength, is involved in sand movement and compaction. Green shear strength of a practical maximum of about 8.0 psi is observed in bentonite bonded sands. Thus, 8 psi of load would be the maximum required to cause shearing of a 1.0 in. sand column for each in. of height. For example, in a 12.0 in. deep flask, 96 psi load would be required to cause the shearing in final compaction. According to experimental measurements using molds and mold sections, molds of 12-20 in. depth can be squeezed with pressure of 60-80 psi ro reach full compaction, if the forces are properly applied as with the contour frame. Practically, a limit of about 150 psi is desirable to accommodate unusual depth and patterns.

Another evidence that moderate squeeze pressure can achieve high density is offered in the results of repetitive squeezing. Repetitive squeezing in a low squeeze pressure can achieve the same density achieved by a single squeeze at higher pressure, as illustrated in Fig. 14. One squeeze at 100 psi produces 95.2 lb/cu ft density, while 14 squeezes at 50 psi will produce the same density. Repetitive squeezing permits relief of locked-in stress, and further compaction after reapplication of load. A single high pressure squeeze with a flat plate is partially diverted to side walls. The contour squeeze frame prevents this by permitting proper distribution of the squeezing force.

ACKNOWLEDGMENTS

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REFERENCES

1. J. B. Caine and C. E. McQuiston, "The Theoretical Concepts of the Packing of Small Particles," AFS TRANSACTIONS, vol. 66, p. 36 (1958).
2. L. J. Pedicini "Packing Characteristics of Typical Foundry Sands," AFS TRANSACTIONS, vol. 66, p. 421 (1958).
3. R. W. Heine and T. W. Seaton "Density of Sand Grain Fractions of the AFS Sieve Analysis," AFS TRANSACTIONS, vol. 66, p. 40 (1958).
4. R. W. Heine – "Molding Sands, Molding Methods and Casting Dimensions," AFS TRANSACTIONS, vol. 64, p. 398 (1956).
5. R. W. Heine, E. H. King, and J. S. Schumacher; "The Jolt Test for Sand," AFS TRANSACTIONS, vol. 64, p. 415 (1956).
6. K. Terzaghi, *Theoretical Soil Mechanics*, J. Wiley & Sons.
7. J. M. Dallavalle, *Micrometrics*, Pitman Publishing Co., N. Y.

SAMPLE FOR RAPID MEASUREMENT OF GAS IN ALUMINUM

By H. V. Sulinski and S. Lipson

ABSTRACT

A rapid quantitative test for gas in aluminum has been developed. The gas content is determined by a single weighing of a 20-cc sample solidified at approximately 0.1 atmospheric pressure. A resin bonded shell mold and knock-off riser system of special design permits close control of as-cast specimen volume. Based on the reproducibility data, empirical values for gas content can be obtained with a standard deviation (σ) of 0.002 cc/100 gr.

The empirical gas content values determined by the test probably approximate the true gas content of the sample. Sampling temperatures between 1250 and 1450 F have no apparent effect on the test results. The constant volume sample reduced pressure solidification test can provide a quantitative measure of the gas contained in a melt within 4½ min from sampling, and with equipment sufficiently rugged to permit use directly on the production foundry floor.

INTRODUCTION

A great deal of effort has been directed, during the past few years, toward the development of techniques for the production of premium quality aluminum alloy castings. These castings are for applications where high strength and reliability are the primary requisites. These applications are for critically stressed castings such as are to be found in missile components, lightweight armored vehicles and in aluminum alloy armor. The techniques that have been developed include close control of the gas dissolved in the molten aluminum.

Gas contained in the melt is removed by chlorine or nitrogen degassing. The termination point of the degassing treatment can only be determined accurately if one has an accurate measuring procedure for the gas content of the melt. Fixed degassing times are not desirable because:

a) There is no assurance that all the gas has been removed if unusual conditions, such as humid atmospheres or poor ingot quality, are encountered.
b) Excess chlorination must be avoided to prevent depletion of certain alloying elements as magnesium in 356 alloy and removal of grain refining additions from the melt.

The reduced pressure solidification test has become an integral part of any process designed to produce high performance castings. The test is performed by permitting a sample of metal taken from the melt to solidify at reduced pressure. Gas precipitated from the melt occupies a volume which is related to the solidification pressure in accordance with Boyle's Law. If 0.1 of an atmosphere is used, the volume occupied by the gas is magnified by a factor of 10. The sample is usually evaluated on a qualitative basis. For these premium quality castings, however, qualitative estimates of the volume of gas in the melt are inadequate.

Either sectioning of the sample to detect evidence of gas holes or density measurements are employed to insure that the degassing operation has been taken to completion. Both of these methods for evaluation of the test specimen leave something to be desired. The sectioning method requires careful polishing of the sample to avoid smearing over the gas holes, and the density measurement is equally time consuming and cumbersome.

Attempts have been made to devise specially adapted balances to reduce the time required for density measurement.[1,2] One such attempt was made at this laboratory through the development of special apparatus designed to measure density by means of liquid displacements.[1] While this method was rapid, its accuracy was limited to $\pm$ 0.02 gr/cc, and the need for this special apparatus impeded its acceptance for production foundry use. In a further effort to improve the sensitivity and ease of specimen evaluation, the concept of a constant volume sample was evolved.

In consideration of the great strides made during the past few years in perfecting precision casting processes, it was postulated that a sample mold could be developed where the volume of the sample could be reproduced within narrow limits. A simple weighing would then suffice for a positive evaluation of the sample. After reviewing the processes which could be used, the shell mold process was selected for casting the constant volume samples. A program was then established for development of a constant volume sample for the reduced pressure solidification test.

METHODS, MATERIALS AND APPARATUS

Selection of Mold Type

The shell mold was selected as the medium into which the constant volume reduced pressure solidification samples were to be cast. This selection was made for the following reasons:

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1) High permeability – mold gases are readily evolved.
2) Non-hydroscopic nature – permits storage under typical foundry conditions.
3) Ease of manufacture – fulfills requirements of expendable mold.
4) Low thermal conductivity – eliminates need for preheating.
5) Dimensional accuracy – close mold volume control.

Sample Cup Design

The design of the specimen must be such as to insure that adequate feeding is provided under all conditions of test. This is because the problem of feeding the sample becomes most difficult and significant when the gas sample contains little or no gas. An improperly risered specimen under these conditions will develop shrinkage cavities, which on the basis of the test parameters will be indistinguishable from voids resulting from gas. A truncated cone design was adopted because of its favorable geometry with respect to feeding.

The size of the specimen was selected on the basis of the needs of the test. A large specimen would provide the maximum in sensitivity; however, the solidification time would become excessive and defeat the purpose of the test which is to provide a rapid measure of the gas contained in the melt. A compromise size of 20 cc was finally selected. This requires a solidification time of slightly better than 2 min, which on the basis of previous work, was considered to be optimum. By controlling the specimen volume at exactly 20 cc the conversion from weight to density could be readily performed by a simple mental computation. For this work, however, time was not taken to adjust the specimen volume to exactly fulfill this requirement.

Figure 1 is a detailed cross-sectional drawing of the sample cup mold. The choke-type gate serves a threefold function. In addition to its function as a channel through which the metal flows into the test specimen segment of the mold, it provides a hot spot for maintaining a fluid channel for feeding and a notch so that the riser segment of the casting can be readily separated from the constant volume specimen. The methods used for production of this mold will be discussed in the following section.

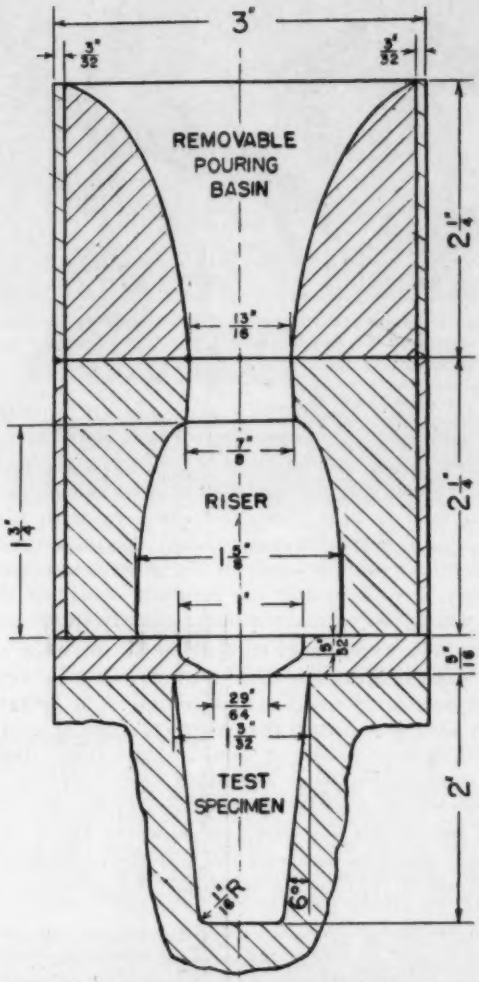


Fig. 1 — Cross-sectional drawing of the constant volume mold. The choke-type gate serves a threefold function: 1) a channel through which the metal flows into the test specimen, 2) provides a hot spot for maintaining a fluid channel for feeding, and 3) provides a notch so the riser segment can be readily separated from the specimen.

Sample Cup Production

Test Specimen Segment. The test specimen portion of the constant volume mold was made by the shell-molding process utilizing a semi-automatic shell-molding machine. The specimen molds were prepared from a shell-molding mixture consisting of 94.8 per cent silica sand (AFS Fineness 120-150), 5 per cent resin and 0.2 per cent kerosene. The process variables were controlled to produce a shell approximately ⅜-in. thick.

Figure 2 is a photograph of the pattern plate used to make the test specimen molds. The plate was designed so that 12 test specimen molds could be made simultaneously by simply cutting the shell (12 in. x 18 in.) up into twenty-four 3-in. squares (12 copes

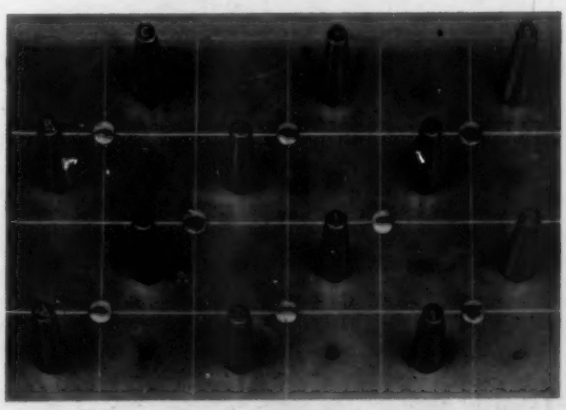


Fig. 2 — Shell mold pattern plate for making the test specimen portion of the constant volume mold. Twelve test specimen molds can be made simultaneously by cutting the shell into twenty-four 3 in. squares.

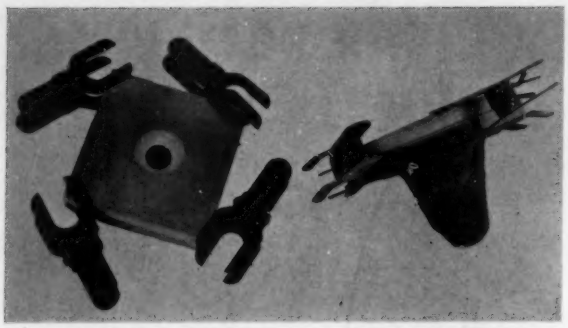


Fig. 3 — Constant volume shell mold showing method of assembly. The shell halves are clamped together by four pinch clamps.

and 12 drags). The pattern plate was an aluminum base material and the specimen inserts were cast iron pieces fastened to the plate by means of machine screws. These inserts produced a specimen cavity of approximately 20 cc. No attempt was made to adjust the individual specimen to produce specimen cavities of identical volume. Each of the inserts was appropriately identified, and the resulting specimen shell and casting also carried the insert identification.

The actual volume of each of the 12 specimen castings was calculated so that the test specimen weight could be related to its actual volume. The specimen shells were grouped so that for each series of tests all the shells were produced from a single insert. Before the shell halves were assembled, the backs of the copes were ground flat (approximately $\frac{5}{16}$-in. thick) on a belt sander and gated, as shown in Fig. 1, using a $\frac{29}{64}$- and 1-in. diameter drill. If the copes were to be made in a blowing machine, the grinding and drilling operations would be unnecessary. Two assembled test specimen molds are shown in Fig. 3. The shell halves were clamped together by four pinch clamps.

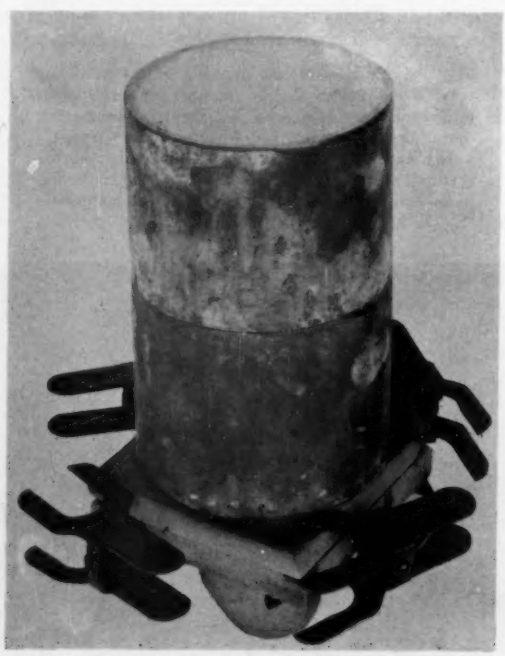


Fig. 4 — Completely assembled constant volume mold with plaster riser and pouring basin in place. The riser and pouring basin are encased in a 3 in. diameter hollow steel cylinder.

TABLE 1 – CHEMICAL COMPOSITION OF 356 ALLOY

Element	Wt. %
Silicon	6.5-7.5
Magnesium	0.2-0.4
Iron	0.6 Max.
Copper	0.2 Max.
Manganese	0.35 Max.
Zinc	0.3 Max.
Titanium	0.25 Max.
Other Elements, each	0.05 Max.
Other Elements, total	0.15 Max.
Aluminum	Remainder

Riser and Pouring Basin Segments. The riser and removable pouring basin were made out of a non-ferrous investment (70 per cent silica – 30 per cent gypsum) material consisting of 1¼ lb investment and 300 cc water. After the plaster molds had set (15 - 20 min) and the patterns had been removed, the molds were dehydrated and calcined in an oven for 16 hr at 600 F. A completely assembled constant volume mold is shown in Fig. 4. The plaster molded riser and pouring basin were encased in a 3 in. diameter hollow steel cylinder, $\frac{3}{32}$-in. thick and 2¼-in. in height. Encasing them in a steel cylinder enables them to be reused many times.

Plaster molds, when allowed to stand at room temperature for a period of time, will absorb some moisture. Thus, before actually using the plaster molds they were given an additional dehydration treatment. This treatment consisted of pouring molten metal into them twice and allowing the heat to soak in for a short period of time.

Melting Equipment, Material, and Procedures

Heats were melted in a 3000-cycle lift-coil induction furnace, and a gas-fired furnace using clay-graphite crucibles. The induction melts weighed approximately 25 lb, and the gas-fired melts approximately 15 lb. Commercial 356 alloy ingot melting stock was used throughout this investigation. The Federal Specification (QQ-A-60lb) of this alloy is given in Table 1.

Degassing treatments, where employed, were accomplished by bubbling chlorine gas through the melt for 10 min with the melt temperature at approximately 1300 F. Gaseous heats were obtained by poling the melt for 2 min with green wood.

Approximately 100 - 120 grams of metal were poured into each constant volume mold. The molds were all poured at 1300 ± 5 F. Temperature measurements were made using a portable potentiometer and a chromel-alumel thermocouple. The thermocouple was encased in a stainless steel tube.

Reduced Pressure Solidification Test Method and Apparatus

The reduced pressure solidification test is a method extensively used in experimental work and in industry for estimating the volume of gas in molten aluminum. The test is performed by permitting a sample

of metal taken from a melt to solidify at reduced pressure. When a sample of a melt is solidified under reduced pressure, the volume of the cavities formed in the metal by the entrapped gas will be inversely related to the solidification pressure (Boyle's Law). The gas content of the sample may be qualitatively evaluated by examining the appearance of the free surface of the specimen. A convex surface is indicative of high gas content, and a concave surface is indicative of freedom from gas. Quantitative information about the gas content can be obtained by measurement of the void volume present in the specimen.

A photograph of the reduced pressure tester used in this work is shown in Fig. 5. The unit conforms basically with the reduced pressure apparatus developed by Baker.[3] Essentially, the apparatus consists of a mechanical vacuum pump, mercury monometer, needle valve,* main vacuum control valve, vacuum release valve, rubber hose lead, pyrex glass vacuum chamber partly filled with brass shot and a rubber O-ring gasket. The mold was partially imbedded (up to its flange) in the shot to seat the mold and to provide a heat sink.

In an actual experiment run, the testing procedure was as follows: After a sample of metal taken from a melt was poured into the constant volume mold, the pouring basin was immediately removed and the vacuum chamber was closed. The main vacuum valve was then opened, and a pressure of 80 mm of mercury on the manometer was obtained by properly adjusting the needle valve. Finally, after 4 min the test mold was removed from the vacuum chamber after closing the main valve and opening the vacuum release valve.

RESULTS AND DISCUSSION

Gate

An investigation was made to determine the optimum size of the gate. To make this study, duplicate constant volume samples were cast using gates varying in diameter from 0.250 - 0.625 in. These samples were cast from an induction furnace heat of metal in the as-melted condition. All of the samples were solidified at atmospheric pressure.

After the riser segment of the casting was knocked off, the center portion of the top surface of each specimen was carefully ground flat to avoid introducing an error due to an irregular fracture. The samples thus obtained were then weighed on an analytical balance.

The results of this work are shown in Fig. 6. From the results obtained it can be seen that a minimum gate diameter of about 0.300 in. is necessary to insure satisfactory feeding of the constant volume segment of the casting. A gate diameter of 0.453 ($^{29}/_{64}$) in. was used for this work. Complete feeding of the casting is mandatory. An improperly fed constant volume sample will develop shrinkage cavities which, on the basis of the test parameters, will be indistinguishable from voids resulting from gas.

Experience has shown that the angle formed between the gate and specimen by virtue of drilling the $^{29}/_{64}$ and 1 in. diameter holes must be sharp and unrounded. Generally, a rounded angle will result in an irregular type fracture.

*The needle valve produced a controlled rate of leak in the system and was effective in control of solidification pressure.

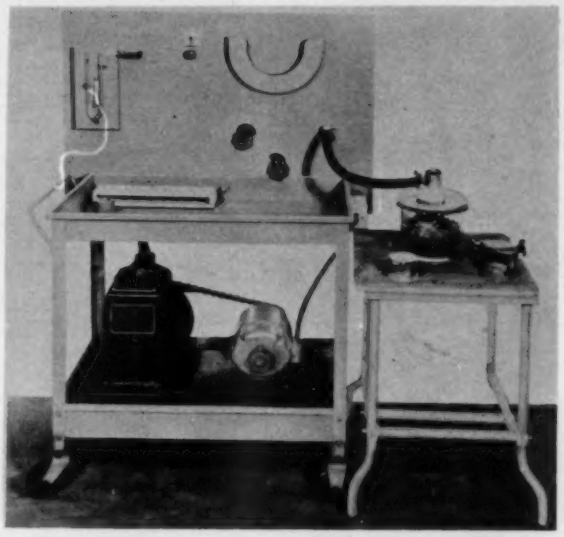


Fig. 5 — Reduced pressure test apparatus used in this work. The unit conforms basically with the apparatus developed by Baker.[3]

Reproducibility Tests

To determine how accurately the volume of the sample can be reproduced, 28 constant volume specimens were cast from two degassed induction furnace melts. Two series of constant volume molds dissimilar in volume were used in this investigation. All of the castings were solidified at atmospheric pressure.

The specimens thus obtained were processed and weighed as previously described. The specimens were weighed in air and ethyl alcohol (U.S.I. 200 proof). From these data their volume and density were then determined by Archimedes' technique.

The results of this work (Table 2) show that the constant volume sample can be reproduced within narrow limits. Maximum volume and density error obtained for the 20-cc sample was 0.054 cc and 0.007

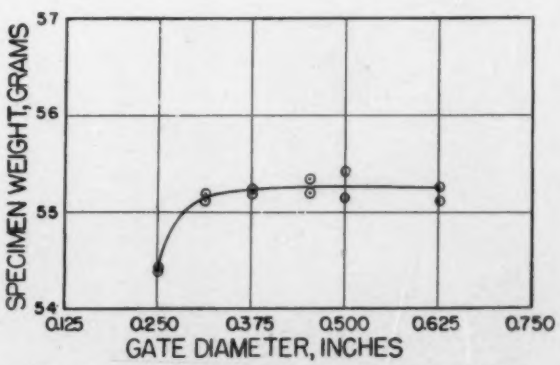


Fig. 6 — Gate diameter effect on the weight of the constant volume specimens. From the results obtained it can be seen that a minimum gate diameter of about 0.300 in. is necessary to insure satisfactory feeding of the castings.

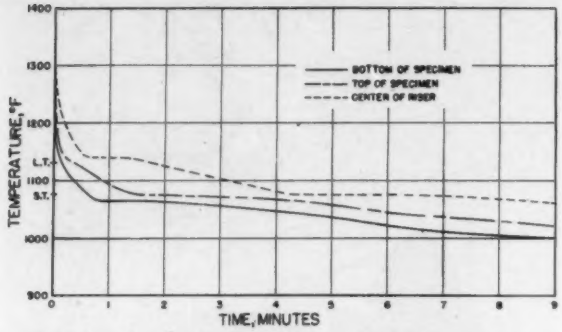


Fig. 7 — Temperature gradients within the constant volume casting when poured at 1300 F (alloy 356). The solidification of the specimen begins after approximately 9 sec, and ends slightly after 2 min. Complete solidification occurs after about 6¼ min.

gr/cc, respectively. These errors represent a gas content error of 0.004 cc of gas per 100 grams of metal (S.T.P.).

Standard volume and density deviation (σ) obtained was 0.028 cc and 0.003 gr/cc, respectively. These figures are equivalent to a gas content error of 0.002 cc of gas per 100 grams of metal (S.T.P.).

On the basis of the reproducibility data, a balance having a sensitivity of $\pm$ 0.05 grams, and a capacity in excess of 60 grams, would be adequate for all specimen weighings. Balances fulfilling these requirements are commercially available, and are sufficiently rugged for foundry floor use.

Solidification Time and Temperature Gradient Measurements

To determine the solidification time and the temperature gradients within the constant volume casting when poured at 1300 F, cooling curves were obtained. These curves were obtained by the use of a multipoint temperature recorder and three chromel-alumel thermocouples. One thermocouple was positioned in the center of the riser, and the other two were positioned just inside the top (⅛ - 3/16-in.) and bottom (1/16 - ⅛-in.) surfaces of the constant-volume specimen.

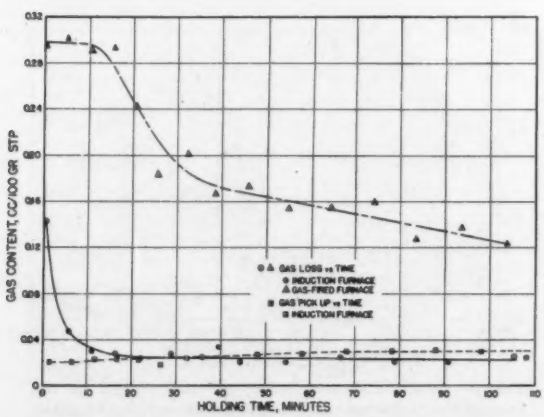


Fig. 8 — Gas loss and pickup vs. time curves for gassed and degassed heats of metal melted in induction and gas-fired furnaces.

TABLE 2—REPRODUCIBILITY TESTS RESULTS

Specimen No.	Vol. cc	Vol. Error cc	Density[1] gr/cc	Density[2] gr/cc	Density Error gr/cc
		Series 1			
1	19.792	−0.018	2.676	2.675	−0.001
2	19.819	+0.009	2.676	2.677	+0.001
3	19.815	+0.005	2.679	2.680	+0.001
4	19.769	−0.041	2.680	2.674	−0.006
5	19.756	−0.054	2.680	2.673	−0.007
6	19.801	−0.009	2.678	2.677	−0.001
7	19.819	+0.009	2.678	2.679	+0.001
8	19.778	−0.032	2.680	2.675	−0.005
9	19.853	+0.043	2.678	2.684	+0.006
10	19.823	+0.013	2.679	2.681	+0.002
11	19.847	+0.037	2.677	2.682	+0.005
12	19.830	+0.020	2.678	2.681	+0.003
13	19.828	+0.018	2.680	2.682	+0.002
14	19.769	−0.041	2.678	2.673	−0.005
15	19.829	+0.019	2.678	2.680	−0.002
16	19.815	+0.005	2.671	2.672	+0.001
17	19.840	+0.030	2.668	2.672	+0.004
18	19.823	+0.013	2.663	2.665	+0.002
19	19.790	−0.020	2.653	2.650	−0.003
	Average vol. of series 1 sample = 19.810 cc				
		Series 2			
20	19.472	−0.002	2.680	2.680	0
21	19.503	+0.029	2.679	2.683	+0.004
22	19.473	−0.001	2.680	2.680	0
23	19.443	−0.031	2.677	2.672	−0.005
24	19.452	−0.022	2.680	2.677	−0.003
25	19.461	−0.013	2.680	2.678	−0.002
26	19.506	+0.032	2.679	2.684	+0.005
27	19.487	+0.013	2.681	2.682	+0.001
28	19.466	−0.008	2.681	2.679	−0.001
	Average vol. of series 2 sample = 19.474 cc				

1) Density value obtained on individual specimens by Archimedes' Method.

2) Density value calculated by assuming the averaged vol. of the samples in each of the series.

The cooling curve obtained at each thermocouple location is shown in Fig. 7. From these curves it can be seen that solidification of the specimen begins after approximately 9 sec and ends slightly after 2 min. Complete solidification of the casting occurs after approximately 6¼ min.

Although complete solidification of the casting occurs after approximately 6¼ min, the casting can actually be removed from the vacuum chamber after 4 min, and the riser knocked off without producing any harmful effects. This is because at that time the specimen segment of the casting is completely solid, and the riser segment except for its center portion is also solid.

The thermal gradient studies show, as expected, that directional solidification does occur in the constant volume segment of the casting. This is desirable with respect to feeding of the test sample.

In addition, the results of this work show that the desired vacuum level is obtained in the chamber before solidification of the constant volume casting actually begins. Solidification of the sample begins after approximately 9 sec. With the reduced pressure tester used in this work a pressure of 80 mm of mercury was obtained in approximately 8 sec.

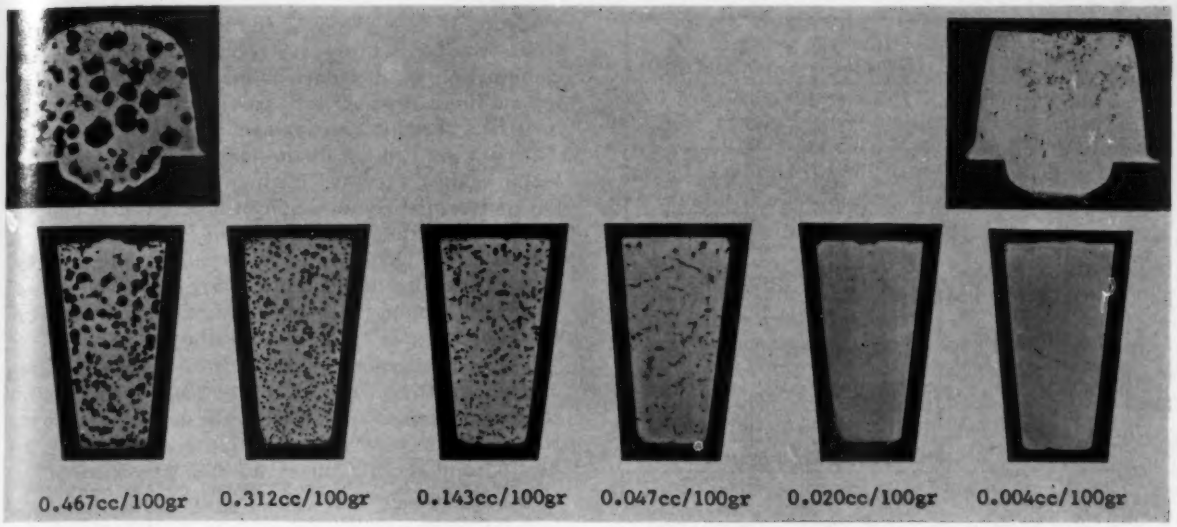


Fig. 9 — Sections of constant volume specimens showing gas contents over the range of 0.467 - 0.004 cc of gas (S.T.P.) per 100 grams of metal. The corresponding risers are shown for the specimens representing the extreme of the range.

Gas Loss and Pickup vs. Time

To determine how sensitive the constant volume sample is to a loss or pickup of gas in a melt, samples were cast at various time intervals from degassed and gassed heats of metal melted in the induction and gas-fired furnaces. All of the samples were solidified at a reduced pressure of 80 mm of mercury.

The volume of gas in the samples at standard conditions of pressure and temperature (S.T.P.) was calculated from the measured weight and the standard density of gas-free 356 alloy. These calculations were made by the use of the following formula (formula derivation given in Sec. 1 of Appendix):

$$\text{cc of Gas per 100 gr of Metal (S.T.P.)} = \frac{W_o - W_1}{W_1} \times K_s$$

where W_o = Weight of gas free sample
W_1 = Weight of reduced pressure sample
K_s = Constant

From the data obtained curves (Fig. 8) were prepared showing the sensitivity of the sample to the volume of gas present in the melts at various time intervals. The results of this work show that the constant volume sample is sensitive to a loss or pickup of gas in a melt. The curves obtained are quite smooth. The small amount of scattering obtained with the gassed gas-fired furnace melt is believed not to be due to the failure of the sample to accurately detect the loss of gas, but may be due to the fact that the melt temperature cycled during the holding time. Although all samples were cast at 1300 ± 5 F, the melt temperature throughout the run cycled between 1250 and 1350 F.

Figure 9 is a photograph of a series of sections of constant-volume specimens showing gas contents over the range of 0.467 - 0.004 cc of gas (S.T.P.) per 100 grams of metal. The corresponding risers are also shown for the samples representing the extreme of the range. Calculation of the apparent gas content of these risers (0.511 and 0.026 cc/100 gr) revealed that the riser segment contained a larger apparent volume of gas than the specimen.

An investigation was therefore conducted to determine whether shrinkage or gas segregation was responsible for this difference. Samples were cast containing various amounts of gas. These castings were solidified under a reduced pressure of 80 mm of mercury for 7 min. From the castings thus obtained, the density of the riser and specimen portions of the castings were determined.

The data obtained were plotted as shown in Fig. 10. Each point represents a specimen and riser density value obtained from one casting. Along with plotting the experimental points, a line was drawn representing the theoretical* specimen and riser density relationship line upon which all of the points should fall. From Fig. 10 it is evident that none of the points falls upon the line, and all are just below it.

Figure 11 shows the microstructure of specimen (a) and riser (b) segments of a relatively gas-free constant-volume casting. An examination of these photomicrographs shows the presence of shrinkage voids in the riser (b), which may be responsible for the excessively low riser density values. The density and gas content calculations made were made with the assumption that all voids present in the specimen and riser were due solely to entrapped gas.

The presence of shrinkage voids in the riser may possibly be explained on the basis that little if any directional solidification occurs in the plaster molded riser.

Sampling Temperature Effect

An analysis of the mechanism of hydrogen evolution[4] from solidifying aluminum alloy melts suggests that the sampling temperature may be a factor affecting the results obtained. When the gas level in the melt is high, higher sampling temperatures would sus-

*See Sec. 2 of Appendix for method used for calculating theoretical specimen and riser density values.

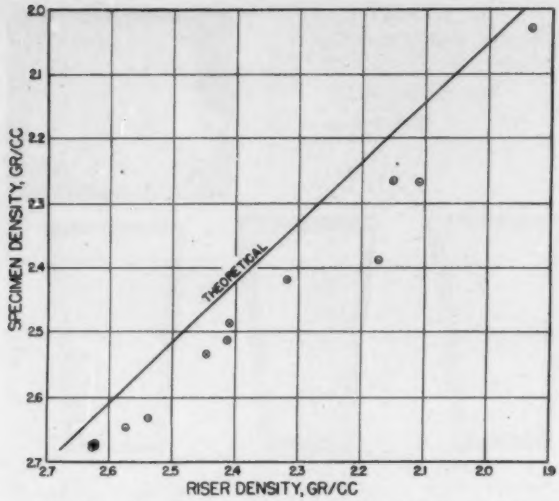


Fig. 10 — Experimental relationship between riser and specimen densities. The theoretical line is included for comparison purposes.

tain the sample in the liquid state for a longer time during the test, increasing the possibility for loss of gas through the metal surface. Practical difficulties, however, prevent a direct evaluation of this variable.

As an indirect method for determining whether sampling temperatures significantly affected the results, a series of melt samples were taken during the time interval that induction and gas-fired furnace melts[4] were being heated and cooled between 1220 and 1495 F. Approximately 34 min elapsed between taking the first and last samples in each melt. The samples were all solidified under a reduced pressure of 80 mm of mercury for 7 min. It was expected from previous holding time data that the gas contents would vary with time, and that if a temperature effect were detected it would manifest itself in an abnormal behavior in the gas content versus melt temperature relationship.

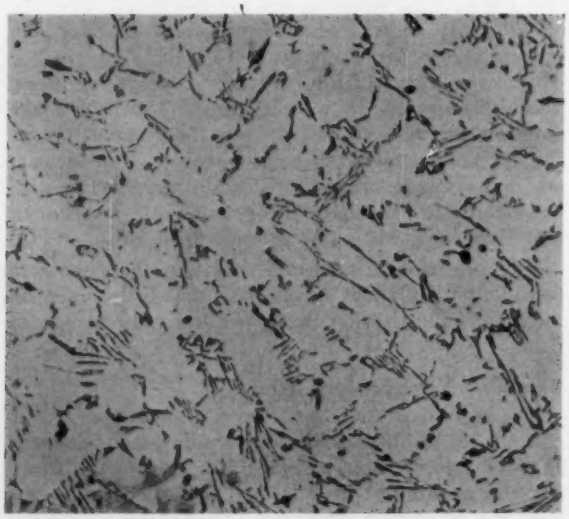
Figure 12 is a graph of the data obtained in the course of the heating and cooling experiments. An examination of this data shows that on heating in the gas-fired furnace the apparent gas content of the melt rises sharply with increase in melt temperature. This is a verification of the long recognized foundry taboo against excessive heating of aluminum melts. On cooling, the combination of lowered melt temperature and holding time results in a loss of gas from the melt. The melts prepared in the induction furnace showed a similar trend but at a greatly lowered level of gas, in keeping with previously observed behavior of induction melted aluminum alloys.

These observations conform with expected trends. It is apparent that rate of loss of gas is a function of time, temperature and the ambient atmosphere in the furnace. With due consideration of these facts, the heating and cooling curves for the gas melted metal (Fig. 12) show a hysteresis type function. Their relative positions indicate an understandable lag in gas pickup upon heating, and a similar lag in gas loss upon cooling. In the experiment conducted with induction melted metal, the gas level is sufficently low and the rates of gas loss sufficiently high (holding time experiment) that the hysteresis effect is not in evidence.

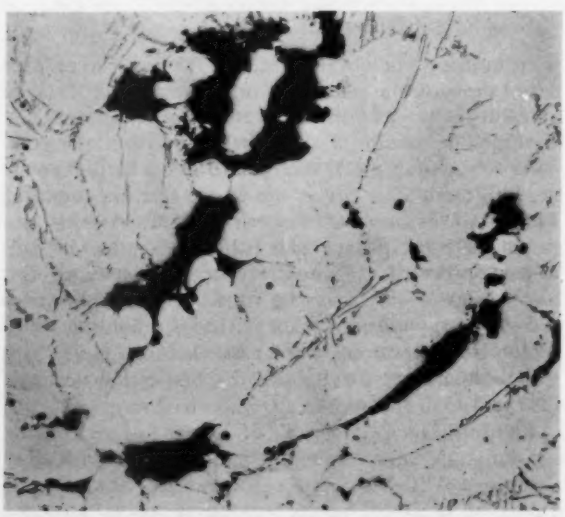
These data give no evidence of any significant gas loss at high sampling termperatures. Since the goal of the constant volume sample test for gas in aluminum melts is the detection of the point where the gas level approaches zero, it is apparent that loss of gas from a melt of low gas content cannot be a factor in the test.

Constant Volume Sample Reduced Pressure Test

This test is a quantitative test for determining the gas content of an aluminum-base alloy melt. With the constant volume sample, the volume of gas contained in a melt is determined from the weight of the sample, and the curve shown in Fig. 13. This curve shows the relationship of sample weight to cc of gas per 100 grams of metal (S.T.P.). The curve was ob-



a) Specimen (density 2.673 gr/cc).



b) Riser (density 2.628 gr/cc).

Fig. 11 — Photomicrographs of the specimen and riser segments of a constant volume casting showing the presence of shrinkage voids in the riser. Light 3% H.F. etch. 100 ×.

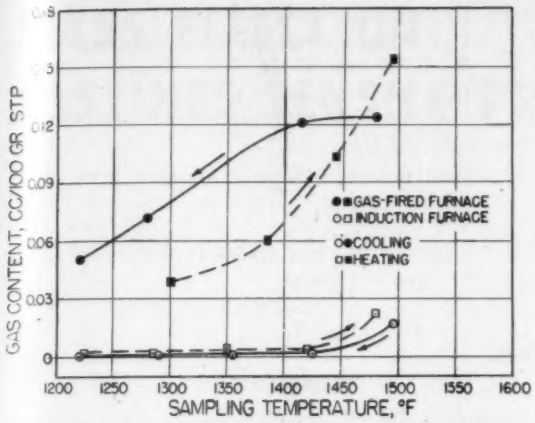


Fig. 12 — Effect of sampling temperature on the gas content of constant volume samples from induction and gas-fired melts. On heating in the gas-fired furnace the apparent gas content of the melt rises sharply with increase in melt temperature.

tained by solving the following equations:

$$\text{cc of Gas per 100 gr Metal (S.T.P.)} = \frac{P_2}{P_1} \times \frac{T_1}{T_2} \times 100 \left(\frac{W_1}{V_1} - \frac{1}{D_0} \right)$$

It is evident from the procedure described that the test can be performed fairly rapidly. In actual practice, it has been found that the gas content of a melt can be obtained within approximately 4½ min.

The simplicity and rapidity of the test is such that it may be used as a production test for controlling the gas content of aluminum-base alloy melts in an industrial foundry.

Measured Gas Content vs. True Gas Content

The usefulness of the reduced pressure solidification test, whether for qualitative or quantitative measurement of gas in molten aluminum, has been demonstrated by its acceptance as a production foundry control test. The question may be raised, however, whether the calculated gas content based essentially upon the volume of voids found in the reduced pressure sample, is the true gas content of the melt. The answer to this question is no. This is so because the number obtained is based upon certain simplifying assumptions, which are related to the characteristics of the test and to the method of computation. These are:

1) That the solid solubility of hydrogen is negligible and that all the gas originally contained in the liquid is expelled during the test.
2) That none of the gas precipitated from the sample is lost.
3) That the voids in the sample contain gas at a pressure equal to the sum of the reduced pressure in the system plus the metallostatic head.
4) That the base density of the metal in the melt is accurately known.

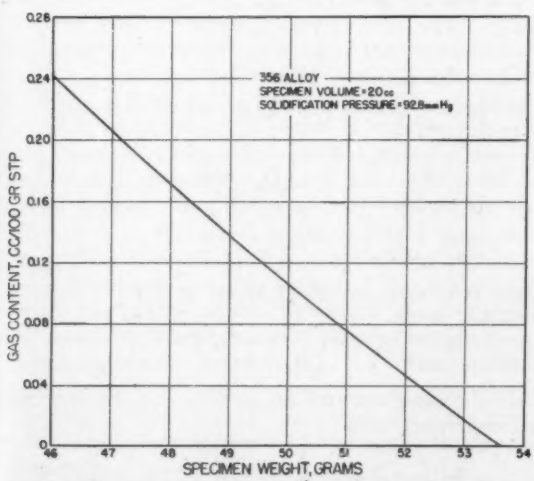


Fig. 13 — Curve showing the relationship of specimen weight to gas content (weight to cc of gas per 100 grams of metal, S.T.P.). The curve was obtained by solving equation given in paper.

There is much evidence that suggests that the use of these assumptions does not lead to any large error. Ransley and Neufeld[5] have determined that the equilibrium solid solubility of hydrogen in aluminum is small, being 0.036 cc/100 gr at 760 mm pressure, or approximately 0.012 cc/100 gr at the test pressure. Since the gas in solid solution does not produce unsoundness in the casting, its presence is not really a factor in the test. Rosenthal and Lipson[4] have shown that the total void volume varies with the atmospheric pressure over the samples as predicted by Boyle's Law.

If gas bubbles were to escape under reduced pressure, this relationship would not hold. The uniform distribution of the gas holes in the sectioned samples, shown in Fig. 9, is further evidence that there is no appreciable tendency for the gas bubbles to float to the surface of the sample.

The careful design of the specimen, together with its highly efficient plaster riser, minimizes the possibility for occurrence of shrinkage voids. The thermal gradient measurements are further evidence of this fact.

Variations in alloy chemistry within specification limits result in differences in base density. This would be a source of error when used as the basis for the gas content calculation. Under reduced pressure solidification of 0.1 atmosphere, this error is reduced by a factor of ten, and can be regarded as negligible.

On the basis of this discussion, it is apparent that the values obtained approximate the true gas content of the alloy. At any rate, the volume of gas, as determined from the sample, is that which is available to produce unsoundness in castings poured from the melt. One additional point should be made relative to the empirical value obtained in the test. The solid solubility of hydrogen at atmospheric pressure exceeds that at the test pressure by 0.024 cc/100 gr. Where the measured gas content of the melt is no greater than this value, it may be assumed that all of the gas in the metal can be retained in solution under conditions of atmospheric solidification.

CONCLUSIONS

1) Based on the reproducibility data, empirical values for gas content can be obtained with a standard deviation (σ) of 0.002 cc/100 gr. This value is based upon the following:
   a) Standard volume deviation (σ) of 0.028 cc in a 20 cc sample.
   b) Standard density deviation (σ) based only on the weight of the sample of 0.003 gr/cc.
2) The empirical gas content values determined by the test probably approximate the true gas content of the sample.
3) Sampling temperatures between 1250 and 1450 F have no apparent effect on the test results.
4) The constant volume sample reduced pressure solidification test can provide a quantitative measure of the gas contained in a melt within 4½ min. from sampling, and with equipment sufficiently rugged to permit use directly on the production foundry floor.

ACKNOWLEDGMENT

The authors are indebted to the Department of the Army for permission to publish this paper. The authors express their appreciation to H. Markus, Dir., Met. Rsch. Laboratory, and H. Rosenthal for their review and assistance in carrying out this work. The opinions expressed in the paper are those of the authors, and are not necessarily endorsed by the Department of the Army.

REFERENCES

1. S. Lipson, A. Saia, and H. Rosenthal, "Gravitometer for Rapid Density Measurement of Solids," A.S.T.M. Bulletin, No. 205, April 1955, pp. 60-63.
2. R. W. Ruddle, "An Improved Technique for Determining the Porosity of Cast Specimens," *British Foundryman, vol. 50,* December 1957, pp. 605-615.
3. W. A. Baker, "Microporosity in Magnesium Alloy Castings," *The Journal of the Institute of Metals, vol. 71,* pp. 165-204 (1945).
4. H. Rosenthal and S. Lipson, "Measurement of Gas in Molten Aluminum," AFS TRANSACTIONS, *vol. 63,* pp. 301-305 (1955).
5. C. E. Ransley and H. Neufeld, "The Solubility of Hydrogen in Liquid and Solid Aluminum," *The Journal of the Institute of Metals, vol. 74,* pp. 599-620 (1948).

APPENDIX

1) Calculation of volume of gas per 100 grams of metal assumption: All voids present in the sample are due to entrapped gas.

$$\text{cc of Gas in Sample} = \frac{W_o - W_1}{D_o}$$

where W_o = Weight of gas free sample, gr
W_1 = Weight of reduced pressure sample, gr
D_o = Density of gas free 356 alloy, 2.68 gr/cc

$$\text{cc of Gas per 100 gr of Metal} = \frac{W_o - W_1}{D_o} \times \frac{100}{W_1}$$

The volume of gas at standard conditions of temperature and pressure in accordance with Boyle's and Charles' Laws is as follows:

$$\text{cc of Gas per 100 gr of Metal (S.T.P.)} = \frac{W_o - W_1}{D_o} \times \frac{100}{W_1} \times \frac{T_1}{T_2} \times \frac{P_2}{P_1}$$

where T_1 = 273 degrees K
T_2 = Solidus temperature of the alloy, 852 K
P_1 = 760 mm of Hg
P_2 = Solidification pressure (vacuum pressure + metallostatic pressure, 92.8 mm of Hg

$$\text{cc of Gas per 100 gr of Metal (S.T.P.)} = \frac{W_o - W_1}{W_1} \times \frac{100 \times T_1 \times P_2}{D_o \times T_2 \times P_1}$$

$$\frac{100 \times T_1 \times P_2}{D_o \times T_2 \times P_1} = \text{Constant } K_s$$

Then,

$$\text{cc of Gas per 100 gr of Metal (S.T.P.)} = \frac{W_o - W_1}{W_1} \times K_s$$

2) Calculation of theoretical specimen and riser density assumptions:
   1) All voids present in the specimen and riser are due to entrapped gas.
   2) The same amount of gas exists in the specimen as in the riser

D_1 = Density of specimen (or riser)
D_o = Density of gas-free specimen (or riser)

$$D_1 = \frac{\text{Wt.}}{\text{Vol.}}$$

$$\frac{1}{D_1} = \frac{\text{Vol.}}{\text{Wt.}} = \frac{\text{Vol. of Metal + Vol. of Gas}}{\text{Wt. of Metal + Wt. of Gas}}$$

Neglecting the weight of the gas, then

$$\frac{1}{D_1} = \frac{\text{Vol. of Metal}}{\text{Wt. of Metal}} + \frac{\text{Vol. of Gas}}{\text{Wt. of Metal}}$$

$$\frac{\text{Vol. of Gas}}{\text{Wt. of Metal}} = \frac{1}{D_1} - \frac{1}{D_o}$$

$$\text{cc of Gas per 100 gr Metal} = 100\left(\frac{1}{D_1} - \frac{1}{D_o}\right)$$

$$\text{cc Gas per 100 gr Metal (S.T.P.)} = \frac{P_2}{P_1} \times \frac{T_1}{T_2} \times 100 \times \left(\frac{1}{D_1} - \frac{1}{D_o}\right)$$

$$\text{Specimen: cc of Gas per 100 gr Metal (S.T.P.)} = \frac{92.8}{760} \times \frac{273}{852} \times 100 \times \left(\frac{1}{D_s} - \frac{1}{2.68}\right)$$

$$= K_s\left(\frac{1}{D_s} - \frac{1}{2.68}\right)$$

$$\text{Riser: cc of Gas per 100 gr Metal (S.T.P.)} = \frac{84.1}{760} \times \frac{273}{852} \times 100\left(\frac{1}{D_r} - \frac{1}{2.68}\right) = K_r\left(\frac{1}{D_r} - \frac{1}{2.68}\right)$$

Since the same amount of gas exists in the specimen as in the riser then

$$K_s\left(\frac{1}{D_s} - \frac{1}{2.68}\right) = K_r\left(\frac{1}{D_r} - \frac{1}{2.68}\right)$$

Knowing the density of either the specimen or riser, the other can be calculated.

MASS EFFECT ON CASTINGS TENSILE PROPERTIES

a rationalization

By G. W. Form, P. J. Ahearn and J. F. Wallace

ABSTRACT

Previously published data on the effect of section size on the tensile properties of cast metals was considered to rationalize the effect of mass on these properties. Analysis of this information indicated that solidification time was of greater significance than section size. Accordingly, section-size data were converted to solidification times, and the influence of this latter casting condition on tensile properties was evaluated. A study of the relations between these factors resulted in the formulation of a hypothesis for predicting the sensitivity of tensile strength to solidification time.

It has been postulated that tensile strength is reduced by increasing solidification times if the ductility during tensile testing is insufficient to exceed the necking strain. However, as long as the metal is sufficiently ductile to exceed this limiting value, the tensile strength is virtually unaffected by changing solidification times. This hypothesis has practical value by permitting the prediction of the tensile behavior of a cast metal at various section sizes on the basis of limited test data.

INTRODUCTION

The influence of section size on the tensile properties has been shown by several investigators.[1-13] It is notable that the effect of mass on strength and ductility differs widely from one metal to another. For instance, tensile strength of gray iron decreases by over 40 per cent if the thickness of the casting is increased from ¼ to 1¼ in.[1,2] The strength of nodular iron of similar chemical composition, on the other hand, is little affected by changing the section size.[3]

The mechanism whereby the tensile strength and ductility is affected by different casting thicknesses is generally discussed in each of the references.[1-13] The role of section size on the graphite flake size and matrix structure has been shown. The slightly lower ductility in heavy cast-steel sections has been attributed to the larger grain size and greater concentration of nonmetallics compared to lighter sections. Recent work has brought out that a more rapid rate of solidification of aluminum alloys[12] refined the size and distribution of intermetallics, thereby improving tensile properties. It has also been shown that faster solidification rates of magnesium-aluminum-zinc alloys reduce the grain size and concentration of intermetallics[11,13] and result in improved properties.

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These references, however, fail to consider some characteristics of the tensile test itself that can explain many factors of this section sensitivity. These features of the tensile test will be discussed, and the characteristics will be related to the section sensitivity of several cast metals. It will be shown that knowledge of the tensile properties of the cast metal at one thickness (eventually two thicknesses) can be employed to predict within reasonable limits the tensile properties of that same metal at other thicknesses.

SECTION SIZE EFFECT

When considering the section size and its effect on tensile properties, it is wise to distinguished between two results of widely different origin. First, the strength of materials is known to be subject to a purely geometrical size effect, illustrated by a decrease in fracture strength as the specimen size is increased.[14] This effect is not associated with variation in metallurgical structure, since it occurs in a series of specimens machined from a stock which is homogeneous throughout.

This drop in fracture strength can be readily explained on the basis of Griffith's fracture hypothesis for brittle materials,[15] which uses the concept that internal flaws are invariably present in a material. The larger such a flaw (incipient crack), the lower the applied normal stress necessary to propagate it, i.e., the lower is the fracture strength. Since the probability of a large flaw of suitable orientation being present increases with the size of the specimen, the fracture strength is expected to decrease as the dimensions of the specimen become larger.

The size effect can also be explained by the probability of the occurrence of inhomogeneities or dislocations rather than cracks. This purely geometrical size effect is particularly pronounced in brittle materials.

The second effect is connected with the size of the casting (not that of the specimen), and can be demonstrated by performing a series of tensile tests on specimens of identical dimensions taken from castings of various section sizes. This effect may be called "metallurgical size factor," in contrast to the geometrical size factor.

SOLIDIFICATION RATE EFFECT

On the basis of many published results,[1-7] one is led to assume that a unique relationship exists between tensile strength and size of a casting. However,

TABLE – CHEMICAL COMPOSITION OF MATERIALS CONSIDERED IN THIS PAPER

Material	Chemical Composition, %	Remarks
Aluminum – 4.5% copper (grain-refined)	Cu-4.5, Fe-0.3 Si-0.4, Ti-0.1	Solution heat treated and aged for 12 hr at 310 F.[9]
Medium Carbon Steel	C-0.27, Mn-0.63 Si-0.22, Al-0.3	Annealed at 1650 F for 1 hr per in. of section.[5]
Magnesium-AZ-92 Alloy	Al-9.21, Zn-2.07 Mn-0.14, Si-0.012	Solution heat treated and aged for 4 hr at 500 F.[11]
Nodular Iron	C-3.30, Si-2.78, Mn-0.26, Mg-0.051, Ni-0.890, P+S <0.1	Heated to 1768 F for 1 hr, furnace cooled to 1350 F and held for 5 hr (Annealed for high ductility).[3]
Gray Iron	A.S.T.M. Spec. A48-56, Class 30	Arbitration bar B. As-cast.[2]

recent studies[10,11] have shown conclusively that this is not necessarily the case, since the tensile strength of aluminum and magnesium sand castings were greatly altered through chill action. In special cases, identical strength values may result in castings of widely differing section size. It follows, then, that tensile strength and size of casting do not relate uniquely to one another for a given cast metal.*

A surprisingly good correlation is obtained, on the other hand, if tensile strength is plotted against time of solidification. This correlation holds irrespective of whether an alloy is cast with or without chills. This evidence suggests that the important parameter controlling the strength of a casting is time of solidification, not section size. In order to check this point further, data were collected from the literature which permit a correlation between tensile strength and freezing time. Since most available results relate strength to section size, it was necessary to convert the latter into time of solidification. Such a conversion was made possible through published data.[9,17]

The Table lists the composition of the five materials for which the desired relationship could be established. The corresponding tensile strength results are plotted vs. solidification time in Fig. 1. The metals were selected in a heat-treated condition to avoid any influence of cooling rate below the solidification temperature. In order to evaluate only the solidification time effect, it was desirable to avoid the effect on tensile properties, produced by more rapid cooling through the critical temperature range (eutectoid transformation) in the ferrous alloys, and differences in cooling from a single solid solution phase or region into a two-phase region in the nonferrous alloys.

Nodular iron and steel were selected in the annealed state, and the aluminum and magnesium alloys in the solution heat treated and precipitation hardened condition for this reason. Only in gray iron was it necessary to use as-cast data because of lack of available information in other conditions. It is evident from Fig. 1 that a unique relationship between strength and freezing time, as proposed,[10] holds for all five materials over the range investigated. Furthermore, if the same data are replotted on a semi-logarithmic scale, as shown in Fig. 2, a straight line relationship results, which can be expressed by the equation:

$$\text{Tensile strength} = a - b \cdot \log t$$

where: t = freezing time in min (always a finite number).

a = tensile strength in psi after a solidification time of 1 min.

The slope of the straight lines, given by the con-

*It has also been shown that average thermal gradients in the solidifying casting do not correlate with tensile strength.[10,16] The actual thermal gradient at the instant of solidification is significant in determining the tensile strength but is difficult to determine experimentally.

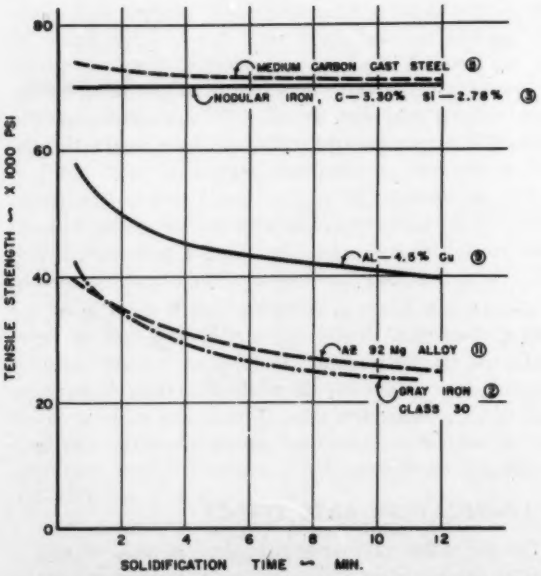


Fig. 1 — Effect of Solidification time on tensile strength of various cast metals.

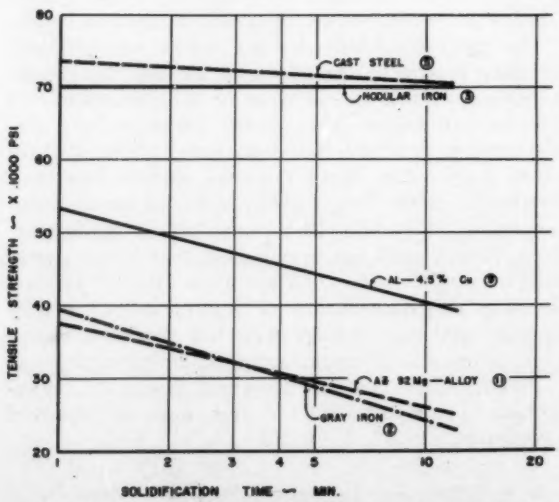


Fig. 2 — Semi-logarithmic plot showing the effect of solidification time on tensile strength.

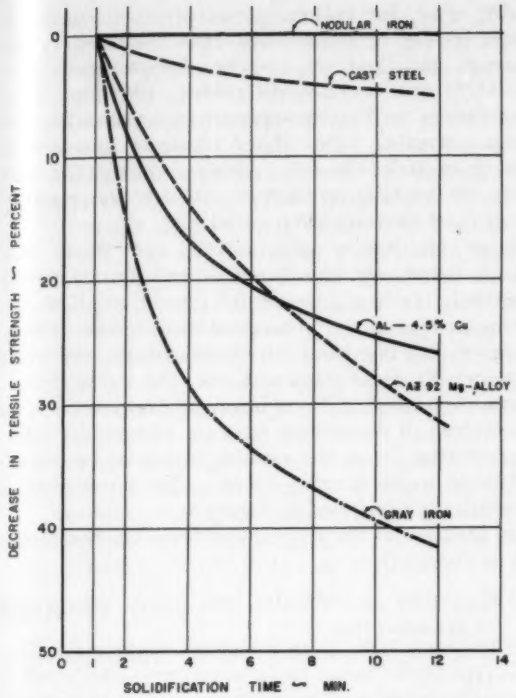


Fig. 3 — Per cent decrease in tensile strength as a function of solidification time for various cast metals.

stant b, could be taken as a measure of the sensitivity of strength to metallurgical size factor, since it indicates the rate of change of tensile strength with freezing time. This is not a practical choice, however, since a drop in strength from 100,000 to 95,000 psi would be expressed by the same value of b as a drop from 10,000 to 5,000 psi over the same interval of t. Yet, the low-strength material would obviously be classified as more sensitive to metallurgical size factor than the high-strength material.

The precentage drop in tensile strength with freezing time was, therefore, considered a more appropriate parameter to express the sensitivity of the tensile strength. A plot of per cent decrease in tensile strength vs. solidification time is shown in Fig. 3. The strength for a freezing time of 1 min was arbitrarily selected as 100 per cent. This graph shows clearly that gray iron, aluminum — 4.5 per cent copper, and commercial AZ-92 magnesium alloy are strongly affected by time of solidification in contrast to nodular iron and cast steel. The tensile strength of the two latter materials was altered little over the range of freezing times investigated.

TENSILE TEST CHARACTERISTICS EFFECT ON SECTION SENSITIVITY

The influence of solidification time on tensile properties is considerably affected by the characteristics of the tensile test itself. The tensile strength is defined as the ratio between maximum load and original cross-sectional area of the test specimen. This implies that the tensile strength can be computed from an engineering stress-strain curve only if the maximum load is really attained in a tensile test. In order that

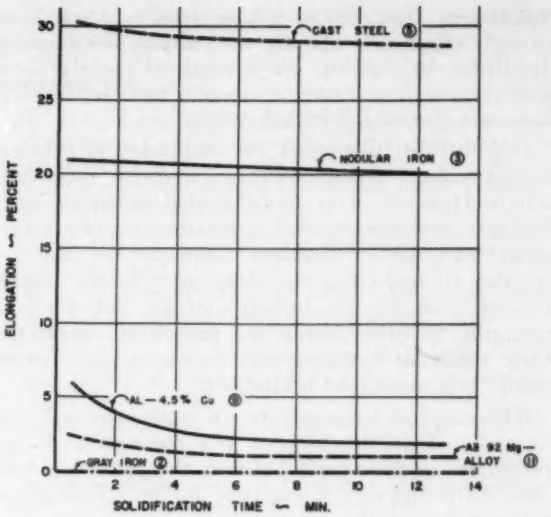


Fig. 4 — Effect of solidification time on per cent elongation of various cast metals.

the maximum load be attained, it is necessary that the ductility of a material at fracture be sufficient to at least reach the necking strain. The necking strain also is the strain at which the rate of increase in load carrying ability due to strain hardening is just equal to the rate of decrease of load carrying ability because of reduction of the cross-section of the test bar. At higher strains, the effect of reduction in cross-section predominates and the total load decreases.

It has been shown that an increase in freezing time will, in addition to producing other modifications in metallurgical structure, produce a larger as-cast grain size. This larger grain size will generally exhibit decreased ductility, since increasing grain size reduces the total grain boundary area. For a given composition, this produces a higher concentration of grain boundary segregates, providing more concentrated planes of weakness in the metal.*

As long as the fracture elongation remains larger than the necking strain, an increase in freezing time (no matter how much it affects ductility) will not reduce the maximum load or tensile strength appreciably. If fracture elongation falls below the necking strain, however, the tensile strength is influenced by the freezing time. This hypothesis proposes that a critical ductility value exists at the necking strain or the strain at which the loss in load carrying ability, because of the smaller cross-sectional area of the specimen, exceeds the increase in load carrying ability from strain hardening. When the cast metal has a lower ductility than this critical value, the tensile strength becomes sensitive to the metallurgical size factor.

In order to check this theory, ductility values for the five metals considered were plotted against time of solidification (Fig. 4). It can be seen, that those materials such as gray iron and the aluminum and

*It must be appreciated that a thorough study of the metallurgical processes taking place during freezing is necessary to assess their resultant effect on mechanical properties.

magnesium base alloys, whose tensile strength is strongly affected by freezing time, display low fracture ductilities. In contrast, the strength of nodular iron and cast steel is little influenced by solidification time, and also exhibit high ductility.

It is difficult to predict the magnitude of necking strains without the stress-strain curves for these materials. However, it is doubtful that materials with less than about 5 per cent elongation reached maximum load before fracture. Consequently, the respective strength data for these more brittle metals in Figs. 1-3 do not indicate tensile but fracture strengths. In other words, the stress-strain curve for these materials intersects the brittle strength curve before maximum load is attained.

The effect of freezing time on tensile strength can be analyzed on the basis of this concept. It is appropriate to distinguish between the following possible influences of freezing time on the tensile properties of cast metals.

1) Neither tensile strength nor elongation are affected by variation in solidification time.
2a) Only elongation is affected.
2b) Only tensile strength is affected.
3) Both tensile strength and elongation are affected.

Behavior (1) is unlikely over a wide range of freezing time, for reasons previously mentioned, although it may be approached over a limited range. Behaviors (2a) and (3) can be demonstrated with the aid of true stress-strain curves. The four schematic curves in Fig 5a represent behavior (2a) for different freezing times. They illustrate how tensile strength is little affected as long as elongation surpasses the necking strain even though the fracture strength and ductility are reduced.

Figure 5b depicts behavior (3) and shows that what is commonly considered a decrease in tensile strength is really a drop in fracture strength, if the maximum load is not attained during tensile testing. Behavior (2a) is typical for ductile materials, while behavior (3) is characteristic of the more brittle metals. A transition from behavior (2a) to (3) is to be expected, if ductility is strongly affected by solidification time or if the necking strain is barely exceeded at short freezing times. This latter case is schematically illustrated in Figure 5c.

On the basis of the above hypothesis, the five metals may be classified:

a) Cast steel and nodular iron follow the pattern of behavior (2a).
b) The magnesium alloy, AZ-92, typifies behavior (3).
c) The aluminum alloy may either be classified as

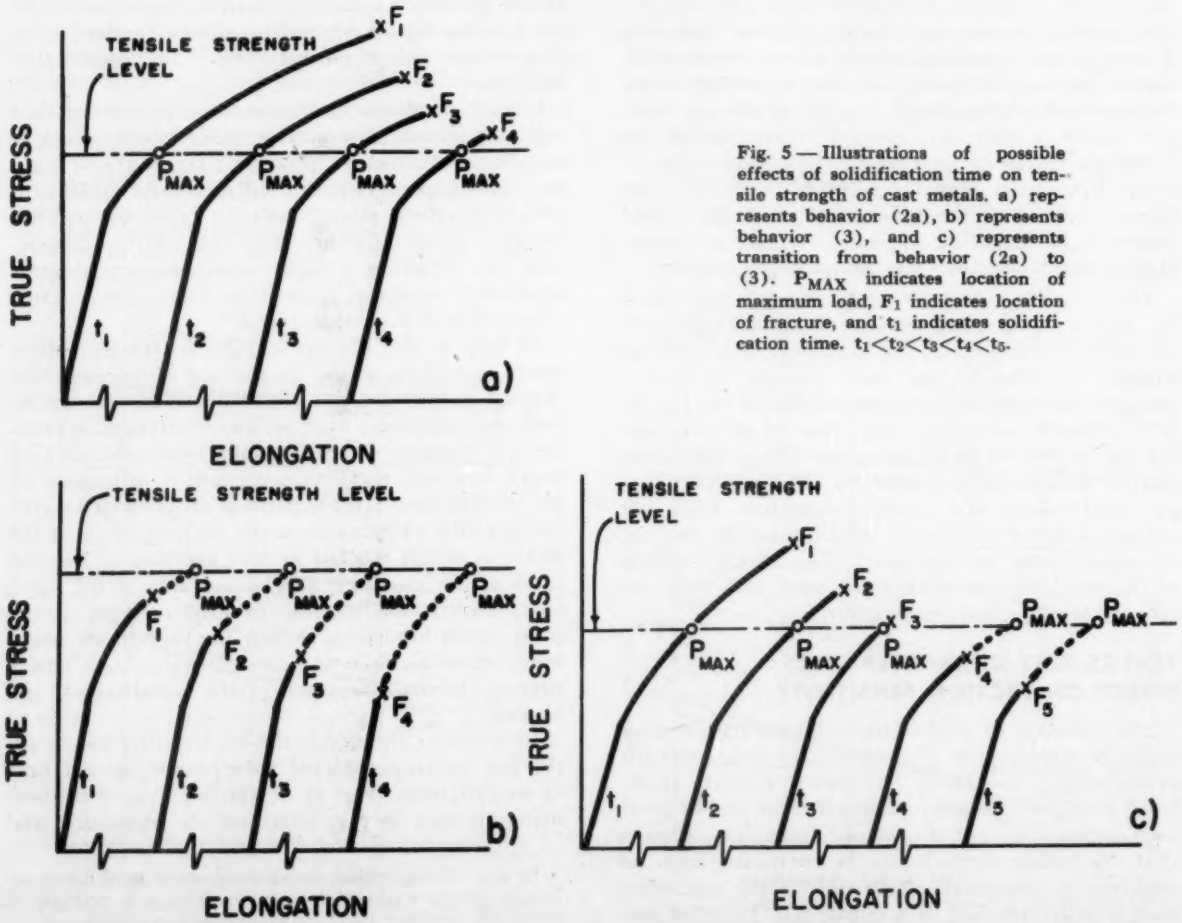


Fig. 5 — Illustrations of possible effects of solidification time on tensile strength of cast metals. a) represents behavior (2a), b) represents behavior (3), and c) represents transition from behavior (2a) to (3). P_{MAX} indicates location of maximum load, F_1 indicates location of fracture, and t_1 indicates solidification time. $t_1<t_2<t_3<t_4<t_5$.

case (3) or as a transition from case (2a) to (3), depending upon whether or not 6 per cent elongation (for t = 1 min) exceeds the necking strain of the alloy.

(d) Gray iron may appear, upon first examination, to be typical of case (2b), since its tensile strength is strongly affected by freezing time and the ductility remains unchanged. However, from the illustration given in Fig. 5, it is unlikely that behavior (2b) is displayed by any material. Gray iron should be classed in case (3). In brittle materials such as gray iron, changes in ductility, if they occur, are of the order of a fraction of 1 per cent, and are therefore not only hard to assess but also subject to considerable scatter. It is also entirely possible that a given material can behave as one class at one strength level and exhibit a different behavior at another strength. It is suspected, for example, that ultra-high-strength steel will have characteristics similar to case (3) whereas low-strength steel falls into case (2a).

SUMMARY

The hypothesis developed states that the sensitivity of the tensile strength of a cast metal to the solidification rate of the casting depends on whether the metal has sufficient ductility to attain or exceed the strain at necking. Utilizing this hypothesis, the sensitivity of a cast metal to solidification time or a metallurgical size factor can be assessed by a limited amount of tensile testing.

If the elongation of the tensile specimen from a rapidly solidified casting is less than the necking strain, the tensile strength will decrease with longer freezing times. If the fracture elongation exceeds the necking strain for this rapidly solidified bar, a second test of a specimen from a thicker section is necessary to classify the metal as sensitive or insensitive to freezing time. This hypothesis is based on the assumption that the stress-strain relationship is not appreciably altered by freezing time. This assumption is correct as far as known, but requires further investigation.

This theory is of practical value to metallurgists, since the prediction of tensile properties at various section sizes is possible from limited data. Such information can be employed to select cast metals for given applications and for alloy development.

REFERENCES

1. H. L. Campbell, "Relation of Properties of Cast Iron to Thickness of Casting," *Procedures*, A.S.T.M., *vol. 37*, Part 2, p. 66 (1937).
2. *Gray Iron Castings Handbook*, Gray Iron Founders' Society, Inc., Cleveland, p. 126 (1958).
3. C. F. Reynolds and H. F. Taylor, "Mechanical Properties of Spherulitic Graphite Cast Iron," AFS TRANSACTIONS, *vol. 60*, p. 76 (1952).
4. R. S. MacPherran, "Effect of Section and Various Compositions on Physical Properties of Cast Iron," *Procedures*, A.S.T.M., *vol. 29*, Part 2, p. 76 (1929).
5. C. W. Briggs and R. A. Gezelius, "The Effect of Mass upon the Mechanical Properties of Cast Steel," *Transactions*, A.S.M., *vol. 26*, p. 367 (1938).
6. H. C. Winte, "Gray Iron Castings Section Sensitivity," AFS TRANSACTIONS, *vol. 54*, p. 436 (1946).
7. J. F. Wallace, J. H. Savage and H. F. Taylor, "Mechanical Properties of Cast Steel as Influenced by Mass and Segregation," AFS TRANSACTIONS, *vol. 59*, p. 223 (1951).
8. D. Chandley, C. M. Adams, Jr. and H. F. Taylor, "A Quantitative Evaluation and Importance of Hydrogen in Aluminum Founding," AFS TRANSACTIONS, *vol. 63*, p. 607 (1955).
9. W. D. Walther, C. M. Adams, Jr. and H. F. Taylor, "Techniques for Improving Strength and Ductility of Aluminum Alloy Castings," AFS TRANSACTIONS, *vol. 62*, p. 219 (1954).
10. M. C. Flemings, P. J. Norton and H. F. Taylor, "Performance of Chills on High Strength-High Ductility Sand-Mold Castings of Various Section Thicknesses," AFS TRANSACTIONS, *vol. 65*, p. 259 (1957).
11. M. C. Flemings, R. W. Strachan, E. J. Poirer and H. F. Taylor, "Performance of Chills on High Strength Magnesium Alloy Castings of Various Section Thicknesses," AFS TRANSACTIONS, *vol. 66*, p. 336 (1958).
12. E. M. Passmore, M. C. Flemings and H. F. Taylor, "Fundamental Studies on Effects of Solution Treatment, Iron Content and Chilling of Sand Cast Aluminum-Copper Alloy," AFS TRANSACTIONS, *vol. 66*, p. 96 (1958).
13. R. D. Green, "The Effect of Cooling Rate on the Grain Size of Magnesium Casting Alloys," AFS TRANSACTIONS, *vol. 66*, p. 380 (1958).
14. W. Weibull, "A Statistical Theory of the Brittle Strength of Metals," Royal Swedish Institute for Engineering Research, *No. 151* (1939).
15. A. Griffith, "The Phenomena of Rupture and Flow in Solids," *Philosophical Transactions*, Royal Society of London, Series A, *vol. 221*, p. 163 (1921).
16. R. W. Ruddle, "Correlation of Tensile Properties of Aluminum Alloy Plate Castings with Temperature Gradients during Solidification," *Journal of the Institute of Metals*, *vol. 77*, p. 37 (1950).
17. R. E. Morey, H. F. Bishop and W. S. Pellini, "Heat Transfer Characteristics of Shell Molds," AFS TRANSACTIONS, *vol. 62*, p. 448 (1954).

RAPID HYDROGEN DETERMINATION FOR STEEL FOUNDRY CONTROL

By C. C. Carson and B. J. Alperin

ABSTRACT

A method is described for the determination of hydrogen in steel which is rapid, simple and economical. The method uses the hot extraction technique for extracting hydrogen from steel, and employs a Pirani gage for the analysis of the extracted gases for hydrogen by the thermal conductivity technique. The furnace section is made a part of the gas collecting volume. Comparison of results with those obtained with a conventional apparatus and procedure employed for the hot extraction technique show satisfactory agreement. For a sample size of approximately 2 g, the determination is completed in 5-10 min from the time that the sample enters the hot zone of the furnace. The time from the sampling of a molten bath to the completion of an analysis is about 15 min.

INTRODUCTION

As a result of the widespread interest in the significance and behavior of hydrogen in steel, there is a growing need for an economical, simple and rapid method for the determination of hydrogen in steel suitable for quality control purposes in a foundry. This paper describes a method that is being used in a steel foundry operated by the authors' company. The method is based on the hot-extraction technique for the extraction of hydrogen from steel, and on the thermal conductivity technique for the analysis of the extracted gases for hydrogen. An analysis can be completed in about 15 min from the time of sampling of a molten bath. The results compare satisfactorily with those obtained with a conventional hot-extraction apparatus for the extraction of the hydrogen, and a fractionation procedure for the analysis of the extracted gases.

The apparatus developed differs from the conventional hot-extraction apparatus in that the furnace section is a part of the gas-collecting volume, thereby eliminating the need for a diffusion pump, or a Toepler pump, for the transfer of the gases from the furnace section. Also, the analysis of the extracted gases by the thermal conductivity technique is accomplished with an economical, commercially available Pirani gage. Only one pump, a mechanical vacuum pump, is used. The cost of the apparatus is about half, and the time for analysis is less than half by comparison with a conventional hot-extraction apparatus employing the fractionation technique for the analysis of the extracted gases.

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ANALYTICAL METHOD CHOICE

Gas Extraction From the Sample

The three techniques commonly employed for the extraction of hydrogen from steel are:

1) Hot extraction.[1,2,3] The sample is heated at 600-1100 C in a vacuum.
2) Tin-fusion.[4,5,6] The sample is fused with tin in a vacuum.
3) Vacuum-fusion.[1,2,7] The sample is melted in a graphite crucible at 1375-1700 C in a vacuum.

The extracted hydrogen is normally collected in a separate volume along with varying amounts of carbon monoxide and nitrogen. The partial pressure of the hydrogen is then determined, and used in the calculation of the weight of the hydrogen by means of the perfect-gas law ($PV = nRT$).

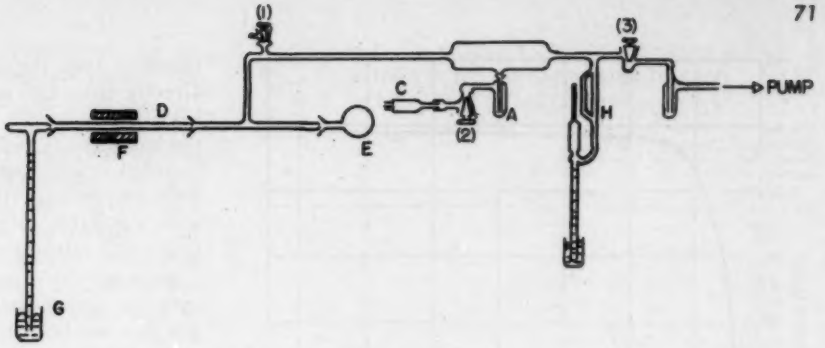
Differences of opinion exist as to the accuracy and suitability of the three techniques, and the absence of standards has not helped to resolve these differences. It is generally conceded, however, that, at least for practical purposes, all of the hydrogen is extracted by the vacuum-fusion technique. Carney, Chipman and Grant[4,8] carried out studies on the tin-fusion technique, and determined that the method had a precision of $\pm$ 0.1 ppm and an overall accuracy (including sampling molten steel) of a few tenths of 1 ppm.

Although the tin-fusion and vacuum-fusion techniques are not as convenient, and require a more cumbersome apparatus as compared to the hot-extraction technique, they have been termed superior by some investigators[4,6] because it was believed that they were faster and more accurate than the hot-extraction technique. However, Carson[9] compared the tin-fusion technique at 1150 C with the hot-extraction technique at 975 C on samples of low alloy steel weighing 2-3 g, and found that the hot-extraction techniques gave the same results in about the same time and with a better precision.

For the purpose of designing a simple, economical and rapid procedure for the determination of hydrogen in steel for quality control purposes, it was decided to make use of the hot-extraction technique.

Extracted Gases Analysis

There are several ways which can be and are used to determine the partial pressure of the hydrogen in



A– Trap
C– Pirani tube
D– Quartz tube
E– Receptacle for analyzed sample
F– Resistance furnace
G– Mercury lift
H– McLeod Gage

Fig. 1 — **Apparatus for hot-extraction technique.**

the mixture of hydrogen, carbon monoxide and nitrogen extracted from the sample. Thus, the hydrogen can be removed as it is in the so-called fractionation technique, and its partial pressure obtained either on the hydrogen removed or from the drop in the total pressure of the mixture. For another approach, a mass spectrometer can be employed to analyze the gas mixture without the necessity of separating the hydrogen from the mixture. Also, the thermal conductivity technique[10] can be employed to analyze the gas mixture for hydrogen without going through a separation step, and it was this technique which appeared to be the most suitable for the type of apparatus desired.

Shields, Chipman and Grant[10] used a special thermal conductivity cell of their own design and reported satisfactory results, obtaining a probable error of 0.12 ppm for the overall analytical method when the tin-fusion technique was employed. Carson[9] used an economical, commercially available Pirani gage for the thermal conductivity technique in conjunction with the hot-extraction technique, and observed only a small increase in the standard deviation ($\pm$ 0.03 to $\pm$ 0.11) as compared to a fractionation technique in which the hydrogen was converted to water and frozen out. This small loss in precision is not as important as the time (10-15 min per determination) that is saved for a quality control test. Accordingly, the gases were analyzed by the thermal conductivity technique using the commercially available Pirani gage.

APPARATUS

Figure 1 shows the hot-extraction apparatus which was constructed.

A major feature is the use of the furnace section as part of the calibrated collecting volume. This eliminates the need for a diffusion pump or a Toepler pump to transfer the extracted gases into a collecting volume. Not only is a simplification thereby effected, but the time for carrying out the transfer of gases is saved.

Except for the furnace section, which is made of quartz, the system is constructed of pyrex glass.

The Pirani gage has a range of 0-2,000 μ Hg. Trap A, which is used to freeze out water and other condensibles from the collected gases, also serves to protect the gage from condensibles. When the system is not in use, stopcock (2) is closed to protect the gage from vapors. Stopcock (2) has a 10 mm bore, and its hollow cavity has a volume of > 100 cc, which provides an appreciable volume of gas in the vicinity of the gage.

Stopcock (1) is used to open the system to the atmosphere and to introduce gases into the system for calibration purposes. The total volume of the system is about 1500 cc. The only pump used was a mechanical pump, and the vacuum obtained, < 0.1 μ Hg, was found to be satisfactory. The stopcocks are greased with a proprietary stopcock grease. The joints are standard taper seal 24/40 ground glass joints and are sealed with a proprietary wax.

CALIBRATIONS

The volume of the system can be determined by using the McLeod gage, according to the method described by Murray and Ashley.[11] Essentially, the method consists of expanding a gas trapped in the McLeod gage at a known pressure into the rest of the system, and calculating the total volume from Boyle's Law ($PV = K$). It was found advisable to attach a known volume of about 250 cc to the air inlet at stopcock (1). This made it possible to substantially increase the amount of gas that is trapped and then expanded into the system, thereby making possible a higher and more precise final pressure reading upon which to base the calculation of the total volume.

The calibration of the volume is carried out with the furnace at the operating temperature of 900 C. The volume thus obtained did not appear to change significantly with the pressure over a pressure range of 50-75 μ Hg.

Since the trap is immersed in either acetone at approximately – 85 C or, more conveniently, in liquid air, a correction must be added to any pressure reading to account for the fact that this part of the volume is below room temperature. This calibration is made simply by starting out with a dry, pure gas such as nitrogen at room temperature and any given pressure, then placing the bath around the trap and noting the drop in pressure. This is done for a series of pressures and a graph is constructed (Fig. 2).

Finally, a small correction is added to an observed pressure of hydrogen, to account for the fact that some hydrogen will diffuse through the hot quartz of the furnace section. This correction is directly proportional to the partial pressure of the hydrogen, and is conveniently measured as the drop in the hydrogen pressure during a 10 min period (about 1 μ Hg for a hydrogen pressure of 35 μ Hg for this apparatus).

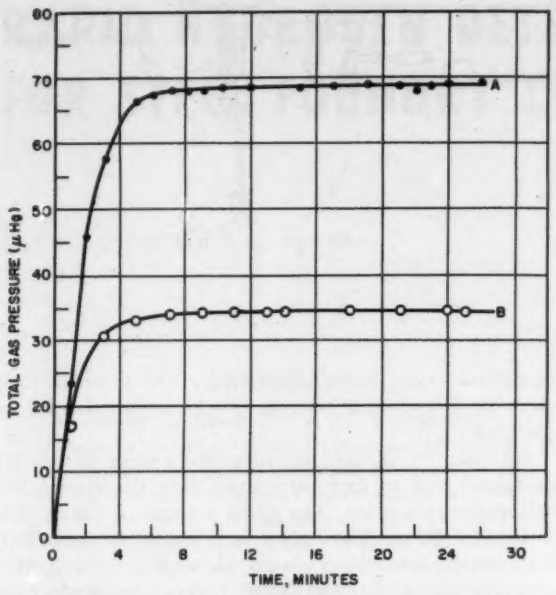


Fig. 2 — Rate of gas extraction. The pressure of the extracted gases is plotted against the time for two typical samples. A good indication of the hydrogen content can be obtained after an extraction period of about 5 min.

The Pirani gage is calibrated with nitrogen, i.e., it is attempted to have the Pirani gage read the same as the McLeod gage when nitrogen is in the system. With the Pirani gage calibrated, hydrogen is introduced into the system, and the reading on this gage is plotted against the reading on the McLeod gage for a series of hydrogen pressures. This curve, which represents 100 per cent hydrogen, is plotted on the same graph with the nitrogen calibration curve, which represents 0 per cent hydrogen. The graph is then used to determine the fraction of hydrogen in the gases extracted from a steel sample. This is possible because of the three gases extracted (hydrogen, nitrogen and carbon monoxide), two of the gases, nitrogen and carbon monoxide, have virtually the same thermal conductivity properties,[10] so that we are in effect dealing with a hydrogen-nitrogen system.

SAMPLING AND ANALYTICAL PROCEDURE

A spoonful of molten steel is obtained with a well-slagged spoon and killed with aluminum. A sample is then immediately taken from the spoon by carefully sucking (by means of a rubber bulb) some molten metal up a pyrex or vycor glass tube (¼-in. in diameter and more than 1 ft in length), then quenching the drawn sample in cold water. The steel "pencil" thus obtained can then be stored in liquid nitrogen until the time of analysis, or it can be subjected to the next step directly.

A portion of about 1 in. of the steel tube is lightly abraded on a dry grinding wheel, being careful to avoid overheating the metal. The steel tube is then placed in a vice, and about ½-in. of the abraded portion of the tube is broken off. Should a void be discovered in the sample, another sample should be taken from another portion of the tube. If care has been taken to keep the sample clean, it can be introduced directly into the analytical system via the mercury lift after its weight is obtained. If it is suspected that the sample has been contaminated (e.g., from being handled with fingers or from coming in contact with an unclean vise, etc.), then it should be washed with a suitable solvent such as acetone, and dried before being introduced into the system.

Stopcock (3) is open when the sample is introduced into the system and stopcock (1) is closed. After the gas that has been introduced into the system has been pumped out, stopcock (3) is closed, and the sample is moved by means of a magnet into the center of the hot zone of the furnace section. Extraction time is 10 min. At the end of that time, a reading is taken simultaneously on the McLeod gage and on the Pirani gage.

From the calibration graph obtained for the Pirani gage (containing one curve for 0 per cent hydrogen and one for 100 per cent hydrogen), the fraction of hydrogen in the gas mixture is determined. The use of the graph, shown in Fig. 3, will be illustrated with an example. Let us assume that the reading on the Pirani gage was 85.0 μ (point b, Fig. 3) when the reading on the McLeod gage was 50.0 μ (point a, Fig. 3). If the gas had consisted of 100 per cent hydrogen, then the reading on the Pirani gage would have been 92.3 μ (point c, Fig. 3). Therefore, the fraction of hydrogen in the gas mixture would be

$$\frac{(85.0 - 50.0)}{(92.3 - 50.0)} \text{ or } 0.829.$$

The McLeod gage reading, after being corrected for the presence of the trap, is then multiplied by

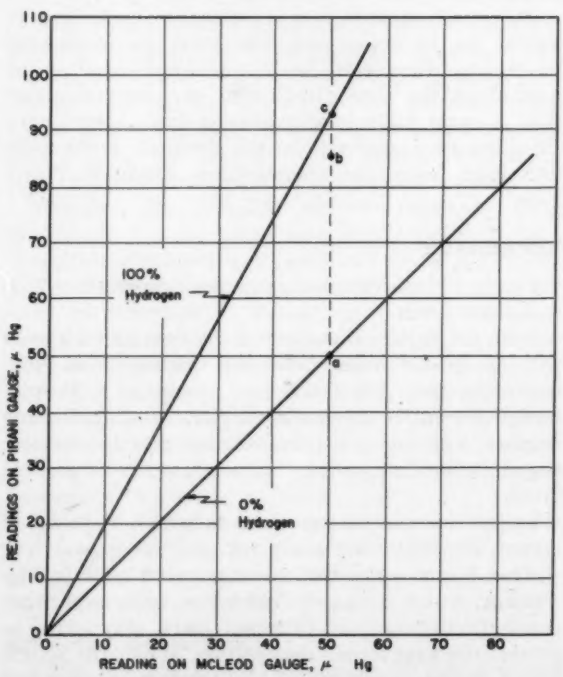


Fig. 3 — Calibration of Pirani gage for thermal conductivity method.

th fraction to give the partial pressure of the hydrogen at the time that the reading was taken. The correction for the amount of hydrogen that has diffused through the quartz is then added to this pressure to give P_H, the hydrogen pressure that is used in calculating the result. The weight of the hydrogen is calculated from the perfect gas law, $P_HV = nRT$ or $P_HV = W_HRT/M_H$, where W_H is the weight of the hydrogen, and M_H is its molecular weight. Asssuming a fairly constant room temperature, this can be put in the form $W_H = k'P_H$. Thus, the equation used in the calculation of the result is

$$(\text{ppm Hydrogen}) = \frac{k'P_H 0.10^6}{W_s}$$

$$\text{or } (\text{ppm Hydrogen}) = \frac{k\,P_H}{W_s}$$

where W_s is the weight of the sample.

After some practice, the entire determination from the time of sampling to the calculation of the result can be carried out in about 15 min.

In Fig. 2, the pressure of the extracted gases is plotted against the time for two typical samples. It can be seen that a good indication of the hydrogen content of the sample can be obtained after an extraction period of only 5 min or so, if the time is critical.

RESULTS

Table 1 compares results obtained by means of the described apparatus and procedure with results obtained with a conventional apparatus[9] employing hot-extraction at 975 C for extracting the gases and a fractionation technique for the analysis of the extracted gases. The chemical compositions of the samples are given in Table 2.

For each comparison, two samples were taken from the same pencil specimen of steel obtained as indicated in the section of this paper on sampling and analytical procedure. Tests had indicated that the hydrogen distribution is uniform along a sound pencil specimen. For example, the results obtained by the conventional apparatus and procedure on three specimens taken from one pencil were 3.2, 3.2 and 3.4 ppm of hydrogen.

It is to be noted that eight of the samples in Table 1 had hydrogen contents or 2-3 ppm. Previous studies on thermally charged specimens had indicated that in this range of hydrogen contents results obtained by the conventional apparatus have a standard deviation of ± 0.07 ppm. The standard deviation of the results obtained by the modified apparatus is therefore calculated to be ± 0.14 ppm in this range of hydrogen contents. Statistical treatment indicated that the average result obtained by the modified apparatus does not differ significantly from that obtained by a conventional apparatus.

TABLE 1 – COMPARISON OF RESULTS OBTAINED BY MODIFIED AND CONVENTIONAL APPARATUS

Sample No.	Hydrogen Content, ppm: Modified Apparatus	Hydrogen Content, ppm: Conventional Apparatus
1	2.92	2.87
2	2.90	3.06
3	6.23	6.07
4	3.49	3.84
5	4.14	4.10
6	5.46	5.47
7	2.83	2.76
8	2.43	2.23
9	2.91	2.82
10	1.97	2.15
11	2.27	2.32
12	2.43	2.67

CONCLUSIONS

The modified apparatus described in this paper for the determination of hydrogen in steel is economical to build and simple to operate. Results obtained do not differ significantly from results obtained with a conventional apparatus. A determination can be completed in about 15 min from the time of sampling with a precision that is satisfactory for quality control purposes.

ACKNOWLEDGMENTS

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When this paper was written, the authors included information and expressed opinions they believed to be correct and reliable. However, because of the constant advance of technical knowledge, the widely differing conditions of possible specific applications and the possibility of misapplication, neither the authors

TABLE 2 – CHEMICAL COMPOSITION OF SPECIMENS

Sample No.	Heat No.	Per Cent: C	Si	Mn	S	P	Cr	Ni	Mo	V	Cu	Al	Ti
1	2195	0.21	0.50	0.73	0.014	0.015	1.26	0.13	1.02	0.04	0.19	0.008	
2	2189	0.25	0.41	0.72	0.009	0.017	1.24	0.14	1.17	0.04	0.19		0.014
3	2191	0.22	0.57	0.65	0.009	0.016	1.18	0.15	1.10	0.06	0.20		0.021
4	2193	0.19	0.42	0.60	0.014	0.012	0.10	0.13	0.54	0.01	0.20		
*5		0.13	0.45	0.60	0.015	0.009	16.25	1.80					
*6													
*7													
8	2178	0.20	0.53	0.70	0.010	0.017	1.25	0.14	1.01	0.06	0.21		
9	2210	0.24	0.63	0.63	0.007	0.010	1.20	0.14	1.13	0.03	0.22		0.020
10	2205	0.18	0.35	0.73	0.008	0.017	1.30	0.16	1.05	0.18	0.20		
11	1820	0.18	0.50	0.54	0.008	0.013	1.29	0.14	1.03	0.20	0.22		0.022
12	1810	0.15	0.40	0.63	0.008	0.012	1.21	0.17	0.96	0.19	0.22	0.008	0.016

* Taken at different times during the same heat.

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REFERENCES

1. American Institute of Mining and Metallurgical Engineers, "Symposium on the Determination of Hydrogen in Steel", *A.I.M.E. Transactions, 162,* 353 (1945).
2. Iron and Steel Institute, Fourth Report of the Oxygen Sub-Committee, Section 2, Part B (a), *Journal of Iron Steel Institute, 148,* No. 2, 280 (1943).
3. Sims, C. E., and Moore, G. A., "Apparatus for the Hot-Extraction Analysis for Hydrogen in Steel", *A.I.M.E. Transactions, 176,* 248 (1948).
4. Carney, D. J., Chipman, J., and Grant, N. J., "The Tin-Fusion Method for the Determination of Hydrogen in Steel", *A.I.M.E. Transactions, 188,* 397-403 (1950).
5. Epstein, H., Chipman, J., and Grant, N. J., "Hydrogen in Steel Making", *Journal of Metals, 9,* 597-608 (1957).
6. Naughton, J. J., "Determination of Hydrogen by Vacuum Extraction and Tin Fusion", *A.S.M.E. Transactions, 162,* 385-90 (1945).
7. Derge, G., Peifer, W., and Richards, J. H., "The Sampling and Determination of Steel for Hydrogen", *A.I.M.E. Transactions, 176,* 219-47 (1948).
8. Carney, D. J., Chipman, J., and Grant, N. J., "The Sampling and Analysis of Liquid Steel for Hydrogen", *Journal of Metals, 2,* 404-13 (1950).
9. Carson, C. C., "Determination of Hydrogen in Steel. Comparison of Tin-Fusion and Hot-Extraction Techniques". Paper presented at 134th A.C.S. Meeting, and to be published in *Analytical Chemistry.*
10. Shields, B. M., Chipman, J., and Grant, N. J., "Thermal Conductivity Method for Analysis of Hydrogen in Steel", *Journal of Metals, 5,* 180-4 (1953).
11. Murray, W. M., and Ashley, S. E. Q., "Determination of Carbon by the Low-Pressure Combustion Method", *Analytical Chemistry, 16,* 242-8 (1944).

TEMPERING MOLDING SAND

By H. W. Dietert, V. Rowell and A. L. Graham

ABSTRACT

Test procedures for actual and compensated moisture tests are described. The actual moisture test is recommended as standard whenever the actual total percentage of the base water and water of evaporation is required. The compensated moisture test is recommended for determining the sand moisture content without including the evaporation water. This test is recommended for moisture control at the sand mill where sand temperature fluctuates, and one desires to determine the sand moisture at the time it reaches the molding station.

A method of determining the riddled density of sand is given, and its relationship to moisture percentage shown.

The use of high-strength sands is on the increase. Tempering a high-strength sand, by holding ultimate deformation the same as for previously used lower-strength sand, yields a strong sand that retains the good flowability of lower-strength sand. The higher-strength sand will produce a better finish casting with greater contour accuracy.

Ultimate deformation is proposed as a new method of controlling the sand temper in place of using moisture percentage as the index for western bentonite, clay and natural bonded sands.

INTRODUCTION

Molding sands are tempered with water and oil, and may also be tempered with silicone liquid. The latter is not in practical use at this time due to cost. Oil is finding some practical use for tempering treated bentonite bonded sands.

Water, at the present time, is the usual means of plasticizing the bonding material in molding sands. Water softens the clay or bentonite particles, causes the particles to increase in size in many cases and enables the bonding particle to acquire a degree of tackiness and plasticity.

This study is devoted to a means for determining the amount of water suitable for molding. The temper point desired will be predicated on the degree of plasticity imparted by the water addition. Plasticity degree will be measured by ultimate green deformation of the sand.

Water added to molding sand may be divided into two parts—base water, and evaporation water.

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Base water is necessary to plasticize the bond, and is almost wholly absorbed by the bond in the sand. A small portion is used to moisten the surfaces of the sand grains.

SAND MOISTURE

Evaporation water is that portion which is added beyond the amount needed to temper a sand, and comprises moisture lost before the sand is used for molding. The amount of evaporation water required varies greatly. The greatest requirement factor is the temperature of the sand. For example, as much as 0.8 lb of evaporation water per 1000 lb of sand may be required for some sand systems for each 10 degree rise in sand temperature up to 210 F. Above that temperature a multiple of the stated amount may be required.

The amount of evaporation water required is also increased as the atmospheric humidity decreases, and as the air current or the aeration increases. Thus, when moisture determinations are made of molding sands, particularly of sands above room temperature, attention should be given to the fact that water in molding sand consists of two parts, base and evaporation water.

At the time of mixing the sand, one is interested in preparing a sand that will reach the molding station with the correct degree of temper. To accomplish this, sufficient evaporation water must be added at the time of mixing. Correctly understood, moisture tests are a great aid.

MOISTURE DETERMINATION METHODS

Water in a sand is expressed as percentage of moisture. Moisture percentage values give foundrymen a numerical value for control purposes only. It does not express any particular mechanical property of the sand in reference to temper, but does give information as to one of the ingredients of the sand. At the present time, it is the almost exclusive means of controlling the sand temper. Thus, one should be well versed in the test procedure, and what the test values obtained really mean.

There are two fundamental test procedures, actual and compensated moisture tests. These two tests come about due to the fact that as the sand temperature increases in a sand system additional water is added to the mix to compensate for evaporation moisture loss while the sand is transported and stored.

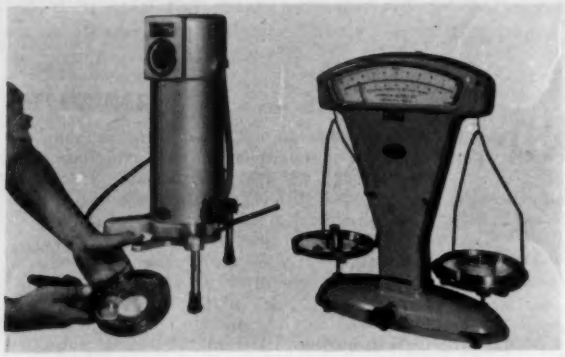


Fig. 1 — Arrangement of equipment for actual moisture test.

The actual moisture test procedure gives moisture percentage of the base and evaporation water present at time of sampling. This test is used whenever one wishes to know the sand moisture.

The compensated moisture test procedure yields results that express the percentage of base moisture in the sand. It allows the evaporation water to vaporize during the first weighing of the compensated moisture test. Thus, this test can be used at the mixing station to ascertain the sand temper when it arrives at the molding stations after it is transported, stored and ready to go into the mold. The hotter the sand, and the dryer the air, the greater will be the moisture evaporated during the first weighing operation. This also applies to the sand in the system.

It is not difficult for one to adjust the evaporation period during the first weighing operation in the compensated moisture test procedure so the test values will express sand moisture as delivered to the molder. Compensated moisture test values for a controlled sand will remain the same throughout the day as the sand temperature varies. Thus, the compensated moisture test is, at the present time, the proper moisture control tool whether one uses manual or automatic tempering.

The actual moisture test is in reality a new moisture test used to check how well the final moisture of the sand was anticipated.

SAND SAMPLING

The representative sand sample is placed in a sand-sample container approximately one quart in size. The container is immediately closed after sample is placed in it with a tight fitting cover to avoid evaporation moisture loss. Make the moisture determination of the sand sample as quickly as possible. Discard top ½-in. layer of the sample. Then quickly mix sand sample by stirring with a spatula. Immediately fill sand weighing dish 2 in. x 1 in. full of sand. Use thumb to slightly pack sand in the weighing dish. Place cover on the weighing dish. The weighing operations are different for the actual and compensated moisture tests.

Actual Moisture Test Procedure

The actual moisture test, which yields test values expressing the exact total moisture in the sand at the time of sampling, is conducted in two slightly different sample treatments during weighing—a standard actual moisture test and a control actual moisture test.

The standard actual moisture test is used for precise work. The closed weighing dish containing the test sample is placed on an analytical or similar balance and a catch weight is obtained. A tare weight is used to counter balance the moisture teller drying pan with filter paper and weighing dish with cover at room temperature. The pan, dish and cover must be stored in a desiccator to keep them moisture free before use.

Figure 1 illustrates, in the right view, how the weighing dish with cover is placed in the drying pan during the first weighing operation. The left view illustrates how the catch weight of the sand test sample is dumped in the moisture teller drying pan together with the emptied weighing dish and cover. This assembly is placed in a moisture teller or oven for drying.

After the sample is dried, cool in a desiccator and reweigh the drying pan with sand, dish and cover. The difference between the first catch weight and the dry weight, divided by the original weight of the moist sand sample and multiplied by 100 is the standard actual moisture percentage.

For control work, one can dispense with the cover on the weighing dish and cooling in desiccator. The amount of moisture loss due to an open weighing dish during the weighing operation is negligible due to the small amount of sand area exposed to the air. The operator also removes most of the top layer of the sand during the weighing operation which further reduces moisture loss during the weighing operation.

Thus, in a control actual moisture test procedure, it is permissible to leave the weighing dish open to facilitate weighing.

Tare the moisture teller pan with filter paper in the bottom and with the weighing dish without cover. Heat the pan and dish in the moisture teller and tare while hot. The use of filter paper in the drying pan does not retard the drying speed. It keeps the pan clean, and the filter paper can be repeatedly used if care is taken to shake it clean after each test.

Place the weighing dish containing the test sample in a heated drying pan with filter paper on the right pan of a balance, and balance to 50 grams of sand.

Dump the sand into the drying pan and leave weighing dish in the pan. Place in moisture teller or oven to dry. Having the moisture teller hot speeds the drying time so that a 3 min drying time as a rule is sufficient.

Re-weigh the dried sample by moving the rider of the lecture balance, and multiply the reading (weight loss in grams) by two to obtain the percentage of moisture. When a speed balance (Fig. 1) is used, the lower scale is read direct as percentage moisture to save time.

For sands above room temperature, one should use the actual moisture test procedure whenever total moisture reading is desired. This procedure is also advisable for moisture tests where accuracy is paramount, and sand is at or near room temperature, pro-

viding one wishes actual moisture content at the time of sand sampling.

Actual moisture percentage values express the total base and evaporation moisture contained in a sand at the time of sampling. It does not yield results that are easily applied in the foundry for tempering sand at the mixing station particularly where the sand temperature varies. Under such conditions, the actual moisture test readings would increase as the sand temperature increases, since a greater quantity of evaporation water is added at the mixer as the sand temperature increases.

It is much easier to use compensated moisture test readings that remain constant throughout the day when the sand is correctly tempered, as sand temperature varies. Thus, the compensated moisture test is of greater value for moisture control work in the foundry at the mixing station.

Compensated Moisture Test Method

Practically everyone is using the compensated moisture test method without giving it sufficient thought to yield accurate results. A better understanding of the compensated moisture test is necessary since it is a useful control tool. It is surprising how easy it is to apply correctly when care is exercised.

The sand sample is taken at the sand mill discharge and placed in a covered container. The simple operations are illustrated in Fig. 2. Quickly discard the top layer of warm or hot sand from container, mix by stirring with spatula and fill the weighing dish full of sand. Dump the sand from weighing dish into the drying pan with filter paper in bottom. Spread sand uniformly in the drying pan rapidly and immediately start weighing. The weighing dish places a little more than 50 grams of sand in the drying pan. Start a stopwatch and remove sufficient sand with a spatula to balance at 50 grams of sand when 60 sec time is reached.

During this first weighing operation, remove sand from the sand sample at one edge only. It is essential to present a constant sand sample contour to the atmosphere to obtain accurate results. Place the weighed drying pan containing the 50 grams of sand in the moisture teller and dry. Re-weigh and read moisture percentage from moisture percentage scale on the speed balance, or multiply rider reading by two when a lecture balance is used.

At the beginning of the test, use a hot pan and place a 5¼-in. diameter filter paper in the bottom of the drying pan. Thus, both weighing operations are made with hot pans so as to cancel out the greater part of the error of weighing with hot pans. The filter paper allows one to use a hot pan and to keep the pan clean. After each test, remove the filter paper and shake it clean for re-use.

USE OF MOISTURE TESTS

The 60 sec weighing time suggested in the compensated moisture test may not fit all sand systems. To find the correct weighing time, make an actual moisture test of the sand as it comes out of the molder's hopper. Then make a compensated moisture test of sand as it comes from sand mill discharge. Compare



Fig. 2 — Arrangement of equipment for compensated moisture test.

the actual moisture test reading at the molding stations with the compensated moisture test reading at the mixing station; both should be the same.

Alter the weighing time in the compensated moisture test to allow sufficient evaporation of moisture to yield compensated moisture test readings at the sand mill, which agree with the actual moisture test readings at the molder's hopper discharge.

Batches of sand which are discharged before being thoroughly mixed will cause erratic moisture percentage readings. Large variations in moisture percentages may be experienced without any exact pattern of error under such inadequate mixing procedure. Uniform moisture distribution must first be secured in the mill before one can apply strict moisture control.

Compensated and Actual Moisture Test Readings Comparison

Additional information on the subject is in the test data in Table 1. Here, for comparative purposes both the compensated and actual moisture tests were made on the same sand at the mixer discharge. Note that as sand temperature increases, the difference between the compensated and actual test readings increases. This illustrates that as sand temperature increases additional water is added to compensate for evaporation.

TABLE 1 – TEST DATA FOR ACTUAL AND COMPENSATED MOISTURE TESTS AT THE MIXER DISCHARGE AS SAND TEMPERATURE VARIES

Sand Temp., F	Compensated Moisture, %	Actual Moisture, %	Difference, %
66	4.95	4.87	0.08
76	5.1	5.04	0.06
88	4.95	5.03	0.08
93	5.0	5.14	0.14
100	4.9	5.1	0.20
105	4.82	5.08	0.26
107	4.8	5.07	0.27
108	4.95	5.3	0.35
116	4.75	5.22	0.47
126	5.1	5.65	0.56
129	5.0	5.86	0.86



Fig. 3 — Equipment for determining riddle density.

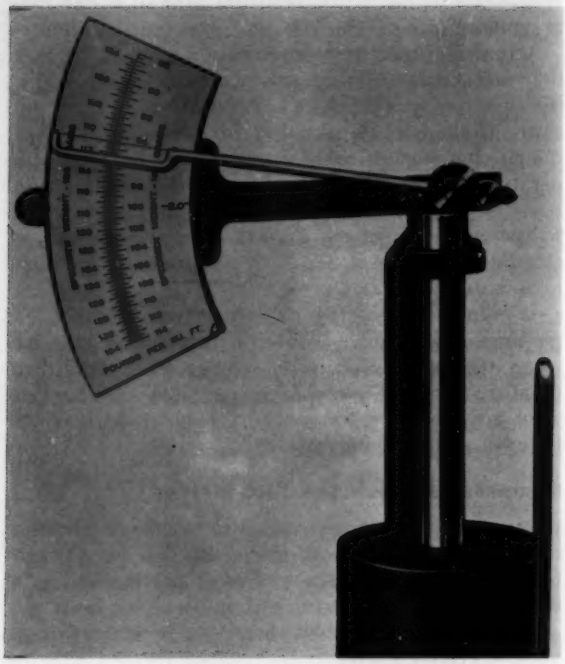


Fig. 5 — AFS Sand rammer with rammed density indicator.

Riddled Density vs. Moisture

Many prefer sand tempered so it is lightest in "feel." Some may not have given much credit to this method. The lightness of sand as affected by moisture, can be easily measured by determining the riddled density of a sand as moisture content is varied. The sand is riddled from a constant height into a container, with equipment shown in Fig. 3.

The sand container is filled with riddled sand and struck off level. The weight of the sand vs. moisture percentage yields a graph, as shown in Fig. 4. The weight of the sand is at a minimum in the temper moisture-percentage range.

The graph does not come to a sharp point at the minimum, so the precise moisture at temper is not shown. However, a useful moisture range yielding the lightest sand is shown.

An alternative method is to use an American Foundrymen's Society sand rammer with density indicator, Fig. 5. Ram the sand with three drops using a constant weight of sand, and vary the moisture. Read the rammed density from the density indicator. The object is not to obtain a sand with minimum density per se, but to moisten the clay substance which swells it to a good workability.

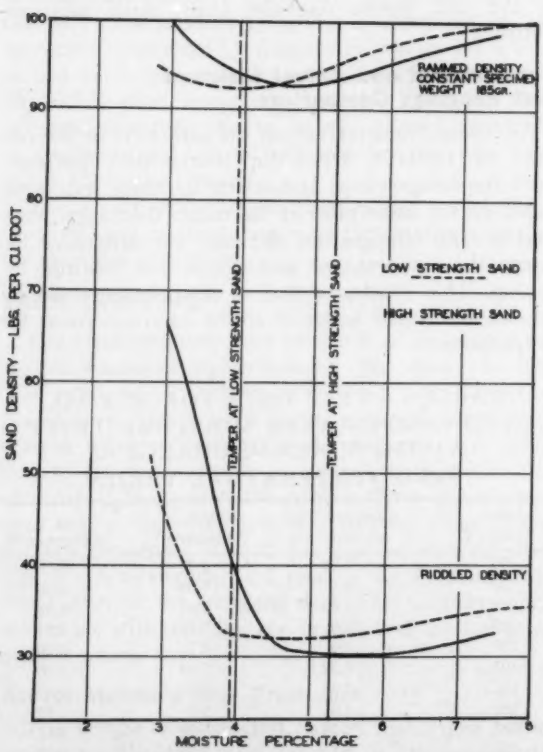


Fig. 4 — Riddled and rammed density vs. moisture content.

Ultimate Deformation as a Temper Control

Ultimate deformation measures the amount the sand specimen shortened at the time of failure under compressive load. Precisely, ultimate deformation is read at the point when the rate of specimen shortening exceeds the rate of load. The rupturing of the sand specimen is actually stretching the plastic bond in the sand to the breaking point. Moisture increases the stretching ability of the clay bond. Thus the ultimate deformation test reading is an index of the degree of wetting of the bond.

The ultimate deformation of a western bentonite bonded sand is linearly affected by the moisture content, as illustrated in Fig. 6.

The higher the moisture percentage, the greater the ultimate deformation. To make use of this rela-

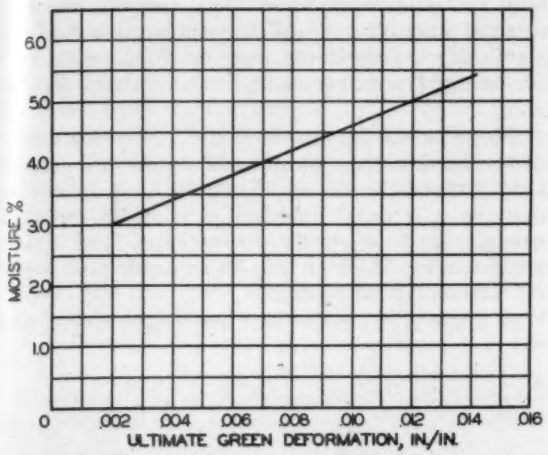


Fig. 6 — Relationship between ultimate deformation and moisture percentage.

tionship, it is necessary to determine the ultimate deformation as moisture is varied for a sand. Then temper the sand to the desired point found better in practice, and at this temper determine the ultimate deformation of the sand. For many sands other than steel, an ultimate deformation range between 0.016 in. - 0.020 in. is found ideal. Some southern bentonite bonded sands may perform better at a lower ultimate deformation, and deformation-moisture control may not be sensitive.

Thereafter, sufficient water is added to each batch of sand to cause sand to possess the predetermined ultimate deformation. Thus, the molders will receive a sand that has constant plasticity or workability, not a sand necessarily of constant moisture content.

Such a sand may actually form a mold of more constant properties than a constant moisture sand. A constant moisture sand, when a change in fineness or bond occurs, need not be at temper. Moisture content should be changed under such conditions to produce an ideally tempered sand of constant workability.

When ultimate deformation is the control index, then moisture content is changed to hold the plasticity of the sand constant. There are a few bonds, southern bentonite for example, where additional research work must be done to apply the ultimate deformation as a temper control.

Water, per se, is not a functional or a useful property that influences the workability of a sand. Thus, it may be found useful to use ultimate deformation as the temper control point for some sand. A sand in a system may not remain constant in fineness, bond, facing or cellulose material content, etc. As these items vary, and the sand is tempered to a constant ultimate deformation, the molders may not notice as much difference in the sand as compared to sand that is tempered to a constant percentage of moisture.

TEMPERING TO ULTIMATE DEFORMATION

The usefulness of tempering sand to a predetermined ultimate deformation reading can be demonstrated by tempering both a normally low-strength

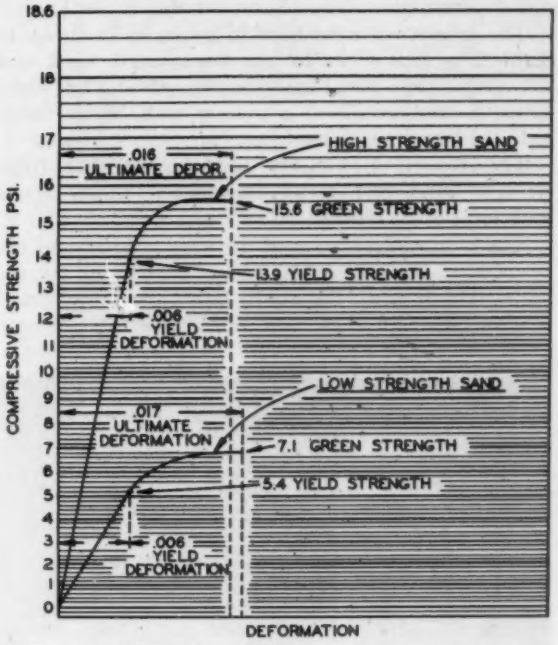


Fig. 7 — Stress-strain curves for 7.1 psi and 15.6 psi strength sands.

sand, for example 7.1 psi, and a high-strength sand, for example 15.6 psi, to good workability.

Assume that a low-strength sand was in use, and workability was found satisfactory. A high-strength sand was desired to obtain better casting contour accuracy and still retain the good workability of the low-strength sand. There are two ways available to temper the high-strength sand, one of which would be to add sufficient moisture by trial and error method to secure good flowability and rammability.

However, a more direct avenue is open. Add sufficient water to the high-strength sand mix to produce an ultimate deformation that is the same as that of the good low-strength molding sand. Using this procedure, one secures stress-strain curves, as shown in Fig. 7. The ultimate deformation of the low-strength sand is 0.017 in., and for the high strength sand it is 0.016 in. The actual moisture percentage of the low-strength sand is 3.95 per cent, and for the high-strength sand is found to be 5.2 per cent. Thus, the correct temper for the high strength sand at room temperature is 5.2 per cent actual moisture, and at elevated temperature it is 5.2 per cent compensated moisture.

Since good flowability is essential in the production of high quality castings, an examination of the flowability is in order for these reference sands. Figure 8 shows both the low and high-strength Rowell flowability specimens of the low- and high-strength sands. The specimens are similar in appearance, showing no excessive false voids.

Test castings were made in type metal with both of the sands, shown in Fig. 9. The castings made in the high-strength sand show less metal penetration than castings made in the low-strength sands. The strength of a sand does affect in some manner the degree of

metal penetration. The reason for this is not known.

The amount of sand burn-in in grams is shown in Table 2 as 28.8 grams for the low-strength sand, and 1.2 grams for the high-strength sand. The high-strength sand produced a casting with better finish than the low-strength sand.

The Rowell flowability test showed that high-strength sand had better flowability, which agrees with test casting results.

The creep deformation of the low-strength sand under a 5 psi load is 0.0053 in., and 0.00187 in. for the high-strength sand. Figure 10 shows the creep deformation curve for these two sands. The high-strength sand has only one-third as much creep as the low-strength sand. This means that molds made from the high-strength sand would have much less mold wall movement, producing castings more accurate in contour.

Therefore, castings made in the high-strength sand will weigh less than those made in the low-strength sand. Referring to the test data in Table 2, it is found that the type metal test castings made in the high-strength sand weigh on the average 778.8 grams, while castings made in the low-strength sand weigh on the average 789.4 grams.

Sand specimens made with the high-strength sand

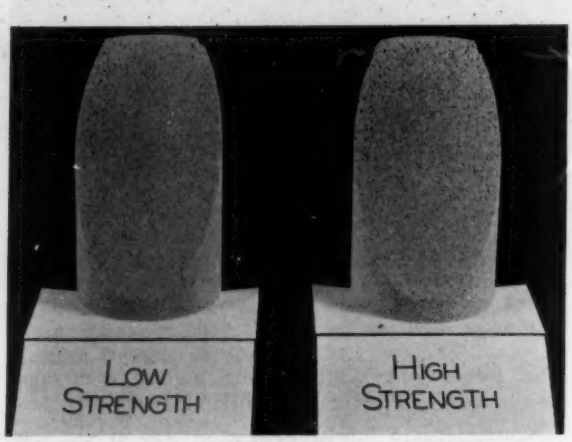


Fig. 8 — Visual flowability of the 7.1 psi and 15.6 psi strength sands.

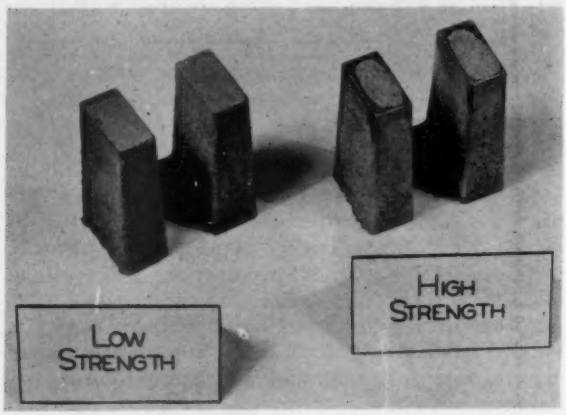


Fig. 9 — Type metal test castings of the 7.1 psi and 15.6 psi strength sands.

TABLE 2 — SAND PROPERTIES FOR 7.1 PSI AND 15.6 PSI STRENGTH SANDS TEMPERED TO SAME ULTIMATE DEFORMATION

Property	Low Strength Sand	High Strength Sand
Mix: Bank Sand %	93.0	89.0
Western bentonite, %	7.0	11.0
Moisture, %	3.95	5.2
Rammed density, lb/cu ft	95.55	92.7
Riddled density, grams/qt.	598.10	579.5
2 in. x 2 in. specimen weight, grams	153.0	154.1
Green permeability	88.6	85.8
Green compression, psi	7.1	15.6
Ultimate green Deformation, in./in.	0.017	0.016
Yield strength, psi	5.3	13.9
Yield deformation, in./in.	0.0065	0.0065
Mold hardness, top	88	91
Mold hardness, bottom	87	90
Indicator flowability, %	82	70
Rowell flowability, %	84	87
Dry compression, psi	148.7	145.0
Air set strength, psi	30.2	32.5
Creep deformation, 5 psi load, 1 min, in./in.	0.00530	0.00187
Expansion, 1 psi load, 1800 F, 22.5 sec, in./in.	0.0067	0.0054
Restrained load, 1800 F, 30 sec, psi	68	59
Hot strength at: 500 F, psi	174	190
1000 F, psi	223	236
1500 F, psi	547	677
2000 F, psi	198	268
2500 F, psi	4	4
Hot deformation at: 500 F, in./in.	0.0085	0.0098
1000 F, in./in.	0.0092	0.0126
1500 F, in./in.	0.0148	0.0190
2000 F, in./in.	0.0983	0.1198
2500 F, in./in.	Plastic	Plastic
Sand burn-in weight, gram	28.8	1.2
Casting weight, gram	789.4	778.8

showed a top mold hardness of 91 as compared to 88 for the low-strength sand. This illustrates the fact that high mold hardness may be obtained with high-strength sands when they are correctly tempered by the ultimate deformation method.

The dry strength and the air-set strength properties are similar for the low- and high-strength sands when they are tempered to the same ultimate deformation, as shown in Table 2.

Mold wall fracture casting losses, such as scabs, rattails and buckles, are directly related to the restraining load and the 1 psi load expansion test at 1800 F.* The high-strength sand has a lower restraining load and a lower 1 psi load expansion than the low strength sand, indicating that one should experience less scabbing, etc., with the high-strength sand. However, both of these test values are too high for a safe sand, thus scabbing losses may be expected with either sand. The addition of a carbonaceous, cellulose or cereal material will lower the restraining load below the 50 psi danger limit for either sand.

The hot strengths of the low and high-strength sands are identical for all practical purposes and are conservative. The high-strength sand has a slightly better hot deformation, which agrees with the slightly better restraining load and 1 psi load expansion test results.

*AFS Transactions, *vol. 65*, 1957, pp. 128-133, and 235-237.

CONCLUSIONS

1. When tempering sand, water must be added for base water and evaporation water requirements. The base water plasticizes the bond and moistens the sand grains. Evaporation water is lost during the sand handling.
2. The actual moisture test procedure should be used whenever the total percentage of moisture in a sand is desired, as it measures both the base water and evaporation water.
3. The compensated moisture test procedure, which yields the moisture percentage consisting largely of only base water, is used for moisture control work at the mixing station.
4. The actual moisture test procedure is used when determining the moisture of the sand at the molding station, in research work or when determining the total water added to a sand.
5. The compensated moisture test procedure is used when determining the moisture at the mixing station. It is the moisture control test that anticipates the moisture of a sand after it arrives at the molding station.
6. Test procedures of the actual and compensated moisture test must be followed carefully to obtain reliable results.
7. Test equipment used in moisture tests must be maintained in excellent working order.
8. Uniformity of sand is essential in applying moisture control by the moisture testing methods.
9. The old time expression of tempering a sand to "feel" lightest is a technical fact.
10. Tempering sand to a constant compensated moisture content is in everyday use. However, it need not be considered the ultimate in sand control. Constant compensated moisture control does not yield the better tempered sand in reference to workability when the sand in the system varies in bond, fineness or in other ingredients. Such sands may have a better workability if sufficient water was added to cause a constant ultimate deformation.
11. A high-strength sand may be tempered to the better workability by using ultimate deformation as the means of determining temper point.
12. A high-strength sand controlled at a constant ultimate deformation will produce castings of better contour, less weight, less penetration and less mold wall fracture losses than a low-strength sand, and with equal molding ease.

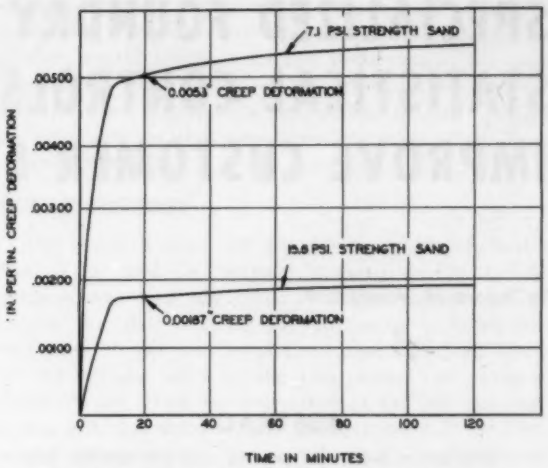


Fig. 10 — Creep deformation curves of 7.1 psi and 15.6 psi strength sands.

SPECIALIZED FOUNDRY STATISTICAL CONTROLS IMPROVE CUSTOMER SATISFACTION

By Buford M. Appleman

ABSTRACT

The major statistical process and job quality controls which have been developed especially for foundry applications at the author's company are described. The $\overline{X}$ and R Charts are not discussed in this paper because many previous articles have been devoted to them.

Techniques presented herein are based on sound statistical relationships, and wherever possible they are applied in simplified "cookbook" form to facilitate line usage right in the foundry by foundry personnel.

Customer satisfaction is, of course, the main objective. However, in order to achieve this goal operational economics must be practical. For this reason the management of the quality control activity is critical. Special techniques are presented which simplify much of the judgment as to the location and amount of effort expended.

The development and installation of this total quality control program has been instrumental in reducing total hard and soft iron scrap from a 27-month average of 10.9 per cent, to a full year of experience at 5.4 per cent with still further gains anticipated. Based on $150 per ton average hard and soft iron scrap, this savings has been worth some $80,000 during the past year. Incidentally, this was done in a jobbing shop, and surprisingly was accomplished by lower quality control man-hours per ton and reduced customer returns.

INTRODUCTION

Manufacturing plants previously accepted the cross of Inspection departments, but with the advent of mass-production machines and munitions manufacturing this activity was re-evaluated. It was not feasible to blow up each shell to see if it was all right before it left the assembly plant. However, in mass production 100 per cent inspection activity sometimes costs as much as the production of the items itself. At this point sampling plans came into vogue along with $\overline{X}$ and R Charts. Both techniques were designed primarily for mass production and machine shops. Naturally, other industries picked up these techniques and adopted them with varying degrees of success. When this was done, they promptly felt that they had installed statistical quality control programs.

In the foundry business these sampling plans and $\overline{X}$ and R Charts are good tools but they leave much to be desired. They do not answer the question of which jobs or processes should be controlled, where to control averages or how much variability is to be allowed. Foundries desire to manufacture quality into the product, which is the basis for the $\overline{X}$ and R concept. However, this is much easier said than done. The gray and malleable iron foundries find this to be particularly true because the casting is either good or bad when it comes out of the sand, and it is too late for much salvage action. It can either be shipped or scrapped. Steel foundries can salvage castings by welding, but this, too, costs money.

Foundry problems are tough and they are a challenge – perhaps that is why so many keep fighting – they just hate to give up. Seriously, foundry problems are just as difficult as those encountered by the petroleum and chemical plants. The main difference is that most foundry units are fractional in size as compared to petrochemical plants, which can afford large engineering staffs and have been progressive in developing special statistical quality control techniques and experimental designs. Perhaps foundries can not go as far as the process industries have gone, but it is possible to incorporate successfully many of their techniques.

The American Society for Quality Control is doing much to foster this type of work, which can be of invaluable assistance to foundries through local chapters and special groups.

CASTING BUYERS WANT IMPROVED QUALITY

Foundrymen realize and feel the pressures for upgrading quality being brought upon them by casting buyers. What are the reasons for this? An analysis of factors which contribute to customer satisfaction includes the following:

The lowest per cent defective is highly desirable. Machine tools are expensive, and are of a higher production type than before. Skilled and semi-skilled labor is costing more money all the time. Tooling and setup costs are more expensive on these more complex machines. Inspection is more critical and assemblies more complex.

Closer tolerances are desired. Less machining is required if casting tolerances are held more closely. This results in a better fit of rough castings in assemblies, and also simplifies the assembly itself. Less total movement in chucks of highly mechanized machinery allows less variability of raw castings than the

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three and four jaw hand-tightened chucks of the past.

Most exacting quantities of delivery keep inventories in balance. Money tied up in inventories is expensive for both the foundry and customer. To maintain a minimum inventory requires more exact predictions of per cent scrap at the foundry or the foundry is forced to ship short and disappoint the customer, or hold the overage of customer's castings in inventory until receipt of additional orders as the past few years have brought evidence of customer reluctance to accept the usually permissible 10 per cent over or under shipment.

Shorter deliveries reduce inventories. Shorter deliveries also reduce investment but these shorter delivery commitments do not allow time for trial-and-error production.

These factors are wanted at lower casting costs to the casting buyer. If it were possible to snap one's fingers and improve quality levels and produce more consistent castings, it would be done because customer satisfaction would increase and foundry costs would decrease.

Many advantages accrue to the foundry with better quality control. Better quality control offers many intangible advantages to the foundry. High quality castings will result in sale of more tons by virtue of their perfection alone. Salesmen are more efficient when they have fewer complaints, and are received more readily in customers' engineering, manufacturing and purchasing departments.

The tangible advantages in better quality control are many. They include lower inspection costs on the final product by acceptance sampling of more lots. There is less rework and salvage as well as fewer returns from the customer. Lower charge-backs for loss of machine time and tools or inspection are a direct result. Lower scrap is significant in calculating the value of better quality control. Fewer customer complaints and service calls are necessary when such a condition does exist, because 100 per cent inspection is seldom as efficient as manufacturing quality into the product.

There is an optimum amount of money which can be spent to achieve better control. The cost of such a program must be repaid by savings accrued through the program. Money can be invested only until the economical balance of total actual dollars of scrap, salvage, charge-backs, returns and inspection labor has reached the optimum from which any change would result in higher total cost of all quality factors. This approach to optimum is the real contribution quality control makes toward quality, delivery, cost or the total financial gain of the business enterprise.

SCRAP HISTORY

The scrap history for 2¼ years in the Malleable Div. of the author's company is shown in Fig. 1. This average was 10.9 per cent. A dollar value was calculated for this scrap at the minimum variable cost rate of $150 per ton total hard and soft iron scrap. At the volume level during this period the value of one per cent scrap was estimated at $15,000 per year. It was felt that some of this total scrap could be eliminated economically. How much could be eliminated and at what cost?

Analysis of this scrap indicated that 15 per cent of the jobs resulted in two-thirds of the total quality loss (Fig. 2). This supported the Pareto principle. Pareto was an Italian economist who discovered that approximately 2 or 3 per cent of the Italian population controlled over 95 per cent of the wealth of the country. In applying the same principle to the amount of crops produced from certain farm areas, he found the same condition existed in that a few acres usually produced most of the crop. There were many other applications in which the same principle applied. His theory was that only a few factors of a large number contribute most of the goodness or badness of the product.

Next it was assumed that a reduction of the scrap losses in the 15 per cent of the jobs producing most of the scrap could be made. Such reduction, estimated at 50 per cent, would result in at least 25 per cent reduction of the total quality loss, and it was believed attainable within one year. This means a goal of 8.2 per cent which would result in savings in excess of $40,000.

Top management recognized that such a program indicated a favorable return on its investment. This factor, plus the awareness of general quality upgrading by casting buyers, resulted in 100 per cent backing of the program by top management. Such an attitude must be vividly apparent to all production personnel if a program of this nature is to succeed.

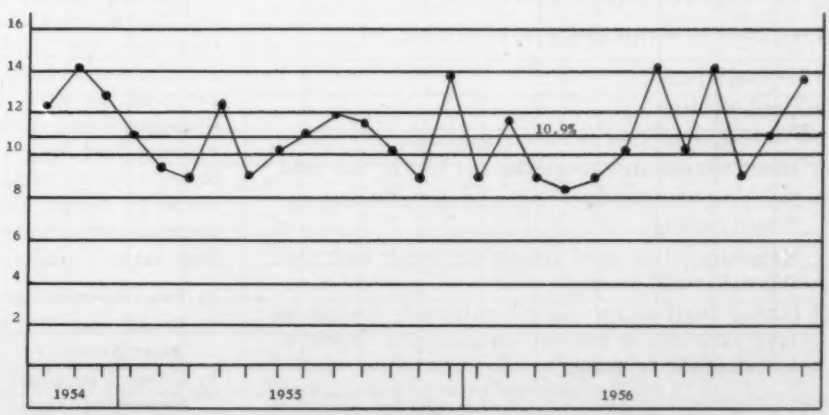


Fig. 1 — Total malleable division hard and soft iron scrap, October, 1954 through December, 1956.

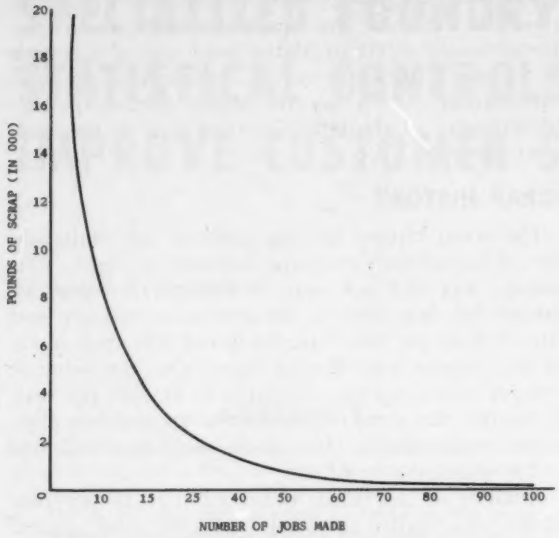


Fig. 2 — The Pareto curve indicates that a small percentage of the jobs run result in a large percentage of the total scrap. Graph shows weekly average malleable scrap, scrap pounds vs. number of jobs made.

THE PROBLEM

Foundries have a severe quality problem due to the fact that a large transformation takes place from melting of raw materials through the liquid pouring of metal into green sand molds around dry sand cores to solidification in the molds shaped by a pattern.

The problem at the author's company was basically this. A job would run two or three times quite well, and then the next time the scrap loss would be heavy. Sometimes this would not show up until after anneal, which is 10 days or so after the job is made in the molding sand. Another phase of this problem is that a defect might be caused by several things. A swell or an enlarged portion of the casting might be caused by:

1) A soft mold due to insufficient jolting or squeezing.
2) A ram-away caused by the machine jolting off center or the pattern being offset.
3) Bad flask due to warpage, etc.
4) Bad jacket or jacket set on the mold incorrectly.
5) Weak sand lacking in green or hot strength.
6) Partially burned out bottom boards.

A hot crack or shrink might be caused by:

1) Soft molding.
2) Hard molding.
3) Pouring at too slow or too fast a speed.
4) Metal temperature might be too hot or too cold.
5) Dumping the mold too hot and not allowing sufficient cooling.
6) Metallurgical or sand conditions being such that this defect will occur.
7) Gating itself might not be quite safe enough to take care of the normal variation for which allowance must be made.

Few of the causes of defects can be analyzed accurately after the castings are made, poured, dumped, cooled, sprued and cleaned, because all of the evidence is gone. If the quantity is small, and the job ordered only periodically such as two or three times per year, the problem is exaggerated.

THE QUALITY IMPROVEMENT PROGRAM

A self-analysis was made to determine the basic problems. Because of the difficulties, as pointed out in previous paragraphs, it was found that often the pattern itself was changed in order to produce better castings. Sometimes this was done when the same pattern had been used previously in a production run of high quality castings. Obviously, this was being done in many cases because of the variability of some other phase of the manufacture, and should not have been necessary in the first place. Of all the factors which can change from one run to the next, the pattern should change the least.

By the the same token, it was known that purposeful changes in carbon, tap-out temperature, moisture content, green strengths, etc., had been made in the shop with the belief that such action would cure some defect which seemed to be rampant. Sometimes many changes were made almost simultaneously. If an improvement did occur, the real reason for it was not always known. It might have been some factor not considered at all, which was varying or that particular group of jobs showing improvement, might have been different from those which suggested the need for a change.

BASIC CORRECTIVE STEPS

Consistency was sought by resisting changes in basic operational methods without careful analysis of the defects and their real importance from a dollar and quality standpoint.

Process control was brought in, and at the same time a consulting engineering firm was retained to assist in the design and evaluation of statistical experiments and training of foundry personnel in these techniques. The largest of these experiments was designed to determine which of seven elements each of sand and metal, plus pouring time and dumping time, were critical from a quality standpoint. Emphasis was placed on solving hot cracking and shrinking tendencies.

This experiment included five patterns representing various types of jobs run at this shop. Some of these were the "stinkers" which was the name given those jobs producing two-thirds of the scrap. This experiment was known as project RAFE (Random Assignment Factorial Experiment). Figure 3 shows the experimental design; the plus (+) and minus (−) indicate levels of each. This experiment took over two weeks to design, two weeks to run and over four weeks to evaluate its major findings. As a result only four major changes were recommended.

1) Pour the metal colder than 2850 F at the tap out.
2) Avoid excessive fines (limit to 5 per cent maximum fines).
3) Control moisture in the sand at a level just sufficient for tempering.

4) Use more than 4.0-5.0 per cent sea coal in the sand.

These were then adopted and operational controls were adjusted to the new balance. $\overline{X}$ and R type of controls are now used to maintain control on some of these critical items, and several other controls were eliminated as being unimportant. It should be emphasized that findings are helpful to the author's company, but this does not necessarily mean that they will help any other malleable shop because of the difference in each foundry of castings run, gating systems, melting practice, sand control, etc.

Tools for Job Control Were Developed

Job control in foundry production does not lend itself to simple single sampling plans, because the

PROJECT RAFE

VARIABLE		A	B	C	D	E	F	G	H	I	J	K	L	M	N	O	P
		PLATE	PIG	OXIDATION	CARBON	SILICON	ALUMINUM	TEMPERATURE	SAND	MOISTURE	SEA COAL	CLAY	MOLD PRESSURE	BENTONITE	CEREAL	POURING RATE	DUMPING TIME	
DAY	RUN																	
1	1	+	+	-	-	-	+	+	-	-	+	+	-	+	+	+	-
1	2	+	+	-	-	-	+	-	-	+	-	-	+	-	-	+	+
1	3	+	+	-	+	-	-	+	+	-	+	+	-	-	-	-	+
1	4	+	+	-	+	+	-	-	-	-	-	+	+	-	+	+	+
2	5	-	+	-	+	+	+	+	-	-	-	-	+	+	+	-	-
2	6	-	+	-	+	-	+	-	+	+	-	-	-	+	-	+	-
2	7	-	+	-	-	-	-	-	-	-	+	-	+	-	-	-	+
2	8	-	+	-	-	+	-	+	+	-	-	+	-	+	+	-	+
3	9	-	-	-	-	+	+	-	-	-	-	+	-	+	-	-	-
3	10	-	-	-	-	-	+	+	-	+	+	-	-	+	-	+	+
3	11	-	-	-	+	-	-	+	-	-	-	+	-	-	-	-	+
3	12	-	-	-	+	+	-	-	+	+	+	+	-	+	+	-	-
4	13	-	+	+	+	+	-	-	+	-	-	+	+	+	-	-	-
4	14	-	+	+	+	-	-	+	+	+	+	-	+	-	-	+	+
4	15	-	+	+	-	-	+	+	+	+	+	-	-	-	+	-	-
4	16	-	+	+	-	+	+	-	-	+	-	+	+	+	-	-	-
5	17	+	-	-	-	-	-	-	+	-	+	-	+	+	+	-	-
5	18	+	-	-	-	+	-	+	+	+	-	+	+	-	+	+	-
5	19	+	-	-	+	+	+	+	-	+	-	+	-	+	+	+	-
5	20	+	-	-	+	-	+	-	+	+	+	+	+	+	-	-	+
6	21	-	-	+	+	-	+	-	-	-	+	-	-	-	+	-	+
6	22	-	-	+	+	+	+	+	+	+	-	+	-	-	-	+	-
6	23	-	-	+	-	+	-	+	-	+	+	-	+	+	+	-	+
6	24	-	-	+	-	-	-	-	-	-	+	+	+	+	-	-	-
7	25	+	-	+	-	+	+	-	+	-	+	-	-	+	-	-	+
7	26	+	-	+	-	-	+	+	+	-	-	-	-	-	+	+	+
7	27	+	-	+	+	-	-	+	+	-	-	-	+	-	-	+	+
7	28	+	-	+	+	+	-	-	-	+	+	+	-	-	-	+	-
8	29	+	+	+	+	+	+	+	+	-	+	+	+	-	+	+	+
8	30	+	+	+	+	-	+	-	-	+	+	+	+	-	+	+	+
8	31	+	+	+	-	-	-	-	-	+	-	-	-	-	+	+	-
8	32	+	+	+	-	+	-	+	+	+	-	-	+	+	+	+	-

Fig. 3 — Project RAFE was designed to determine which of seven elements of metal, plus pouring rate and dumping time made the highest quality castings.

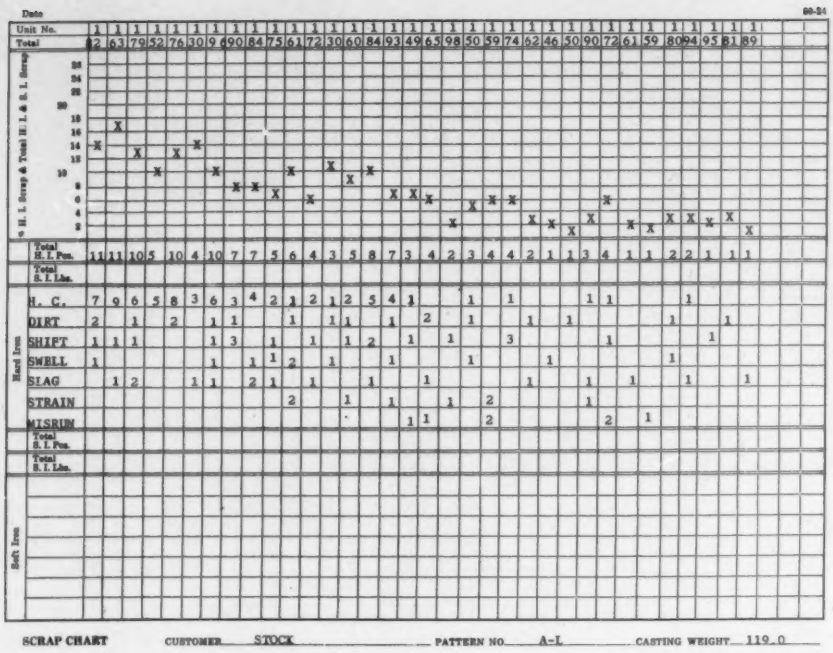


Fig. 4 — Job history of a "stinker."

entire production run is not complete at the time the decision of run or stop must be made. For this reason signals were necessary to decide which jobs would be controlled during production.

"Stinkers" were defined as those jobs having 500 lb or more of scrap in one week. Figure 4 shows the history of a stinker. The original concept of 5 per cent maximum scrap per job was not good. The trouble with this approach was that if a small order was run at 25 per cent scrap, it might take as much quality control and production supervision time for control as would be required to reduce from 5 to 4 per cent a comparatively large production item with much greater total scrap reduction.

One example of this was the permanent mold production of tee-head bolts. Since production was over 50,000 pieces per week, it was worth while to reduce scrap appreciably. It now runs at less than 1½ per cent under normal conditions. This also results in a savings of hard iron and soft iron inspection labor which is very significant.

Sequential Charts

Sequential charts like the one in Fig. 5 were designed for the control of stinkers, and they were set to the wishes of management regarding acceptable and unacceptable limits. If the line enters "Question Continuation of Operation," the odds are that nine times out of ten the job is running worse than 12 per cent. By the same token there is only one chance in 20 that a job going below the lower line with inspection stopped will be worse than 2 per cent.

All stinkers are watched daily by quality control and production supervision. Sequential charts are

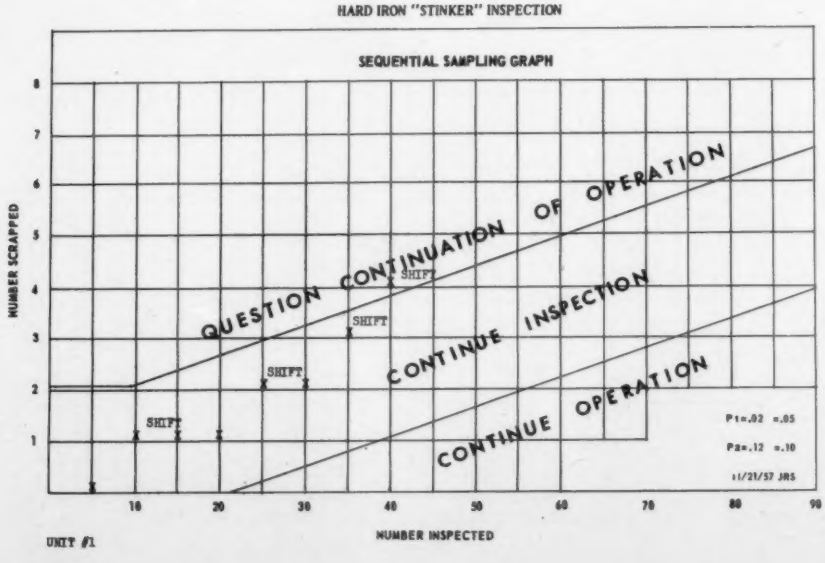


Fig. 5 — "Stinker" sequential chart shows job going out of control.

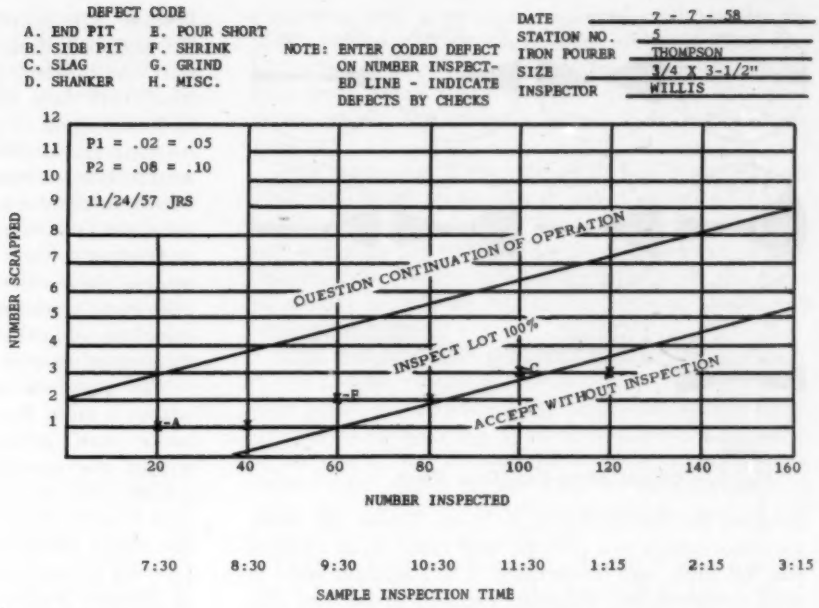


Fig. 6 — Sequential chart on ¾-in. T-Head bolt shown entering "Accept Without Inspection" area.

used to chart job progress each day, and if a stinker is running good enough to enter the "Question Continuation of Inspection" area three production days in a row, even if separated by 6 months between each day of production, it is dropped from the stinker list. If it goes out of control into "Question Continuation of Operation," the job is immediately stopped, or else multi-vari data are accumulated and then it is stopped. If the job stays in the middle of the band, inspection is continued. These charts are revised as total scrap is lowered. The bolt chart is far tighter than the regular stinker chart, as shown in Fig. 6. These charts can be designed for any desired level of acceptance of rejection.

New job control falls into the same category and a new job is guilty of being a stinker until proven innocent.

Reporting Systems

Reporting systems determine the location and magnitude of dollars of quality loss. Table 1 shows the malleable division weekly report. Units, classes, and defects are fairly good, but there are stinkers. Again, the Pareto curve is used in this analysis to point out the major area of defect by dollar loss. The clerical efforts of the entire quality control team are most efficient when tackling the largest item which is out of control.

The reporting system tells whether one of the molding units, certain of the defects by class of defect, a particular class of casting, or a few individual pattern numbers seem to be out of control. The Pareto principle says there will be one of these out of control, and so far there has been. When this is no longer true, then it is obvious that further reductions will be uneconomical.

The scrap goal is not zero and efforts are made to state carefully that the goal is always the economic optimum. This is true because the lower the per cent scrap, the harder it is to push it still lower, and the more effort is required to maintain that level. This is a dynamic system and efforts must be constantly shuffled around to achieve the greatest gain for the least expenditure of time and effort.

Special Problem Solving Techniques

One of the special problem solving techniques used when a job goes out of control on the sequential chart

TABLE 1 — MALLEABLE DIV. HARD IRON SCRAP SUMMARY
AUG. 19 - AUG. 22, 1957

Unit	Scrap, lb	Produced, lb	Scrap & Good, % True
1	3,532	24,985	14
2	735	13,323	6
3 & 4	1,280	33,442	4
5B & 5C	511	21,900	2
5A-7 & 8	5,179	74,181	7
6	6,141	112,688	5
S.F.	650	2,644	25
Samples		14,009	
H.I. break	909		
Total	18,937	280,003	6.8

Class	Scrap, lb		Defect	Scrap, lb	
Chunky	5,059	27%	Hot crack	3,744	20%
Ring	2,895	15	Dirt	3,096	16
Brake shoe	2,793	15	Shift	2,293	12
Pipe fitting	1,677	9	Swell	1,908	11
Hub	1,141	6	Misrun	1,464	8
Spool	970	5	Slag	1,010	5
H.I. break	909	5	H.I. break	909	5
Spoke wheel	696	3	Draw	852	4
Misc.	2,797	15	Pour short	773	4
			Strain	631	3
			Misc.	2,257	12
Total	18,937			18,937	

Hard Iron Stinkers

Customer	Pattern no.	Scrap, lb	Scrap, %
A	606-2453	3,330	18
B	UB 1177	2,899	15
C	2A-167	1,127	6
D	PP 5172-5	734	4
E	1366	696	3
F	UB 1002	639	3

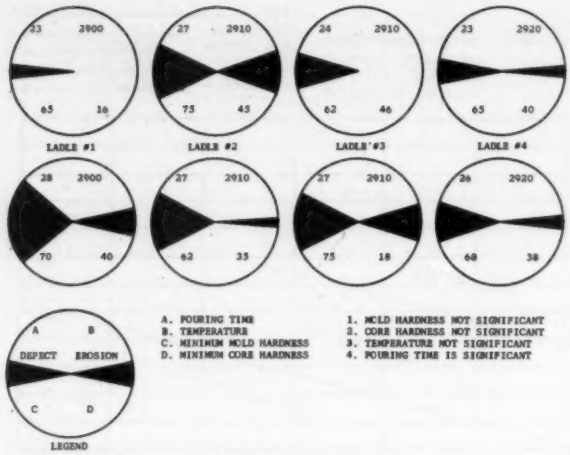


Fig. 7 — Typical Multi-Vari Data Chart.

involves the accumulation of data. There are many variables which can interact and cause scrap castings just by their own variability. The variables such as sand temperature, moisture control, green and hot strength, mold hardness, pouring time, pouring temperature, pouring technique, dumping times, etc., are all constantly changing within certain limits. To assume that each one will stay exactly where it was while making two or three molds is unrealistic. There is still a lot of "art" necessary in this business, but castings do react to certain scientific rules and patterns.

The problem is to determine what the ground rules are for good production of each job and then each process. The accumulation of multi-vari data is for the purpose of determining these rules. Figure 7 is only one of many different types of multi-vari charts. Multi-vari data include numbering within a mold, and each mold is numbered in sequence. All variable information known to be of common importance is accumulated during that time. This includes core hardness, mold hardness, pouring time, pouring temperature, pouring technique, jacket sequence, etc.

Analysis of the type and severity of the defects against the multi-vari data will show a pattern. The time cycle of the magnitude of the defect will answer questions as to the basic problem area. Figure 8 illustrates another type multi-vari chart.

The questions are these. Is it a matter of position within a mold that pattern 1 is consistently better or worse than another? Is there any period of time within the manufacture of the castings when the quality level seems to change? Is it that one ladle of iron is better than another? Is it that fast pouring of hot metal results in better castings than does slow pouring of cold metal? Or is it any other combination of factors? Pareto principles again say that a few factors make the job right or wrong. These clues will point out the variable area that must be controlled, and where that variable must be controlled to give quality in subsequent production.

Such a procedure minimizes the chances of making two or three good castings after a change and releasing a job which might still have heavy scrap. It is possible to roll a pair of dice twice and have "snake eyes" both times, but if we can see only the tops of

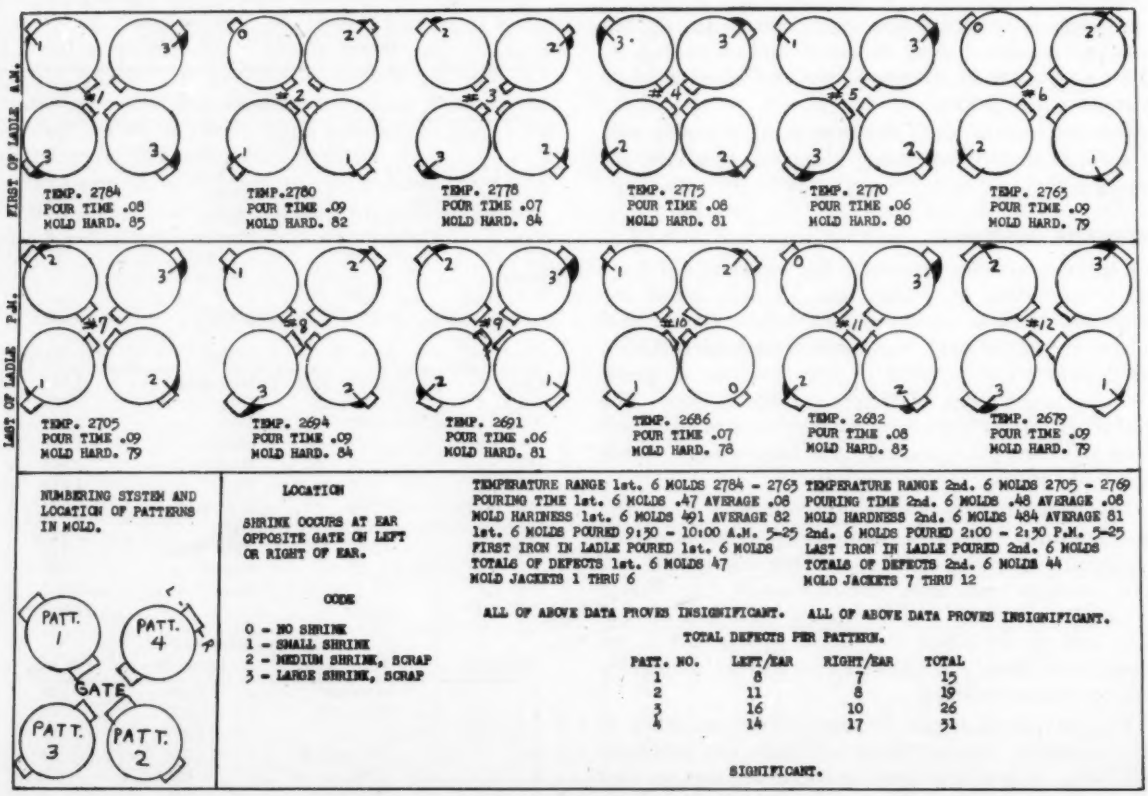


Fig. 8 — More detailed Multi-Vari Data Chart.

the dice after they are thrown, we can not be sure that all sides of both dice do not have just one dot each. On "legitimate" dice it is possible to do the same thing, but the odds are 1,300 to 1 against it. It is the objective of this program to make decisions where the odds of success are favorable.

Pilot runs are essentially production runs of multi-vari type which sometimes go all the way through the customer's machining and testing facilities. This helps determine the critical foundry controls necessary to assure final customer satisfaction. Such test runs are usually 50 pieces. There is no nondestructive testing technique in the world any better than this. Foundry control on subsequent production is thus mimimum in coverage, but specifically centered on those factors which make that casting right.

X-Plots

As processes are proved to be critical, they should be tied down and not removed later because of economic pressures. This is easier said than done. To withstand pressures, controls must be proved, and this requires forced experiments. If the defect can be turned off or on at will, it is sure to be controlled. The author's company uses the X-Plot which is merely a "cookbook" form for the analysis variance. An oversimplified definition of this statistical quality control tool is that it enables one to level out all of the natural variables of a process of job, and compare the results of one, two or three pairs of ideas mathematically. It then gives the mathematical odds of this relationship occurring by chance.

The experiment shown in Fig. 9 is balanced with six trials performed by each method. Each trial, or part of a test, is at least two pieces which can be either one or two molds. If there are more than two pieces per mold, then only one mold is required for each trial. Since there are four pieces in this particular mold, the example shows four castings per trial.

This particular X-Plot shown in Fig. 9 was run on a job which had no pattern of defect except a "within mold" difference. Rather than completely regating and rerigging the pattern, which is expensive, it was thought that a different sand might improve the overall quality to such an extent that regating would not be necessary.

Each of the pieces is rated by the rating scale shown at the bottom of the X-Plot (Fig. 10). Such a rating scale is not perfect, but quite acceptable where there are several castings per trial. Ranking from the best to the worst, one through 24, is more accurate as a general rule, and is usually done for that size experiment, but lots of 48 or more are seldom ranked because of the extra work involved.

The first trial was rated 0, 1, 3, and 3 for a total of seven points. Figure 10 is the last half of the X-Plot. All of the encircled figures for each idea are totaled and shown as 24 for the No. 3 facing and 41 for the No. 6 system. The new idea is lower, but this could still happen by chance. For instance, if a coin were flipped 65 times, the result might be 24 heads and 41 tails. Therefore, the statistical testing formula is followed, which will answer this question.

Naturally, the statistical calculation of this problem could be done by the usual analysis of variance format as shown in Table 2, but the principal idea of the X-Plot is to enable production line foremen to

TRIAL	METHOD TO BE USED	MOLD NOS.	REMARKS: (FOR EXAMPLE POUR TEMP, POUR TIME, MOLD HARD, TIME OF DAY)	MOLD CASTING NUMBER	DEFECT DESCRIPTION			MOLD CASTING NUMBER	DEFECT DESCRIPTION		
					TYPE	LOCATION	RATING		TYPE	LOCATION	RATING
1	# 6	1		1	SHRINK	NEXT TO	0				
	SYSTEM			2	"	EAR	1				
				3	"	OPPOSITE	3				
				4	"	GATE	3				
2	#3	2		1	"	"	0				
	FACING			2	"	"	1				
NEW IDEA				3	"	"	1				
				4	"	"	1				
3	#3	3		1	"	"	1				
	FACING			2	"	"	1				
NEW IDEA				3	"	"	1				
				4	"	"	3				
4	#6	4		1	"	"	1				
	SYSTEM			2	"	"	1				
				3	"	"	1				
				4	"	"	3				
5	#6	5		1	"	"	1				
	SYSTEM			2	"	"	1				
				3	"	"	1				
				4	"	"	3				
6	#3	6		1	"	"	0				
	FACING			2	"	"	1				
NEW IDEA				3	"	"	1				
				4	"	"	1				

Fig. 9 — X-Plot experimental layout first six trials.

set up and run an experiment, and then calculate the significance of the experiment themselves. This is virtually impossible if they must be taught the full analysis of variance principles and calculation. The answer to this problem has been the author's company's "X-Plot Analysis Form."

The "X-Plot Analysis Form" shown in Fig. 11 is filled out from the data shown on the X-plot experiment. The new idea total of 24 is written in and becomes (A) and is multiplied times letter (A) which is therefore 24 again, which equals 576. The present method total of 41 is written in and becomes letter (B) and is multiplied times (B) which equals 1681. The total of these two figures, 576 plus 1681, equals 2257 as letter (C). The grand total of 65 is handled similarly, and eventually equals 4225 as letter (E).

The individual trial totals are squared so that the seven becomes 49, the three becomes nine, etc., for a total of 395 as the total or letter (F).

Then the blocks are filled in on the right side of the "X-Plot Analysis Form" with the appropriate letter number. By following the divisions and subtractions indicated, a significance ratio is calculated. In this case the significance ratio was 12.8 which meant that no. 3 facing was proven superior to no. 6 system sand for this job. This distribution of scores could have happened by chance less than one time in 100 had there been no difference between the sands. Sometimes, however, such an experiment showing scores of 40 and 20 will not be significantly different due to the individual trial total variations. The X-Plot analysis form, by virtue of dealing in probability estimates, will clearly differentiate between the two cases described.

Thus, an overall improvement on this job was effected by the use of No. 3 facing, but the problem was not completely solved without changing the gate size and shape on certain patterns on this plate. This ultimately resulted in quality castings being produced consistently.

This technique gives the line foreman a bat and a chance to hit statistically. He is thus an active member of the quality improvement squad, not just a labor foreman. In fact, he is not allowed to make changes without using this technique and finds it necessary to use it almost daily.

Special statistically designed experiments* are utilized to solve specific problems when there are more variables or the operation is such that the X-Plot is not applicable.

*Dorian Shainin, "The Statistically Designed Experiment: A Tool for Process and Product Improvement," *Harvard Business Review*, July-August 1957.

TRIAL	METHOD TO BE USED	MOLD NOS.	REMARKS: (FOR EXAMPLE POUR TEMP, POUR TIME, MOLD HARD, TIME OF DAY)	MOLD CASTING NUMBER	DEFECT DESCRIPTION			MOLD CASTING NUMBER	DEFECT DESCRIPTION		
					TYPE	LOCATION	RATING		TYPE	LOCATION	RATING
7	#6	7		1	SHRINK	NEXT TO	0				
	SYSTEM			2	"	EAR	1				
				3	"	OPPOSITE	1				
				4	"	GATE	3				
8	#6	8		1	"	"	1				
	SYSTEM			2	"	"	1				
				3	"	"	3				
				4	"	"	3				
9	#3	9		1	"	"	0				
	FACING			2	"	"	1				
NEW IDEA				3	"	"	1				
				4	"	"	1				
10	#6	10		1	"	"	1				
	SYSTEM			2	"	"	2				
				3	"	"	3				
				4	"	"	3				
11	#3	11		1	"	"	0				
	FACING			2	"	"	1				
NEW IDEA				3	"	"	1				
				4	"	"	2				
12	#3	12		1	"	"	0				
	FACING			2	"	"	1				
NEW IDEA				3	"	"	1				
				4	"	"	3				

RATING SCALE
0 - NO DEFECT
1 - SMALL DEFECT - NOT SCRAP
2 - MEDIUM DEFECT - SCRAP
3 - HEAVY DEFECT - SCRAP

RANK TOTALS:
NEW IDEA 24 #3 FACING
PRESENT METHOD 41 #6 SYSTEM
GRAND TOTAL 65

Fig. 10 — X-Plot experimental layout last six trials.

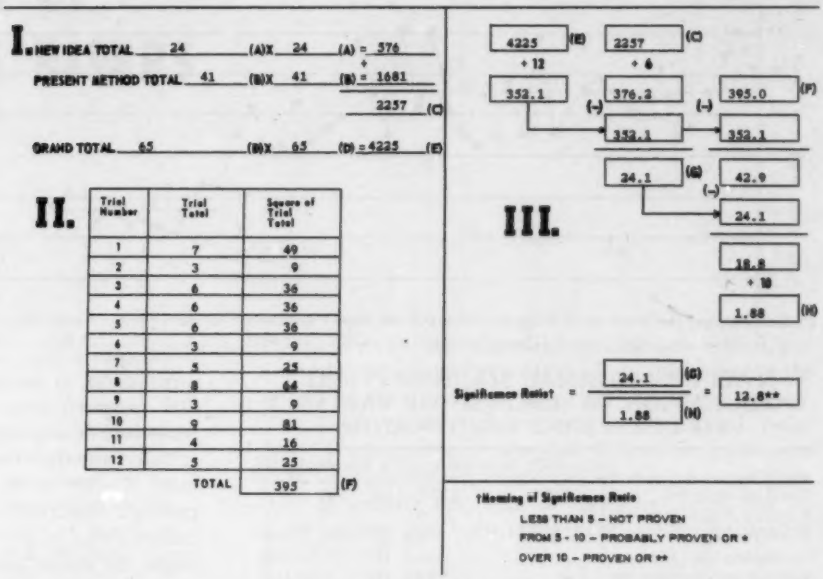


Fig. 11 — Author's company X-Plot Analysis Form for simplified testing of significance for methods tried in X-Plot.

CONCLUSIONS AND RECOMMENDATIONS

Results from this overall program have been extremely good. Figure 12 represents the history of the program showing that customer returns have been reduced substantially. This was not the main objective of the program but is certainly desirable. On the same chart is shown the history of the man-hours per ton expended in the quality control effort. Again a reduction in man-hours per ton was not the objective, but they are now less than ever before. This also includes clerical work on report systems and most of the experimental design and analysis for experiments. Only one salaried person was added into this entire program.

Figure 13 shows the total hard and soft iron scrap history. Results from this overall program have been beyond the most optimistic dreams, and it is not certain that the economic low per cent scrap has yet been achieved. The 12 months ending August 31, 1958, had a weighted average of 5.4 per cent. This is less than half the original 10.9 per cent average experienced prior to the inception of our quality improvement program.

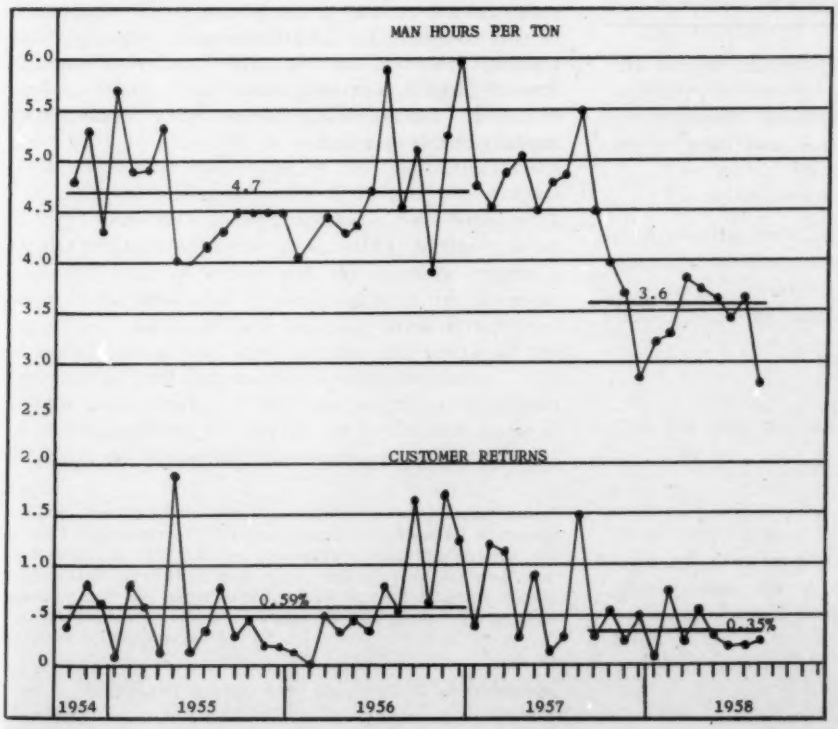


Fig. 12 — Quality control departmental man-hours per ton and customer returns as a percentage of shipments were both reduced during this program.

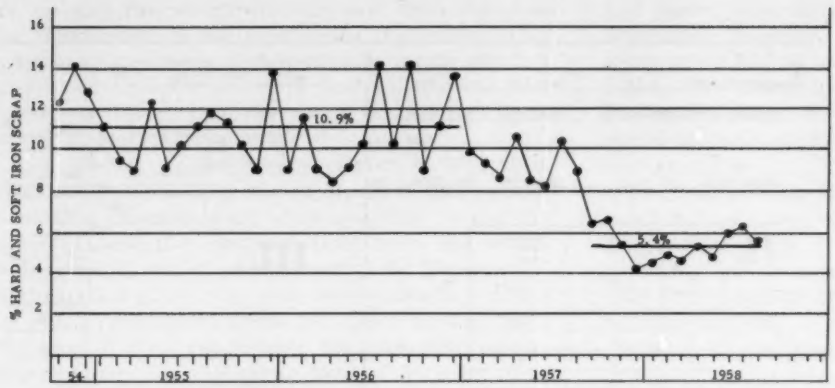


Fig. 13 — Total malleable division hard iron and soft iron scrap was reduced by 50 per cent during this program with further reductions still economically feasible.

TABLE 2 – NORMAL MATHEMATICAL CALCULATION OF ANALYSIS OF VARIANCE TO DETERMINE SIGNIFICANCE

	Mean (DF)	Main Effect (DF)	Residual (DF)
Total of squares	4,225	2,257	395
Number values per total	12	6	1
Uncorrected sum squares	352.1 (1)	376.2 (2)	395 (12)
Correction for the mean	352.1 (1)	352.1 (1)	352.1 (1)
Sum of squares corr. for mean	24.1 (1)	42.9 (11)	
Correction for main effect		24.1 (1)	
Sum of squares corr. for main effect		18.8 (10)	
Mean square, variance, or sigma squares	24.1	1.88	

Significance ratio $\dfrac{24.1\ (1)}{1.88} = 12.8^{**}$

$$F = \frac{1}{10} = \frac{4.96^{*}}{10.004^{**}}$$

*4.96 equals the 0.05 level which means that one time in 20 could a number as high as 4.96 occur by chance. This is rounded off to 5 on the X-Plot Analysis Form for sake of simplicity.

**10.04 equals the 0.01 level which means that one time in 100 could a number as high as 10.04 occur by chance. This is rounded off to 10 on the X-Plot Analysis Form for sake of simplicity.

The statistical aspects of the program herein are vitally important to the success of such a program, but must be combined with dynamic management of the efforts of quality control and production supervision to be most effective. The techniques of this program can be learned rapidly by qualified personnel of any organization.

Calculate the true costs for lack of quality. Determine if a few of the jobs, processes, defects, classes of castings, machines or other categories show that the Pareto concept is valid. Estimate the savings which might be made and the cost of such a program. If the savings are worth the expenditure of time and money, go after them because by so doing, greater customer satisfaction will be achieved with castings in contrast to fabrications, forgings, and the like. Foundries will be forced to produce better and more consistent castings if they are to be a progressive rather than a declining industry.

ACKNOWLEDGMENT

The author wishes to express his appreciation to the management of Texas Foundries, Inc., for its permission to carry out this program; to J. R. Speer, Jr., and F. W. Jacobs of Texas Foundries for their technical advice, and to the photographic department of this company for its assistance in preparing illustrations presented in this paper; and to all quality control department employees and foundry supervision at Texas Foundries for their wholehearted support of this program.

DIMENSIONING OF SAND CASTING RISERS

By Harish D. Merchant

ABSTRACT

A new approach for risering sand castings whereby a general equation for risers is obtained is developed. Using the equation, the size of riser or risers for a given casting can be easily found. The general equation is expressed in graphic form which is convenient to use, provided the values of the parameters of the equation are known. These are the solidification shrinkage at pouring temperature,* the ratio of height to diameter of riser and number of risers required.

INTRODUCTION

The dimensioning of sand casting risers has been more of an art than science in the foundry industry. The transition of risering from art to science was started in 1948 by Caine.[3] The approach to the dimensioning was empirical in character, and was applicable to a narrow set of conditions for certain specifications of steel. In actual working practice, such formulation for the dimensioning of risers quite frequently may fail to give satisfactory results. The reason is obvious. The working conditions of a foundry are extremely dynamic as compared to a laboratory. Thus, foundrymen prefer to use excessively large instead of analytically exact risers.

There are three problems involved in riser dimensioning: size, shape and position. Caine's[3] work throws light on size, while Bishop and others[4] in their work in 1955 tell in addition something about what should be the shape of the cylindrical riser. The later work also is partly empirical. The workers at Naval Research Laboratory have done some work on the position of risers on a casting. Again, it is an empirical type of study, and the authors suggest a set of rules, successful in the laboratory, for positioning a riser. Each of these problem areas is confined to nodular iron and steel for size and shape of the riser, and to steel only for positioning the riser.

This paper makes a different empirical approach to to dimensioning of risers as far as size and shape of the risers are concerned. The method is general in the sense that it can be applied to any type of metal or alloy. As in any scientific approach to a problem, a number of assumptions are made in the course of the paper as they tend to make the problem simpler. The risers considered are only top-insulated cylindrical risers, and data available in literature are used. A similar general approach can be used for the question of feeding distances of different risers, but no such attempt has been made here due to insufficient data.

CHVORINOV'S RULE

The risering requirements of a sand casting depend mainly upon two factors, viz., the thermal properties and the solidification characteristics of the metal. But the riser size would also be affected by the size and shape of casting and riser, type of sand used, the position of gate and riser and other infinite variety of situations under which the casting is made. In this study, in which cylindrical top-insulated risers are considered, the following variables are considered:

V_c – volume of casting.
S_c – surface of casting.
V_r – volume of riser.
S_r – surface of riser.
b – solidification shrinkage at pouring temperature.
c – contant for the metal dependent on solidification behavior.

According to Chvorinov,[1] the time a metal body remains molten is proportional to the square of the ratio, volume (V) to surface area (S) of the body. Thus, the time molten for casting is

$$T_c \alpha \frac{V_c^2}{S_c^2}$$

and the time molten for riser is

$$T_r \alpha \frac{V_r^2}{S_r^2}$$

Most frequently a top-insulated riser is used, as this type of riser is little affected by the position of gates,[2] and for a number of other obvious advantages. Now, taking into account the top-insulated riser,

$$T_r = K_r \frac{V_r^2}{S_r^2}$$

*The term "solidification shrinkage at pouring temperature" is used repeatedly. It does not mean solidification shrinkage that occurs at the pouring temperature, but implies that the solidification shrinkage may vary with varying pouring temperature, and this paper considers only that solidification shrinkage which occurs when the metal is poured at the pouring temperature in question.

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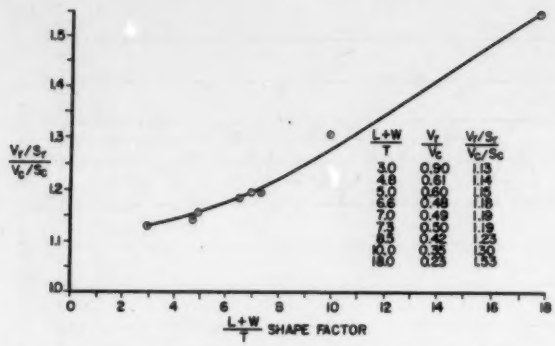


Fig. 1 — Relationship between shape factor and freezing ratio where L+W/T is plotted against the freezing ratio $\frac{V_r/S_r}{V_c/S_c}$. As the former increases, the latter also increases.

$$T_c = K_c \frac{V_c^2}{S_c^2}$$

$$\therefore \frac{T_r}{T_c} = \frac{K_r}{K_c} \quad \frac{V_r^2/S_r^2}{V_c^2/S_c^2} = K \frac{V_r^2/S_r^2}{V_c^2/S_c^2}$$

where, $K_r/K_c = K$

if $K_r = K_c$, $K = 1$

and $$\frac{T_r}{T_c} = \frac{V_r^2/S_r^2}{V_c^2/S_c^2}$$

Obviously, $K_r = K_c$ when at any moment the rate of heat dissipation of riser is same as that of casting, which in turn can happen only when the riser and the casting have the same shape, both are poured at the same temperature and are, in general, cooling under similar conditions. Sometimes, using the last equation, attempts have been made to dimension a riser for a given casting by an argument that the riser should remain molten 1½ times (or some such arbitrary number) the time the casting may remain molten. This is the same as selecting an arbitrary T_r/T_c and substituting in the equation

$$\frac{T_r}{T_c} = \frac{V_r^2/S_r^2}{V_c^2/S_c^2}$$

to find the riser dimension. This equation is only true when $K = 1$, which happens only under the restrictions explained. Hence the Chvorinov's rule[1] cannot be applied rigorously without knowing K or K_r and K_c.

$\frac{V_r/S_r}{V_c/S_c}$ or $\sqrt{\frac{1}{K} \cdot \frac{T_r}{T_c}}$ is known as a freezing ratio, which is low for chunky castings like cubes and spheres, and high for thin and long castings like plates and bars. For steel, this can be easily shown by combining the curves obtained by Caine[3] and Bishop and others.[4] The former has presented a curve of freezing ratio vs. V_r/V_c and the latter, a curve of L+W/T vs. V_r/V_c

where: L = length of casting
W = width of casting
T = thickness of casting

Instead, here L+W/T is plotted against freezing ratio $\frac{V_r/S_r}{V_c/S_c}$ (Fig. 1). It is obvious that, as L+W/T increases, freezing ratio increases. The curve by Bishop and others[4] is only true when the ratio of height of the riser (H) to the diameter of the riser (D) is between 0.5 and 1. Hence, Fig. 1 is only true when H/D is between 0.5 to 1. Actually, a series of such curves may be obtained for different values of H/D, if exact data are available.

From the above discussion we can conclude that Chvorinov's rule[1] is a complex mathematical relation for a given set of casting conditions between the freezing ratio, shape of riser and shape of casting. Curves obtained by Caine[3] and Bishop and others[4] are thus a restatement of Chvorinov's rule; the only difference being that Chvorinov did not consider the shape variable, and considered freezing ratio constant for a metal body irrespective of its shape.

RISERING FOR STEEL

The results obtained for steel by Bishop and others[4] were thoroughly studied. From their data, riser volume (V_r), riser surface (S_r), casting volume (V_c) and casting surface (S_c) were calculated (Table 1). A graph of V_r/V_c and S_r/S_c was plotted (Fig. 2). A linear relation obtained is given by

$$\frac{S_r}{S_c} = m \frac{V_r}{V_c} - c' \qquad (1)$$

where m and c′ are constants. For steel, m = 0.81, c′ = 0.041; making the equation,

$$\frac{S_r}{S_c} = 0.81 \frac{V_r}{V_c} - 0.041 \qquad (2)$$

Substituting, $m = \frac{1}{c}$ and $\frac{c'}{m} = b$, $x = \frac{V_r S_c}{S_r V_c}$ and $y = \frac{V_r}{V_c}$ in equation (1) we get,

$$x = \frac{cy}{y - b} \qquad (3)$$

where b and c are constants. For steel,

$$x = \frac{1.265\ y}{y - 0.050} \quad \text{(Fig. 3)} \qquad (4)$$

Now, Caine[3] obtained the relation in the form of,

$$x = \frac{a}{y - b} + c$$

where a, b and c, are constants, or

$$x = \frac{0.12}{y - 0.05} + 1 \quad \text{(Fig. 4)} \qquad (5)$$

What he presumably did was to reason two points (∞, b) and (c, ∞), and then fit a curve on the scatter diagram obtained by the experiments. But, mathematically speaking, two points are not sufficient to define a curve. From the points (∞, b) and (c, ∞) a number of curves can be obtained such as,

TABLE 1 – RISER DATA, STEEL

No.	Size	V_c	S_c	Min. Riser D	Min. Riser H	V_r	S_r	$\frac{V_c}{S_c}$	$\frac{V_r}{S_r}$	$\frac{V_r}{V_c}$	$\frac{S_r}{S_c}$	$X = \frac{V_r S_c}{S_r V_c}$
1	2x4x8	64	112	3½	4	38.4	49.6	0.57	0.72	0.60	0.48	1.40
2	2x8x8	128	192	5	2½	49	58.9	0.67	0.83	0.38	0.31	1.24
3	2x4x12	96	160	4½	4	63.6	72.4	0.60	0.88	0.66	0.45	1.47
4	2x12x12	288	384	6	3½	99	94.1	0.75	1.05	0.34	0.25	1.40
5	2x24x24	1152	1344	6	8	230	179	0.86	1.28	0.20	0.13	1.46
6	2x25x25	1250	1450	7	4¼	175	131.9	0.86	1.33	0.14	0.091	1.54
7	2x25x25	1250	1450	8	3	150	125.7	0.86	1.19	0.12	0.087	1.38
8	2x36x36	2590	2880	8	6½	330	213.6	0.90	1.54	0.12	0.74	1.71
9	2x36x36	2590	2880	12	4½	510	282.7	0.90	1.80	0.20	0.098	2.00
10	3x3x9	81	126	4½	3½	56	65.4	0.64	0.86	0.69	0.52	1.34
11	3x9x9	245	270	5½	6	143	127.4	0.91	1.12	0.59	0.59	1.23
12	3x6x12	216	252	7	2	79	82.5	0.86	0.96	0.36	0.36	1.12
13	3x9x12	325	342	6	4½	127	113.1	0.95	1.11	0.39	0.33	1.16
14	3x12x12	435	432	6	5½	156	131.9	1.00	1.18	0.37	0.30	1.8
15	3x12x12	432	432	8	3	151	125.6	1.00	1.20	0.35	0.29	1.15
16	3x28x28	2350	1904	9	10	635	346.2	1.23	1.83	0.27	0.18	1.49
17	3x28x28	2350	1904	12	6	680	339.2	1.23	2.00	0.29	0.18	1.63
18	4x4x4	64	96	5	3	58	66.8	0.67	0.88	0.92	0.69	1.31
19	4x4x4	64	96	6	1½	42.5	56.6	0.67	0.75	0.66	0.59	1.12
20	4x4x8	128	160	6	3½	99	94.3	0.80	1.05	0.77	0.98	0.789
21	4x4x12	192	224	6	5	141	122.5	0.86	1.15	0.72	0.55	1.33
22	4x4x18	288	320	6	4	113	103.7	0.90	1.09	0.40	0.32	1.21
23	4x4x20	320	352	7	4	155	126.1	0.91	1.23	0.48	0.36	1.35
24	4x4x28	448	480	11	1½	143	146.8	0.93	0.97	0.32	0.37	1.04
25	4x4x31	495	528	7	5	193	148.4	0.94	1.30	0.38	0.28	1.38
26	4x20x20	1600	1120	10	11	863	423.9	1.43	2.04	0.54	0.38	1.43
27	4x20x20	1600	1120	13	5½	735	357.2	1.43	2.09	0.46	0.32	1.46
28	4x25x25	2400	1650	10	13	1020	486.7	1.45	2.17	0.42	0.29	1.50
29	7x7x7	342	294	9	5	320	204.9	1.16	1.56	0.92	0.70	1.34
30	4x4x12	192	224	6	5	141	122.5	0.86	1.15	0.73	0.55	1.33
31	4x4x28	448	480	7	6	232	170.4	0.93	1.36	0.53	0.35	1.46

$$x = \frac{a}{y-b} + c \qquad A$$

$$x = \frac{a}{(y-b)^2} + c \qquad B$$

$$x = \frac{a}{(y-b)^3} + c \qquad C$$

$$x = \frac{y-b+c}{y-b} \qquad D$$

$$x = \frac{cy}{y-b} \qquad E, \text{ etc.}$$

None of these curves can be expressed in the form of equation (1), $\frac{S_r}{S_c} = m\frac{V_r}{V_c} - c'$, except for equation (E).

A comparison between equation (4) and (5) is given in Fig. 5a. The differences in two curves obtained probably may be due to one or more of the following reasons:

1) Bishop[4] judged defective castings by radiographs, Caine[3] used no radiographs.
2) In calculating the surface areas of riser and casting, the area of the plane of intersection of riser and casting was excluded from riser surface but included in casting surface. In doing so, a better fit on the scatter diagram was obtained. Caine[3] may or may not have included this area into riser and casting surface areas.
3) Bishop and others[4] made an assumption from their laboratory experience "that the minimum riser required for a given shape which had been fed with a slightly oversize riser could be established with reasonable accuracy by subtracting from the riser height the distance by which the tip of the riser shrinkage missed the contact line."[5] It is not known how Caine[3] obtained the minimum risers.
4) The equations (2) and (4) are obtained by the least square method so as to get the best fit on the scatter diagram. It is not known how Caine

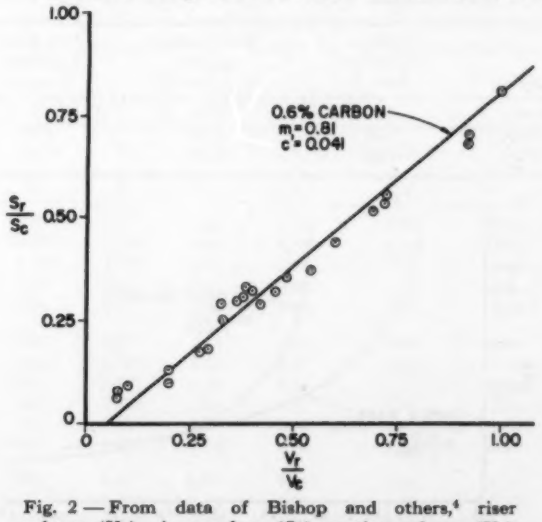


Fig. 2 — From data of Bishop and others,[4] riser volume (V_r), riser surface (S_r), casting volume (V_c) and casting surface (S_c) were calculated, and the scatter graph of steel data ($S_r/S_c \rightarrow V_r/V_c$) was plotted.

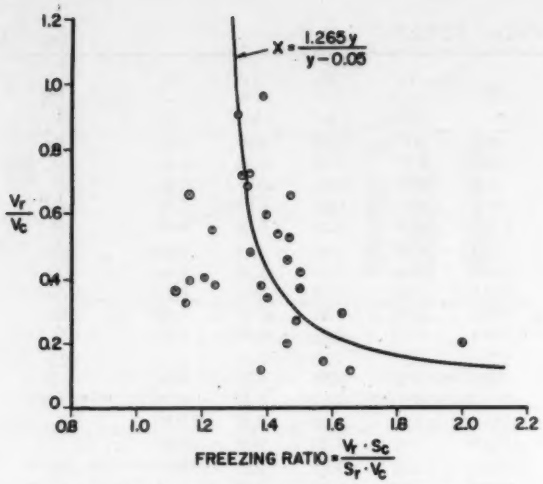


Fig. 3 — The scatter graph of steel data plotted with freezing ratio (F.R.) — F.R. $= V_r/V_c$.

obtained the experimental curve in Fig. 4. By the method of least squares, equation (5) is almost unobtainable unless one of the constants *a*, *b*, or *c*, is known. If the dotted experimental curve is roughly drawn on a scatter diagram, and a theoretical curve is made to fit on it, this can be misleading. It would also be incorrect to assume that any point falling on right of the curve would be invariably sound, and on left, invariably defective. Within certain limits,[4] all the "just sound" and "just defective" points will fall on both sides of the curve between a narrow strip. The limits, however, are difficult to define as the casting conditions change from foundry to foundry. It would be safe to utilize this average curve, unless a casting absolutely free from microshrinkage is to be obtained.

5) Bishop[4] varied the ratio of height to diameter of the riser (H/D). As explained later (Fig. 6), the ratio H/D changes riser requirements considerably, particularly in chunky castings. H/D ratios used by Bishop[4] are mostly less than one.

TABLE 2 – (OPEN INSULATED) RISER DIAMETER FOR STEEL WHEN D = H

Casting	Caine's Equations[3] (Equ. 5)	Bishop and others[4] (Ave. Curve)	New Riser Equation (Equ. 4)
4x4x4	4.50	4.70	4.3
2x4x8	4.05	3.50	3.75
1x8x8	3.30	2.75	3.00

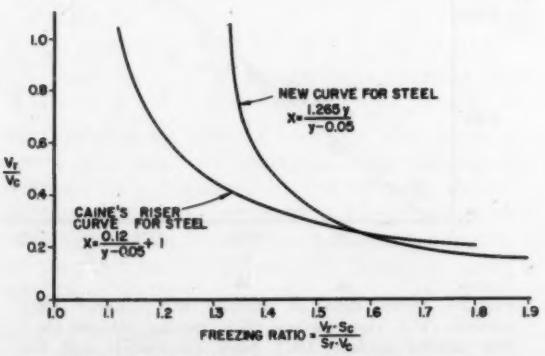


Fig. 5a — A comparison between equation (4) and (5) showing Caine's[3] curve and the new curve for steel.

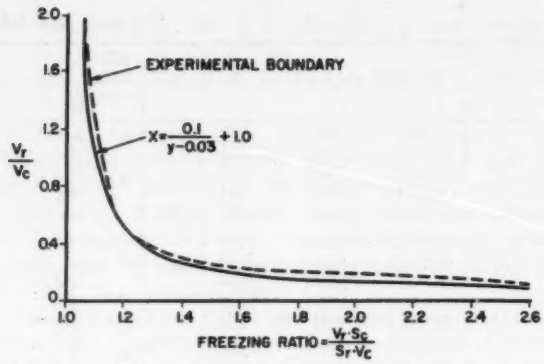


Fig. 4 — Graphs from Caine[3] where a, b and c are constants, or $x = \dfrac{0.12}{y - 0.05} + 1$.

Riser dimensions for steel were calculated by using equations (4) and (5). In equation (4), the area of plane of intersection is excluded from riser surface area but is included in casting surface area. In equation (5), the area of plane of intersection is included in both riser and casting surfaces. The riser dimensions for the same castings are also calculated from the average graph given by Bishop and others.[4] The results obtained by the three methods are presented in Table 2.

With the practical point of view, the results are not quite identical. Mathematically, equations (4) and (5) are identical when a = bc. When the area of the plane of intersection of riser and casting is included in both riser and casting surface areas, c = 1, b = 0.05 for both equations; and when the same area is included in casting surface only, c = 1.25, b = 0.05 for both equations (4) and (5) (Fig. 5b). According to equation (4), in the first case a = bc = 0.05 and in the second case a = bc = 0.0625. According to equation (5), a = 0.12.

This means that the only difference between Caine's curve (equ. 4) and new riser curve (equ. 5) is that the value of *a* is less in latter, so that the

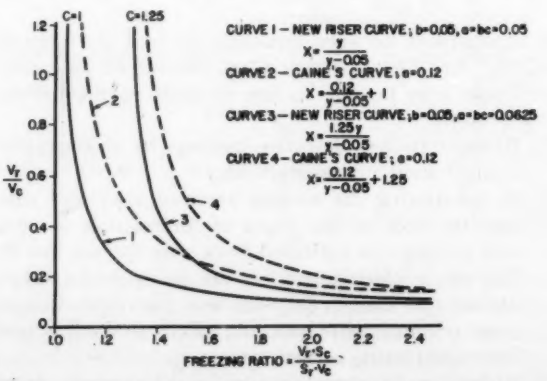


Fig. 5b — New curve for steel, and Caine's[3] curve, for C = 1 and C = 1.25.

knee of the Caine's curve is shifted a little toward X and Y axes to give the new riser curve (Fig. 5a). As actual difference in the dimenstions of risers obtained by using equations (4) and (5) is not great, it would not make much difference using any of them (irrespective of which one gives "just sound" castings), if castings completely free from microshrinkage are not to be obtained. The area of the plane of intersection of riser and casting must not be included in riser surface, because often the riser is necked down, and the value of c obtained without the inclusion of this common area further helps in understanding the physical meaning of b and c (this will be explained later Fig. 12a).

For different values of $\frac{L+W}{T}$, $\frac{V_r}{V_c}$ was plotted against H/D. The data were obtained from the work of Bishop and others[4] (Fig. 6). $\frac{L+W}{T} = 2-4$ is for cubes or short bars, $\frac{L+W}{T} = 6-8$ is for long bars or short plates and $\frac{L+W}{T} = 10-12$ is for plates. It is apparent that we can use a lower riser volume, and hence obtain higher yield if low H/D ratio is used. This is particularly true, and perhaps is it necessary, for chunkier castings where $\frac{L+W}{T}$ is lower and savings in volume of riser by reducing H/D ratio are higher (Table 3).

Although all three lines (Fig. 6) are obtained by the least square method, the results, are not quantitatively exact due to insufficiency of data. However, the emphasis is clear that the risers with height smaller than diameter should be used for chunkier castings to obtain greater yield. In practice, the ratio H/D less than 0.5 becomes impracticable, due to difficulties of removing riser from the casting, and to the formation of a hot spot at the junction of riser and casting. Even necking down of the top riser does not help when the riser diameter is too large as compared to the height. Thus, it may be suggested that H/D between 0.5 and one should be used as it may be practicable.

TABLE 3 – EFFECT ON V_r/V_c OF USING DIFFERENT RISER HEIGHTS

$\frac{L+W}{T}$	H/D=1	H/D=0.5	(H/D = –1) – (H/D = ½)
2-4	1.6	0.78	0.82
6-8	0.58	0.34	0.24
10-12	0.40	0.28	0.12

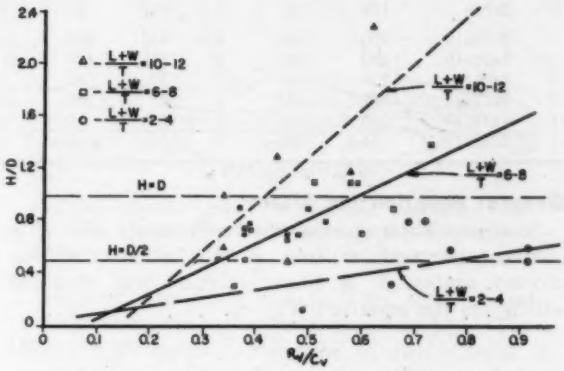


Fig. 6—Graph showing effect of H/D on riser volume.

RISERING FOR NODULAR IRON

Similar treatment, as for steel, was given to the results obtained by Bishop and Ackerlind[5] for nodular irons of different carbon equivalent. The data are given in Tables 4a, 4b and 4c. The graphs are again obtained by the least square method, and plotted in Figs. 7a and 7b. The values of m, c', b and c in equations (2) and (3), for nodular iron as well as steel, are summarized in Table 5.

It is easy to conclude that: (a) the values of b for nodular irons with different carbon equivalents conform fairly well with those obtained by the method proposed by Reynolds and others[6] (see Appendix A); (b) the riser requirements of nodular iron change drastically with change in carbon equivalent.

TABLE 4A – LOW-CARBON NODULAR IRON DATA (C.E. = 3.54-3.82)

No.	Size	V_c	S_c	Min. Riser D	Min. Riser H	V_r	S_r	$\frac{V_c}{S_c}$	$\frac{V_r}{S_r}$	$\frac{V_r}{V_c}$	$\frac{S_r}{S_c}$	$X = \frac{V_r S_c}{S_r V_c}$	C.E.
1	4x4x4	64	96	6	3¾	106	98.8	0.67	1.07	1.67	1.03	1.60	3.73
2	3x6x6	108	144	7	4½	173	137.4	0.75	1.26	1.60	0.95	1.68	3.66
3	3x6x6	108	144	7	4½	173	137.4	0.75	1.26	1.61	0.96	1.68	3.66
4	3x6x6	108	144	7	4½	173	137.4	0.75	1.26	1.61	0.96	1.68	3.66
5	3x3x15	135	198	7	3½	135	115.4	0.68	1.17	1.00	0.58	1.72	3.66
6	3x3x15	135	198	7	3⅞	150	123.6	0.68	1.21	1.11	0.62	1.77	3.66
7	3x3x15	135	198	6	4½	127	113	0.68	1.12	0.94	0.57	1.65	3.66
8	2x8x8	128	192	7	3	116	105.5	0.67	1.10	0.90	0.55	1.64	3.66
9	2x8x8	128	192	7	3½	135	115.4	0.67	1.17	1.06	0.60	1.75	3.66
10	2x8x8	128	192	6	4¼	120	108.3	0.67	1.11	0.94	0.56	1.66	3.66
11	4x4x4	64	96	6	3½	99	94.1	0.67	1.05	1.54	0.98	1.57	3.73
12	2x14x14	392	504	7	4¼	279	197.4	0.78	1.44	0.71	0.39	1.85	3.67
13	1x8x8	64	160	3½	3	29	37.4	0.4	0.775	0.45	0.23	1.94	L3.54
14	1x12x12	144	336	5	3¼	64	70.7	0.43	0.905	0.44	0.21	2.11	L3.54
15	1x14x14	196	448	5	3	59	66.8	0.44	0.883	0.30	0.30	2.02	3.77
16	1x14x14	196	448	5	3¼	64	70.7	0.44	0.905	0.33	0.14	2.07	H3.82
17	1x14x14	196	448	5	3¼	64	70.7	0.44	0.905	0.33	0.14	2.07	H3.82

TABLE 4B – MEDIUM-CARBON NODULAR IRON DATA (C.E. = 3.88-4.32)

No.	Size	V_c	S_c	Min. Riser D	Min. Riser H	V_r	S_r	$\frac{V_c}{S_c}$	$\frac{V_r}{S_r}$	$\frac{V_r}{V_c}$	$\frac{S_r}{S_c}$	$X = \frac{V_r S_c}{S_r V_c}$	C.E.
1	6x6x6	216	216	6	5	140	122.5	1	1.15	0.65	0.57	1.15	L3.88
2	6x6x6	216	216	8	4½	226	164	1	1.38	1.05	0.76	1.38	L3.88
3	3x6x6	108	144	5	3¼	64	70.7	0.75	0.905	0.60	0.49	1.21	H4.09
4	4x4x20	320	352	8	5	235	176	0.91	1.34	0.80	0.5	1.46	L3.88
5	4x4x20	320	352	6	4¼	128	113	0.91	1.13	0.40	0.32	1.45	Hy4.32
6	2x8x8	128	192	5	3¼	64	70.7	0.67	0.91	0.50	0.37	1.36	H4.09
7	2x10x10	200	280	6	4½	128	113	0.71	1.13	0.64	0.40	1.59	L3.88
8	2x10x10	200	280	6	4½	128	113	0.71	1.13	0.64	0.40	1.59	L3.88
9	2x6x18	216	308	6	3	85	84.8	0.70	1.00	0.39	0.28	1.43	H4.09
10	2x14x14	392	504	8	5	235	176	0.78	1.34	0.64	0.35	1.72	L3.88
11	1x14x14	196	448	5	3½	69	64.7	0.44	1.01	0.35	0.15	2.30	L3.88
12	1x12x12	144	336	5	3½	44	57	0.43	0.77	0.30	0.17	1.80	H4.09

General Equation for Risers

Studying Table 5, one can easily notice that c' is constant for nodular iron as well as steel. c was plotted against b. A linear relation was obtained which has the equation (Fig. 8),

$$c = 25b \qquad (6)$$

Now, the new riser equation is given by,

$$\frac{S_r}{S_c} = m\frac{V_r}{V_c} - c' \qquad (1)$$

Substituting,

$$S_r = \pi/4\ D^2 + \pi DH$$
$$V_r = \pi/4\ D^2H$$
$$H/D = p$$

equation (1) can be expressed in the form,

$$\frac{(4p+1)}{S_c} = \frac{p\,m\,D}{V_c} - \frac{1.275c'}{D^2} \qquad (7)$$

Substituting, $m = \frac{1}{c}$ and $c' = \frac{b}{c}$, equation (7) can be expressed in the form of,

$$\frac{(4p+1)c}{S_c} = \frac{p\,D}{V_c} - \frac{1.275b}{D^2} \qquad (8)$$

Substituting, equation (6) (c = 25b) in equation (8),

$$\frac{25(4p+1)}{S_c} = \frac{p\,D}{b\,V_c} - \frac{1.275}{D^2} \qquad (9)$$

This is the general equation of the riser. The results obtained by using it are almost identical to those of Caine[3], or Bishop and others.[4,5]

If the casting is big, one riser, however big, is not capable of supplying metal to all parts of the casting. In such cases, more than one riser is needed. If n = the number of risers needed, equation (9) can be expressed as,

$$\frac{25(4p+1)}{S_c} = \frac{p\,D}{b\,V_c} - \frac{1.275}{n\,D^2} \qquad (10)$$

In other words, 1/nth part of casting be treated as a single riser system by dividing S_c and V_c each by n.

Equations (9) and (10) cannot be expressed graphically as there are five variables p, b, D, S_c, V_c. If p, b are known, a graphical relation between D, V_c and S_c can be established (Figs. 9, 10, 11 and 11a).

Equation (9) was used by the author* for risering gray iron of carbon equivalents 3.5, 3.75 and 4 (silicon

*This experimental work was done by the author at Larson Foundry, Grafton, Ohio.

TABLE 4C – HIGH-CARBON NODULAR IRON DATA (C.E. = 4.63-4.81)

No.	Size	V_c	S_c	Min. Riser D	Min. Riser H	V_r	S_r	$\frac{V_c}{S_c}$	$\frac{V_r}{S_r}$	$\frac{V_r}{V_c}$	$\frac{S_r}{S_c}$	$X = \frac{V_r S_c}{S_r V_c}$	C.E.
1	4x4x4	64	96	5	1	20	35.4	0.67	0.57	0.31	0.37	0.85	H4.81
2	4x4x4	64	96	4¼	1¼	20	33.4	0.67	0.60	0.31	0.35	0.90	H4.81
3	4x4x13½	216	248	3½	6	58	75.4	0.87	0.77	0.27	0.30	0.90	4.63
4	4x4x13½	216	248	4	4	50	63.2	0.87	0.79	0.23	0.26	0.91	4.63
5	4x4x13½	216	248	4½	3¼	52	61.9	0.87	0.84	0.24	0.25	0.96	4.63
6	3x3x15	135	198	5	1¾	34	47.2	0.68	0.72	0.25	0.24	1.06	H4.81
7	3x3x15	135	198	5	1¾	34	47.2	0.68	0.70	0.25	0.24	1.06	H4.81
8	2x8x18	128	192	3½	2¾	26	39.8	0.67	0.65	0.20	0.21	0.97	4.62
9	2x10x10	200	280	5	1¾	34	47.2	0.71	0.72	0.17	0.17	1.01	4.62
10	2x10x10	200	280	4	2¼	28	41.3	0.71	0.68	0.14	0.15	0.95	4.62
11	2x6x18	216	308	6	2	56	65.9	0.70	0.85	0.26	0.32	1.21	L4.41
12	1x8x8	64	160	3	2¼	16	28.2	0.40	0.57	0.25	0.18	1.42	H4.81
13	1x8x8	64	160	3	2	14	25.9	0.40	0.54	0.22	0.16	1.35	H4.81
14	1x8x8	64	160	3	1½	11	21.2	0.40	0.52	0.17	0.13	1.29	H4.81
15	1x12x12	144	336	4	2½	31	44.4	0.43	0.70	0.22	0.13	1.63	4.49
16	1x12x12	144	336	3½	3¼	31	45.5	0.44	0.68	0.22	0.13	1.57	4.49
17	1x14x14	196	448	4	3	38	50.7	0.44	0.75	0.19	0.11	1.71	H4.81
18	6x6x6	216	216	4	4½	57	69.5	1	0.82	0.28	0.32	0.82	4.3
19	6x6x6	216	216	6	2	57	66	1	0.87	0.28	0.31	0.87	4.3

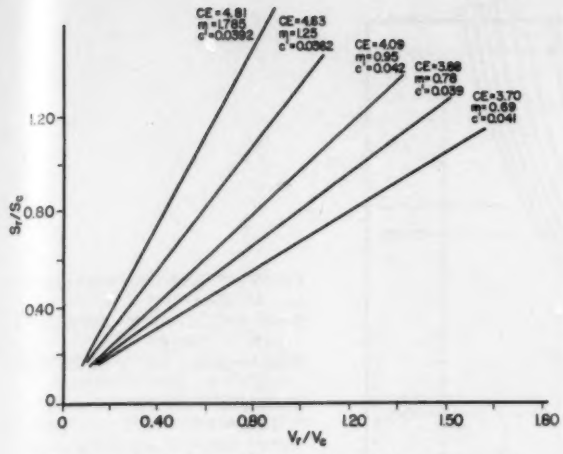


Fig. 7a — Scatter graph for nodular iron data ($S_r/S_c \rightarrowtail V_r/V_c$).

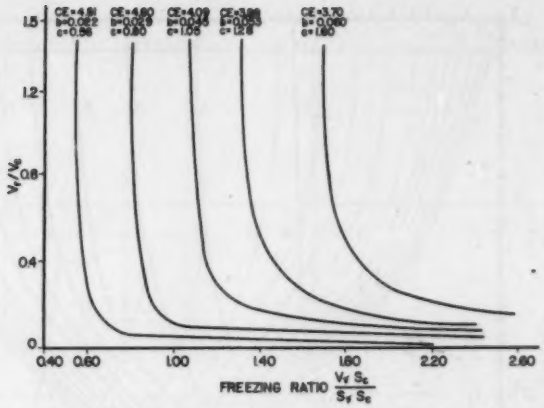


Fig. 7b — Scatter graph for nodular iron data (as Fig. 7a). The graphs were obtained by the least square method.

= 1.5 per cent). The results were quite satisfactory. The graphs obtained from equation (9), when D = H, for these grades of gray iron are plotted in Figs. 9, 10, 11 and 11a.

The Constants b and c

It is suggested that equation (9) may be used for any metal or alloy if its solidification shrinkage for pouring temperature, i.e., *b* is known. The solidification shrinkage for different metals can be found in the literature, or can be calculated by a number of formulas available. It is supposed that the liquid shrinkage does not play an important role in riser size for these reasons:

(1) Considerable amount of metal to compensate for the liquid shrinkage is supplied through the ingates before they freeze.

(2) The metal, when it reaches the riser, after passing through the gating system and the casting, is considerably colder than pouring temperature and much nearer to the liquidus temperature. The riser remains molten for more time just due to its shape, and not because it has a far higher temperature than the ingate before the liquidus temperature is reached.

(3) The solidification shrinkage increases with the superheat in some metals. This is particularly true when the superheat is excessive.[10,11] There is a high degree of variation in shrink depth in a riser due to superheat particularly when riser is insulated and risering is marginal.[11]

Actually, what we are interested in is that all the shrinkage, gross or microshrinkage, should be confined to the riser. Just adequate risers can only be obtained by a close control of the pouring temperature. It should not be below a certain minimum, nor should it be above a certain maximum.[11] In other words, a certain range of superheat is preferable for sound castings. The gross shrinkage often closely follows the superheat, but microshrinkage does not necessarily do so. The relation between superheat and shrinkage becomes quite confusing, particularly when a casting free from microshrinkage is required.[11] However, high-temperature gradients can reduce microshrinkage to a certain extent, while low-temperature gradients may transform gross shrinkage into microshrinkage. The methods, for finding solidification shrinkage of different grades of gray iron poured at 2500 F, and different grades of nodular iron poured at 2550 F, are described in the Appendix A.

TABLE 5—VALUES OF GENERAL EQUATION CONSTANTS

Metal	C.E.	m	c'	$b=\frac{c'}{m}$	$c=\frac{1}{m}$	$\frac{c}{b}=\frac{1}{c'}$	Comparative area of dend. mushy zone Ref. 7,8,9*
Nodular	3.7	0.69	0.041	0.060	1.44	24.0	9
Nodular	3.88	0.78	0.039	0.053	1.28	26.4	6
Nodular	4.09	0.95	0.042	0.045	1.05	24.3	—
Nodular	4.30	—	—	0.042	—	—	0
Nodular	4.63	1.43	0.041	0.029	0.70	24.7	—
Nodular	4.81	2.00	0.040	0.020	0.50	24.7	—
Steel	—	0.81	0.041	0.050	1.24	24.8	5.5
Gray Iron	3.5	—	—	0.033	—	—	9.75
Gray Iron	3.86	—	—	0.024	—	—	8.75
Gray Iron	4.13	—	—	0.015	—	—	4

*This is the comparative area of dendritic mushy zone between 2 and 3 in. from the interface.

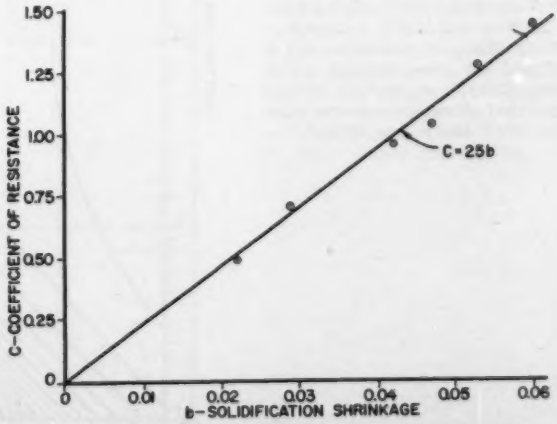


Fig. 8 — Graph showing relationship between coefficient of resistance against solidification shrinkage.

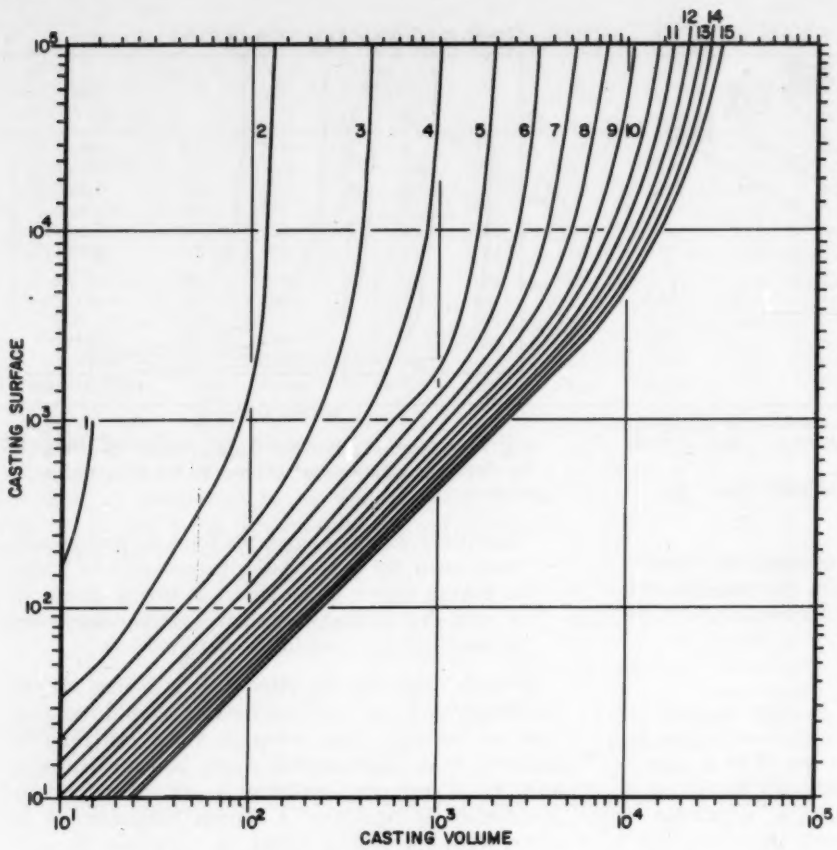


Fig. 9 — Graph for riser diameter, b = 0.05.
Steel: 0.6% C, 0.51% Mn, 0.3% Si. P.T. — 2800 F.
Nodular Iron: 3.17% C, 2.5% Si. C.E. — 4. P.T. — 2550 F.
Malleable Iron: 2.5% C, 1.0% Si. C.E. — 2.83. P.T. — 2550 F.
NOTE: Gray or malleable iron, if poured at temperatures higher than 2500 F, an increase of 0.25 per cent shrinkage must be made for 200 F excess over 2500 F.

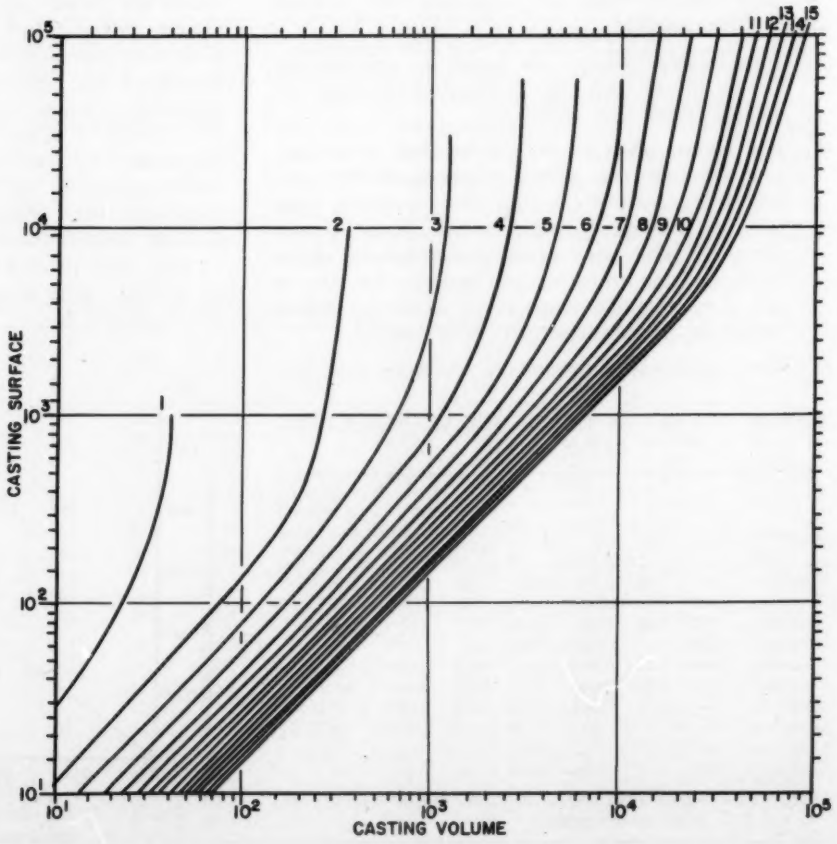


Fig. 10 — Graph for riser diameter, b = 0.0175, when D = H.
Gray Iron: 3.5% C, 1.5% Si. C.E. — 4. P.T. — 2500 F.
Nodular Iron: 4.09% C, 2.5% Si. C.E. — 4.92. P.T. — 2550 F.
NOTE: Gray or malleable iron, if poured at temperatures higher than 2500 F, an increase of 0.25 per cent shrinkage must be made for 200 F excess over 2550 F.

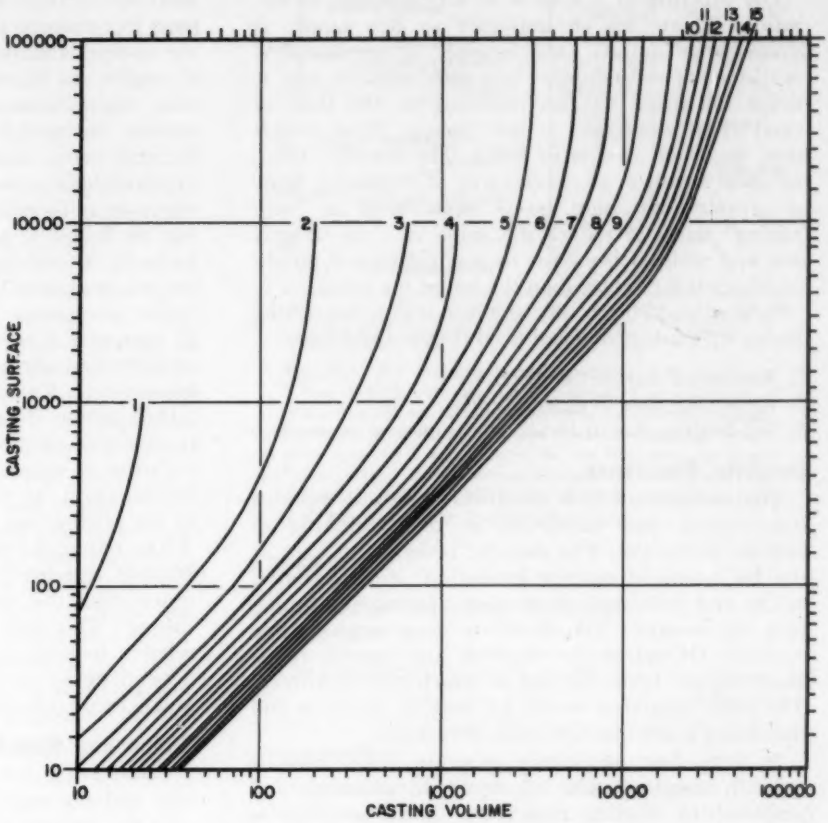


Fig. 11 — Graph for riser diameter, b = 0.033, when D = H.
Gray Iron: 3.0% C, 1.5% Si.
C.E. — 3.5. P.T. — 2500 F.
Nodular Iron: 3.7% C, 2.5% Si.
C.E. — 4.53. P.T. — 2550 F.
NOTE: Gray or malleable iron, if poured at temperatures higher than 2500 F, an increase of 0.25 per cent shrinkage must be made for 200 F excess over 2500 F.

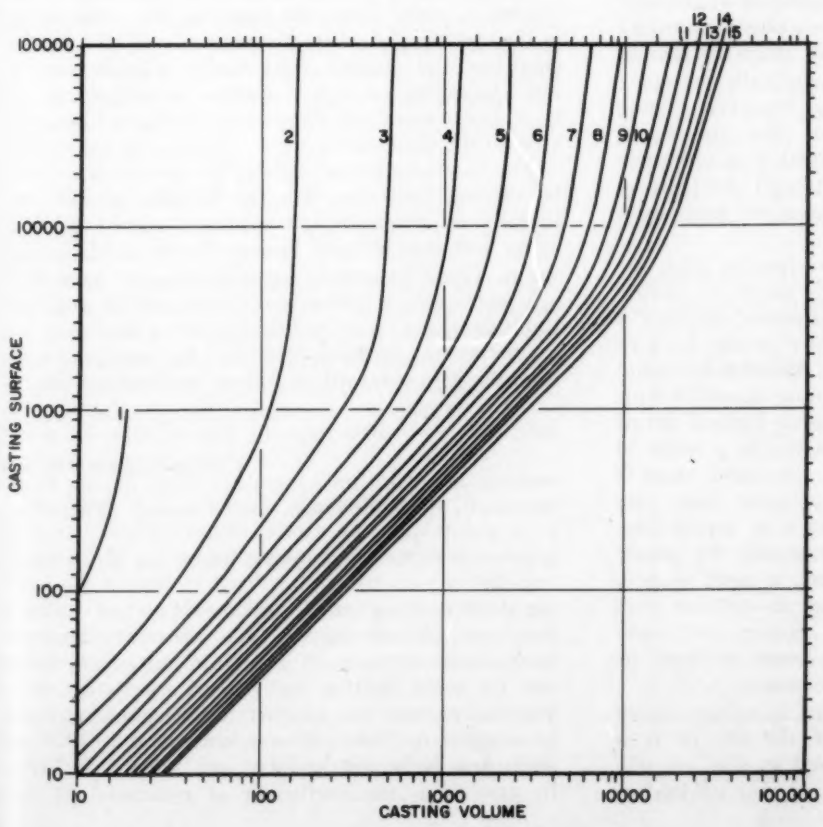


Fig. 11a — Graph for riser diameter b = 0.04, when D = H.
Nodular Iron: 3.37% C, 2.5% Si.
C.E. 4.2. P.T. — 2550 F.
Pure copper, (if b=0.04).
Gray Iron: 2.75% C, 1.5% Si.
C.E. — 3.25, P.T. — 2500 F.
NOTE: Gray or malleable iron, if poured at temperatures higher than 2500 F, an increase of 0.25 per cent shrinkage must be made for 200 F excess over 2500 F.

The function of a riser is to supply metal to the casting. There can be resistance to this supply if dendrites or eutectic are formed. If we say that coefficient of resistance for any such metal is one, it means the metal has no resistance to the flow of metal from the riser to the casting. Most metals form dendrites or eutectic when they solidify. Thus, for most metals *c*, i.e., coefficient of resistance, must be greater than one, except some kind of "self-feeding" tends to reduce the value of *c*, as in gray iron and nodular iron, due to precipitation of graphite, which is far lighter than the rest of the metal.

In gray iron, there can be three things happening during the passage of metal from riser to the casting:

1) Resistance due to dendrites.
2) Resistance due to eutectic cells.
3) Self-feeding due to localized volumetric expansion.

Dendrite Resistance

The resistance due to dendrites is zero at liquidus temperature, and maximum at the beginning of eutectic formation. The eutectic resistance is zero at the beginning of eutectic formation, and maximum at the end of eutectic formation. During the formation of eutectic, the dendritic resistance remains constant. Of course, the dendritic and eutectic resistances depend upon the way in which each is formed. The total resistance would be zero at liquidus, and maximum at the end of eutectic formation.

In gray iron, ordinarily graphite arises from a eutectic transformation of liquid to austenite and graphite. In mottled iron some of the graphite is formed indirectly by the decomposition of a carbide after solidification. Type D undercooled graphite may be formed by one of these two possible ways.[12] The growth and three-dimensional shape of normal and undercooled graphite is fundamentally the same, i.e., "from the common center by branched crystallization interconnected";[12] while the growth of graphite obtained by the decomposition of cementite is by the diffusion of graphite through the layer of austenite, and graphite and the flakes are supposedly separated from one another.

In nodular iron, the carbon precipitates from the unstable undercooled liquid as a cloud of minute platelets which, after being transformed through a phase of liquid crystal, agglomerate in the form of nodules due to high surface energy between the liquid crystal of graphite and the remaining liquid.[13] This is called the primary graphite, and is formed out of liquid. The secondary graphite occurs as a result of decomposition of cementite in the solid state.[13] Thus, in gray iron, the passage of metal from riser to metal during eutectic formation is by liquid flow, while in nodular iron it is presumably by plastic flow. Hence, the eutectic resistance is more in nodular iron than in gray iron. Also, in nodular iron, under the conditions of rapid cooling, the interdendritic feed channels may become blocked by carbide, increasing the eutectic resistance.[14,15]

In gray iron most of the graphite is formed out of liquid and interconnected, while the rest of it is formed in solid form and separated so that less self-feeding is possible. It must be noted that self-feeding does not actually reduce the resistance to metal passage, but reduces the quantity of metal required by the casting, and thus superficially reduces the value of coefficient of resistance (c), sometimes even less than unit. Sometimes, heavy dendrites altogether prevent the passage of metal to or from the casting. In such cases, inadequate or no self-feeding results in unavoidable interdendritic shrinkage, and over-adequate self-feeding results in swells.[16] Such swells may be found in gray or nodular iron, though more frequent in nodular iron, as it is unable to relieve the self-feeding by liquid flow as does gray iron.

The constant *c* in equation (3) is the coefficient of resistance discussed above, and constant *b* is the solidification shrinkage of the metal at the pouring temperature. Caine[3] determines the constants *b* and *c* in equation (5) thus: "If the riser freezes infinitely slowly in relation to the casting, the riser volume required is equal to the volume of contraction on solidification. If the riser freezes at the same rate as the casting, the riser volume required is infinity." Then, citing the Chvorinov's[1] rule, he deduces that relative freezing time to complete solidification is unity when the riser freezes at the same rate as the casting. This may not always happen, because the relative freezing time to complete solidification also depends upon the resistance offered by the dendrites and eutectic cells formed during solidification.

Shape and Resistance Factors

Considering the shape factor in the Chvorinov's rule and the resistance offered, it may be concluded that the relative freezing time to complete solidification, when the riser freezes at the same rate as the casting, is unity when the riser has the same shape as the casting, and when the resistance offered by the dendrites and eutectic cells during solidification is zero. As explained before, Caine's equation (equ. 5) is a restatement of Chvorinov's rule taking into account the shape factor.

The resistance factor can also be taken into account by varying the value of *c* for different metals and foundry conditions as the resistance offered by dendrites and eutectic cells formed during solidification varies. Thus, the general equation for riser (equ. 9) is a restatement of Chvorinov's rule taking into account the shape factor and the resistance factor.

Steel freezes forming dendrites. For steel $c = 1.25$, from the data obtained by Bishop and others.[4] Therefore, $b = \frac{1.25}{25} = 0.05$ (equ. 6). For an alloy or metal forming no dendrites or eutectic cells during freezing, $c = 1$ (but no such metal exists). Therefore, $b = \frac{1}{25} = 0.04$. The values *b* and *c* for alloys forming dendrites, e.g., steel, and for gray and nodular iron are plotted against the area of dendritic mushy zone between 2 and 3 in. form the interface (to avoid chilling due to end effect), the areas being taken from the solidification diagram obtained by workers at Naval Research Laboratory,[7,8,9] for steel, gray iron and nodular iron (Figs. 12a, 12b). In gray iron, the coefficient of resistance (c) and

solidification shrinkage (b) are consistently less than those for dendritic alloys. The curves for gray iron, when produced, intersect the dendritic alloys curves at c = 1.55 and b = 0.062 which roughly corresponds to carbon = 2.3 per cent and silicon = 1.5 per cent (Figs. 12c, 12d). From the Maurer[17] diagram, we find this is mottled iron.

The curves of nodular iron intersect the dendritic alloys curve at c = 1.25, b = 0.05 which roughly corresponds to carbon = 3.17 per cent, silicon = 2.5 per cent or C.E. = 4. This means that at the point of intersection, gray iron (C.E. = 2.8) and nodular iron (C.E. = 4) have self-feeding just enough to compensate for the eutectic resistance to the passage of metal from the riser to the casting. On the right of the point, self-feeding is insufficient to compensate for the eutectic resistance completely; on the left of the point, self-feeding compensates all the eutectic resistance and a part of dendritic resistance.

The values of *b* and *c*, for gray and nodular iron, are plotted against carbon equivalent and percentage graphite by weight (Figs. 12c, 12d, 12e, 12f). It is evident that for a given carbon equivalent, the values of *b* and *c* are higher for nodular iron; but for a given per cent graphite, the values of *b* and *c* are higher for gray iron. This means that gray iron will precipitate more graphite, for a fixed carbon content than nodular iron, but the graphite in the nodule form has far more effective self-feeding than the flake graphite.

CONCLUSION

In this study, an analytic method is formulated to determine size and shape of risers. It may be used for any metal or alloy under certain restrictions and assumptions. It does not answer the questions of how many and where, but it does answer the questions what diameter and how high using the general equation (10),

$$\frac{25(4p+1)}{S_e} = \frac{p \cdot D}{b \cdot V_e} - \frac{1.275}{nD^2} \qquad (10)$$

Also, an attempt was made to explain the meaning of the constants *a*, *b* and *c* in Caine's equation for steel risers. A relationship was found between these constants and the equation restated in terms of constant *b*, which is solidification shrinkage of the metal at pouring temperature.

ACKNOWLEDGMENTS

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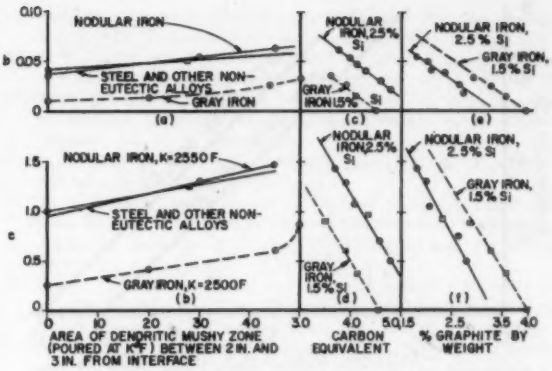


Fig. 12 — Relationship between mushy zone, constants *b* and *c*, C.E. and per cent graphite. The values of *b* and *c* for alloys forming dendrites (steel and nodular iron) are plotted against the area of dendritic mushy zone between 2 and 3 in. from the interface, the areas being taken from reference data[7,8,9] for steel and nodular iron (a and b). *c*, *d*, *e* and *f* are plotted against C.E. and percentage graphite by weight.

APPENDIX A

Solidification Shrinkage in Gray and Nodular Iron

The solidification shrinkage for any composition of gray iron poured at 2500 F, and nodular iron poured at 2550 F, can be found readily by the following methods:

Nodular Iron:[6] Figure 13 shows a curve of solubility of carbon in austenite of 2100 F. Carbon and silicon percentages of the metal being known, the percentage graphite by weight can be determined. Per cent graphite by volume is per cent graphite by weight divided by 0.3.

Figure 14 shows relation between per cent graphite by volume and solidification shrinkage of nodular iron, poured at 2550 F, for green and dry sand molds. It should be noted that there is a difference of about 1½ per cent shrinkage in green and dry sand molds. This is mainly due to mold wall movement.

Gray Iron:[18] The solidification shrinkage for gray iron poured at 2500 F in dry sand molds is given by,

$$\text{Total Shrinkage} = 200 = -\frac{100}{1+16.5\times10^{-6}\times\text{F.R.}}\left(\frac{100-3.9}{100-Y}\right)[3.3(C-Y)+(100-C)]$$

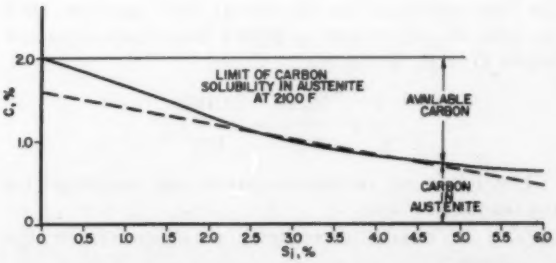


Fig. 13 — Solubility of carbon in austenite curve at 2100 F, from reference data.[6]

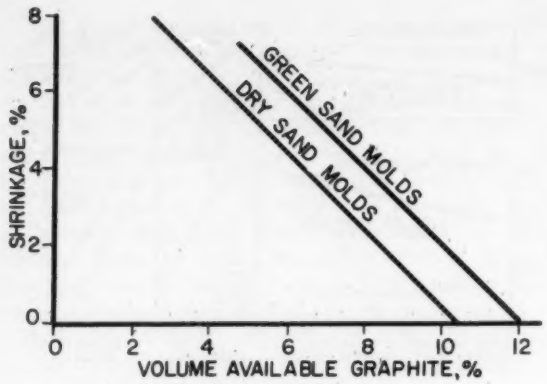


Fig. 14 — Comparison of shrinkage in green and dry sand molds for different amounts of available graphite, from reference data.[6]

where:

F.R. = freezing range, i.e., liquidus to eutectic temperature.

16.5×10^{-6} = average volumetric coefficient of expansion for liquid metal and austenite.

3.9 = Percentage shrinkage associated with the solidification of austenite.

C = per cent carbon.

Y = per cent carbon soluble in austenite.

3.3 = ratio of density of austenite to the density of graphite.

4.3 = per cent carbon in eutectic iron carbon alloy.

Schmidt and Taylor[10] found the difference in shrinkage in green and dry sand molds about 1.5 per cent.

The freezing range (F.R.) can be found in the formula by the following equations:

Freezing range F = Liquidus temp. F − 2050

where, 2050 = average eutectic temp. (F) and liquidus temp. F = 2981−218 F,[19] where, F = %C+½%P+¼%Si.

APPENDIX B

A Proposed Method for Risering Casting of Different Metals

From the solidification diagrams for different metals obtained by the research workers at Naval Research Laboratory, [7,8,9,15,20,21,22] and using the Fig. 12a, the values of solidification shrinkage for different metal are obtained and tabulated in Table 6.

Substituting the value of b in equation (9) gives the riser equation for any metal. For example, 99.9 per cent copper poured at 2050 F has a riser equation (when D = H, b = 0.04),

$$\frac{1}{S_c} = \frac{0.2D}{V_c} - \frac{0.0102}{D^2}$$

The following recommendations are suggested for top-insulated risers:

1) Try to obtain directional solidification in the casting.*
2) Place the riser in thickest section of the casting.
3) Neckdown the riser for easy removal from the casting and to avoid hot-spot at the junction of casting and riser.
4) Establish the proper H/D ratio feasible so as to need minimum riser weight.
5) Insulate all the risers at the top.
6) Use exothermic compounds in the riser for castings above 500 lb. A few dollars worth of exothermic compound can save the whole casting worth hundreds of dollars.
7) H/D ratio for light metal alloy should be more than one to obtain a proper hydrostatic pressure for the flow of metal from the riser to casting.
8) Do not use exothermic compounds in the metals where the extra heat generated by the compound promoted oxidation and drossing of the metal. Such metals are aluminum and magnesium alloys.

TABLE 6 – SOLIDIFICATION SHRINKAGES FOR DIFFERENT METALS

Metal Composition, %	Solid Shrinkage, %	Remarks
Aluminum 99	5.7	
0.08 C Steel	4.5	0.5% Si, 0.7% Mn, P.T.=2900 F
0.30 C Steel	4.7	0.5% Si, 0.7% Mn, P.T.=2900 F
0.60 C Steel	5	0.5% Si, 0.7% Mn, P.T.=2900 F
Copper 99.9	4	P.T.=2050 F
8.5 Mg - Al	6.2	P.T.=1200 F
4.5 Cu - Al	6.5	0.22% Si, P.T.=1250 F
Bronze (88-10-2)	6.5	P.T.=1900 F
Brass (60-40)	4.5	0.63% Sn, 0.51% Mn, 0.70%, P.T.=1779 F
Stainless Steel (12Cr)	0.13%	C, 2.7% Ni, P.T.=2950 F
Stainless Steel (18-8)	0.20%	C, P.T.=2800 F
Monel	0.14%	C, 6.2% Fe, 0.65% Si, P.T.=2500 F
Lead	0.5%	Cu, 0.5% Sn, P.T.=650 F

REFERENCES

1. N. Chvorinov, "Theory of the Solidification of Castings," *Giesseri, vol. 27*, p. 177 (1940); British Iron and Steel Institute translation No. 117.
2. H. F. Bishop and W. S. Pellini, "The Contribution of Riser and Chill-edge Effects to Soundness of Cast Steel Plates," AFS TRANSACTIONS, *vol. 58*, p. 185 (1950).
3. J. B. Caine, "A Theoretical Approach to the Problem of Dimensioning Risers," AFS TRANSACTIONS, *vol. 57*, p. 492 (1948).
4. H. F. Bishop, E. T. Myskowski and W. S. Pellini, "A Simplified Method for Determining Riser Dimensions," AFS TRANSACTIONS, *vol. 63*, p. 724 (1955).
5. H. F. Bishop and C. G. Ackerlind, "Dimensioning of Risers for Nodular Iron Castings," Naval Research Laboratory Report 4737, 19 pp. (April 27, 1956).
6. C. Reynolds, J. Maitre and H. Taylor, "Feed Metal Requirements for Ductile Iron Castings," AFS TRANSACTIONS, *vol. 65*, p. 386 (1957).
7. H. F. Bishop, F. A. Brandt and W. S. Pellini, "Solidification of Steel from Sand and Chill Walls," AFS TRANSACTIONS, *vol. 59*, p. 435 (1951).
8. R. P. Dunphy and W. S. Pellini, "Solidification of Gray Iron in Sand Molds," AFS TRANSACTIONS, *vol. 59*, p. 425 (1951).
9. R. P. Dunphy and W. S. Pellini, "Solidification of Nodular Iron in Sand Molds," AFS TRANSACTIONS, *vol. 60*, p. 783 (1952).
10. W. A. Schmidt and H. F. Taylor, "Risering of Gray Iron Castings," Progress Report No. 4, AFS TRANSACTIONS, *vol. 61*, p. 131 (1953).

*Directional solidification does not necessarily assure soundness. Depending on the geometery of the casting, a minimum gradient condition in excess of that required for directional solidification may be necessary for feeding to soundness,[10,24] e.g. 6 F per in. of gradient for bars[24] and 1 F per in. for plates.[2]

11. R. C. Schnay and S. L. Gertsman, "Risering of Nodular Irons, Part 3 — The Effect of Pouring Temperature on Shrink Depth," AFS TRANSACTIONS, *vol. 64*, p. 271 (1956).
12. H. Morrogh, "Graphite Formation in Cast Iron," British Cast Iron Research Assoc., *Journal of Research and Development, vol. 5*, No. 12 (June, 1955).
13. P. H. Anderson, "Theory of Nodular Graphite Formation in Ductile Cast Iron," AFS TRANSACTIONS, *vol. 63*, p. 163 (1955).
14. H. Morrogh, "The Solidification of Nodular Iron," *Journal of Iron and Steel Institute, vol. 176* (IV), p. 378 (April 1954).
15. R. P. Dunphy, C. G. Ackerlind and W. S. Pellini, "Solidification, Feeding Characteristics of Gray and Nodular Irons," *Foundry, vol. 82*, No. 6, p. 108 and 249 (June 1954).
16. R. C. Schnay and S. L. Gertsman, "Risering of Nodular Iron, Part 2 — Effect of Silicon Content on Feeding Distance," AFS TRANSACTIONS, *vol. 63*, p. 632 (1955).
17. E. Maurer, *Kruppsche Monatshefte, vol. 5*, p. 115 (1924).
18. E. J. Sullivan, Jr., C. M. Adams and H. F. Taylor, "Risering of Gray Iron Castings," Research Committee, Gray Iron Division, American Foundrymen's Society, Progress Report No. 6, AFS TRANSACTIONS, *vol. 63*, p. 8 (1955).
19. L. F. Porter and P. C. Rosenthal, "Factors Affecting Fluidity of Gray Cast Iron," AFS TRANSACTIONS, *vol. 60*, p. 725 (1952).
20. F. A. Brandt, H. F. Bishop and W. S. Pellini, "Solidification of Various Metals in Sand and Chill Molds," AFS TRANSACTIONS, *vol. 62*, p. 646 (1954).
21. H. F. Bishop and W. S. Pellini, "Solidification of Metals," *Foundry*, p. 86, (Feb. 1952).
22. W. S. Pellini, "Solidification of Various Metals in Ingot and Sand Molds," submitted in Electric Furnace Steel Conference, American Institute of Mining, Metallurgical and Petroleum Engineers, Chicago (Dec. 5-7, 1956).
23. R. A. Flinn, D. J. Reese and W. A. Spindler, "Risering of Ductile Cast Iron — Riser Dimensions, Feeding Distance and Other Data," AFS TRANSACTIONS, *vol. 63*, p. 720 (1955).
24. H. F. Bishop, E. T. Myskowski and W. S. Pellini, "Contribution of Riser and Casting End Effects to the Soundness of Steel Bars," AFS TRANSACTIONS, *vol. 59*, p. 171 (1951).

MOLD MATERIAL EFFECT ON COOLING RATE AND PHYSICAL PROPERTIES OF CAST METALS

Second Report of Sub-committee T.S.46
of the Technical Council, Institute of British Foundrymen

ABSTRACT

The investigation started ten years ago as a result of some members of the technical council doubting the efficacy of the craftsman's practice of using raw red sand on bosses and junctions of thick and thin sections to prevent sinking in gray iron castings. This posed the question, "If it was effective in preventing sinking, why was it effective?" Basically, this was thought to be tantamount to asking "What is the effect of mold material on the solidification rate of cast metals?"

The original work was carried out with the interrupted-freezing or slush-casting method. This proved to be insufficiently sensitive and the investigation was then pursued with a pyrometric method. When certain precautions are taken this method is satisfactory and the general conclusions are that the coarser the sand the greater its cooling power and the harder a mold is rammed the greater its cooling power, other variables being equal.

The third stage of the work and the subject of the present report was to study the effect on the physical properties of various cast irons when cast into molds made in typical molding sands at various ramming densities, and also the effect when using other types of molding materials of varying thermal properties. The effect of inoculation and changing the volume to surface area ratio of the test bars was also investigated. Finally, a series of practical tests was carried out in four foundries, each engaged in different classes of work and using standard materials available in each foundry.

The results of the tests confirm the previous experimental work and show that increased coarseness of grain size and increased mold density increases tensile strength in all cases, when other variables are equal. The exception to this was found when the test casting was not of sufficient volume to avoid being mottled when the cooling power of the mold material was increased.

The conclusions reached in this report may assist in elucidating certain previously unexplained phenomena encountered in foundries. The report also sets out some variables which must be controlled when investigations into cooling rates of castings are being carried out.

The sub-committee T. S. 46 of the Institute of British Foundrymen is at present constituted as follows:

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PREVIOUS WORK BY THE COMMITTEE

Work started on this problem in March, 1947, the original investigation being carried out with the interrupted freezing or slush casting method, the idea being to measure the speed of movement of the solid-liquid metal interface away from the mold wall by slushing the casting a predetermined time after casting. This method proved to be insufficiently sensitive, as in all but pure metals and alloys freezing at a constant temperature, there exists a zone of partial solidification which produces a ragged surface to the shell formed in the mold after slushing. It was therefore not possible to determine the difference in cooling power of the different types of sand due to the existence of this zone of partial solidification between the all-solid and all-liquid regions.

Experiments were then carried out with a pyrometric method which consists of placing one or more thermo-couples in selected positions in the mold cavity, or molding sand, or both. Temperature changes were recorded with a sensitive temperature recorder from the moment of pouring the metal into the mold until solidification was complete. It was found that the results obtained had poor reproducibility. Further investigation showed that this was due to two main factors, one of which was the paramount importance of closely controlled mold density, and the other the fact that the cast iron used had a long freezing range and it was not possible to determine the exact end point of solidification.

An investigation was then carried out to find a satisfactory method of obtaining a mold of uniform density throughout its depth so that its density figure was correct at all positions in the mold and not an average of soft and hard areas. The method finally adopted was called the three-squeeze method and is described in the sub-committee's first report.[1]

To overcome the other cause of failure in reproducibility aluminium-bronze was used as the casting alloy. The tests showed that when poured at 1250 C

this alloy had approximately the same solidification time as cast iron.

The conclusion[1] drawn from the investigation into the suitability of the pyrometric method for studying the effect of sand mold materials on the solidification rate of cast metals was that it is satisfactory provided the following precautions are taken:

a) The casting must be sufficiently large to give a reasonable solidification time.
b) Sand bulk density and moisture content must be accurately controlled.
c) Gating and pouring temperature, including superheating temperature of metal, must be standardized.
d) Casting flash must be eliminated.

The sub-committee then carried out work on the effect of mold material on the cooling rate of cast metals using the pyrometric method and observing the foregoing precautions. From this work the following main conclusions were drawn:

1) The cooling power of the sand mold increases with increased ramming density for all sands investigated.
2) The coarser the grade of sand the greater its cooling power.
3) The cooling power of green sand is greater than that of dry sand and the addition of coal dust to either further increases the chilling power but this is more pronounced at low ramming densities.

WORK PROGRAM COVERED BY PRESENT REPORT

The salient feature to emerge from the discussion on the first report by the sub-committee was a need to produce some practical evidence of the physical effect on castings of varying the mold material and the density of the mold. In consequence, the technical council gave the sub-committee the following extended terms of reference:

> "To examine the effect of molding materials on the cooling rate of cast iron, and the resulting influence on physical properties."

It was agreed that the work should, in addition to the normal sands, include the following as mold materials: Shell molding materials; silicon carbide; oil sand; graphite; sodium-silicate bonded sand; zircon; olivine sand; and cement-bonded sand. It was also agreed that piston rings, valve guides and bars ½-in. to 1¼-in. dia. and 8 in. long be used as test castings.

The work covered in the present report falls into two groups as follow:

a) The work carried out by the Department of Industrial Metallurgy of Birmingham University under closely-controlled conditions, and
b) The work done by members of the sub-committee in their own foundries to ascertain if the results obtained at the University could be reproduced under production conditions.

Work Carried Out at University

This work consisted of tests using the following molding materials:

1) Natural-bonded Bromsgrove red sand with and without coal dust.
2) Coarse and fine grain size silica sands bonded with bentonite, cement, linseed oil and dextrine, and sodium-silicate.
3) Zirconium silicate and olivine sand bonded with bentonite.
4) Two grades of graphite each bonded with linseed oil and dextrine and sodium-silicate.
5) Shell process molds in Redhill 'H' and Leighton Buzzard sands.
6) Silicon-carbide bonded with sodium-silicate or with bentonite and subsequently vitrified.

Eight per cent coal dust was added to the bentonite-bonded sands to improve casting finish. The grades used are listed in Table 2.

Round test bar castings of various diameters were made in each of the mold materials described in the foregoing. The mold and gating systems are shown in Fig. 1. The bars were cast ½-in., ¾-in., 1-in., and

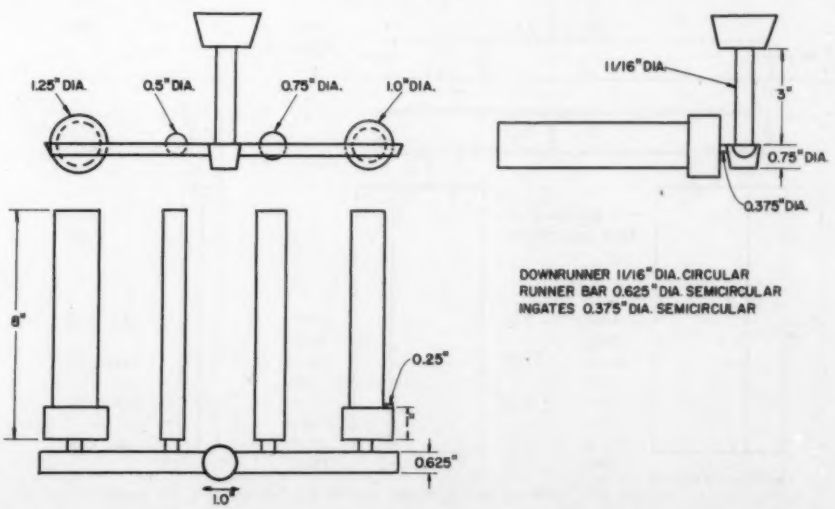


Fig. 1 — Mold layout and gating system as used at the University (results given in Table 2).

TABLE 1 — ANALYSIS OF IRONS OF VARIOUS CARBON EQUIVALENTS USED AT THE UNIVERSITY
(analysis of all the melts was within the following limits)

Iron		Composition, %									
Mark	Nominal carbon equivalent	T.C.	±	Si	±	P	±	Mn	±	S
J.	4.2	3.30	0.04	2.44	0.04	0.21	0.03	1.40	0.03	0.01
K.	4.5	3.51		2.96		0.20		1.38		0.01
L.	3.7	2.93		2.16		0.19		1.37		0.01
M.	3.7*	2.89		2.25		0.19				

*Inoculated with calcium silicide

1¼-in. dia. and 8 in. long; the gating system was of the choked type. Preliminary experiments showed that no feeders were necessary to produce sound ½-in. and ¾-in. bars, but small feeders were placed at the ingate ends of the 1-in. and 1¼-in. bars as shown in Fig. 1.

A test-bar casting was also designed with four square bars having the same volumes as the 1¼-in., 1-in., ¾-in. and ½-in. dia. bars. The length of these bars was 8 in. and the lengths of the sides 1.11, 0.89, 0.67, and 0.44 in. The square test-bar castings were machined to standard test-bar sizes for tensile strength testing.

The base iron used was refined pig iron containing C, 3.35%; Si, 2.22%; Mn, 1.3%; P, 0.3%; and sulfur 0.01%. This iron had a C.E. (carbon equivalent) of 4.2 (iron 'J') and to produce a nominal C.E. of 4.5 (iron 'K') an addition of 0.5 per cent silicon-carbide was made to the base iron. For the melt of nominal C.E. 3.7 (iron 'L'), an addition of 15 per cent of mild steel was made.

All the irons were superheated to 1450 C. The 4.2 C.E. iron 'J' was poured at 1350 C and the 3.7 and 4.5 C.E. irons 'K' and 'L' at 1375 C. High-frequency induction melting was used throughout and the temperatures measured by means of Pt-Pt.Rh couple and a direct-reading instrument.

The effect of carbon equivalent was investigated by making the round bar castings in Redhill 'H', Redhill $1\frac{1}{28}$ and Bromsgrove sands with irons 'J', 'K' and 'L' of carbon equivalents of 4.56, 4.2, and 3.71, respectively. All melts were analyzed and the results are detailed in Table 1. The effect of inoculation with 0.5 per cent ferro-silicon,* 0.25 per cent of silicon carbide, and 0.25 per cent of calcium silicide, respectively, was studied with the iron of C.E. 3.6 (iron 'M').

Tests using Redhill 'H', Redhill $1\frac{1}{28}$ and Congleton sand molds were also made with square bars to obtain data on the effects of the different sands on bars with different volume-to-surface-area ratios.

The work carried out in members' foundries as described in the following was also duplicated at the University to act as a check on the reproducibility of results.

All the various bars produced were machined to corresponding standard test bars and tensile tested. Brinell hardness readings were taken across the diameter of the 1¼-in. bars cast in coarse synthetic sand, natural sand, and zirconium-silicate molds bonded with bentonite.

In addition to the tensile testing of all bars, microexamination was carried out on transverse sections adjacent to the tensile fractures.

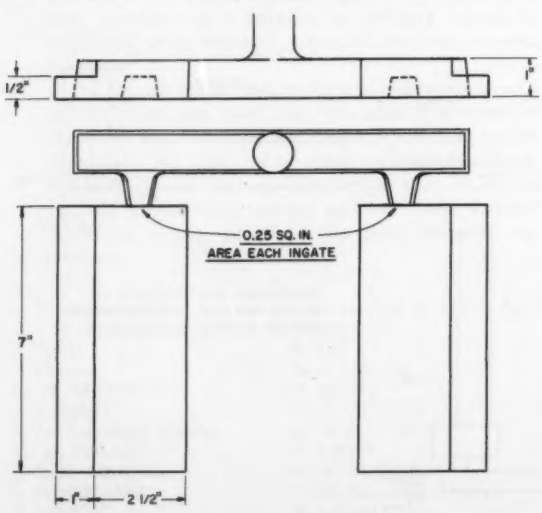
*Ferro-silicon 75 per cent silicon.

TESTS CARRIED OUT BY MEMBERS

Stepped Bar Castings

This work consisted of making the castings shown in Fig. 2 in four different foundries and at the University. The pattern and runner system were identical in each case and each foundry made two pairs of castings using coarse sand hard-rammed and fine sand soft-rammed. Each investigator used the coarsest sand and the finest sand available in his foundry, except that at the University the original sands already referred to were used and at one foundry only one type of natural sand was available.

In this latter case the same type of sand was used in each mold, one mold being rammed hard and one soft. All molds were hand-rammed, and the sand contained 8 per cent coal dust with a moisture percentage to suit molding conditions. The molds were made in a 4-in. deep molding box and cast in the typical gray



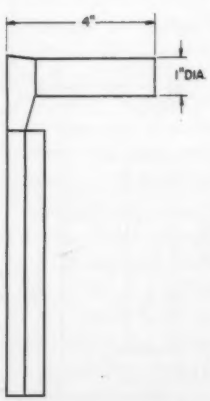


Fig. 2 — Step castings and runner system (results in Table 4).

iron available in each foundry. Two test bars, one from each outside edge of each test casting, were machined from the test block. The casting temperatures were read with an immersion pyrometer by the same investigator who visited each foundry when the castings were being made.

To supplement the tensile figures obtained from the test bars cast at the foundry which used only one type of natural sand, further work was carried out by two other member foundries who repeated their tests, one using Congleton base synthetic sand and the other Bromsgrove natural sand. In these tests the only variable was ramming density, each foundry making two test castings, one rammed hard and one soft. Similar tests were repeated at the University using one sand and extremes of ramming density.

TABLE 2 – DETAILS OF MOLD AND MOLD MATERIALS AND TENSILE STRENGTHS OBTAINED ON ROUND TEST BARS WITH IRON OF C.E. 4.2 MARK 'J' USED AT THE UNIVERSITY

Mold Material	Grain Size	Density g/cc	Bond	Moisture, %	Drying treatment	Coal dust	Tensile strengths, tons per sq. in. Bar diameters, in. 1¼	1	¾	½
Leighton Buzzard(C)	Coarse	1.95	10% cement	10			15.9	18.5	21.2	mottled
Zirconium silicate	Very fine	2.61	3½% bentonite	3½		8% fine	15.7	18.2	20.8	24.8
Olivine	Coarse – medium	1.82	3½% bentonite	4		8% med.	15.6	17.7	20.0	24.4
Leighton Buzzard(C)	Coarse	1.72	Oil[a]		8 hr 200 C		15.6	18.0	20.8	mottled
Redhill 14/28	Coarse	1.62	3½% bentonite	3		8% Coarse	15.5	17.6	20.4	25.0
Redhill H	Fine	1.70	10% cement	10			15.2	17.5	20.0	24.4
Redhill H	Fine	1.60	Oil[a]		8 hr 200 C		15.1	17.2	19.8	23.5
Leighton Buzzard(C)	Coarse	1.65	CO_2 5%				15.0	17.4	20.2	mottled
Bromsgrove black	Medium	1.52	clay approx. 11%	7		8% med.	14.9	16.7	19.9	24.0
Redhill H	Fine	1.56	CO_2 5%				14.6	16.5	19.0	23.0
Leighton Buzzard (22)	Fine	Shell 0.5 in. thick	6% resin				14.5	16.0	18.2	22.0
Redhill H	Fine	1.31	3½% bentonite	4		8% fine	14.4	16.1	18.9	23.0
Redhill H	Fine	Shell 0.4 in. thick	6% resin				13.7	15.3	17.7	21.5
Graphite B.B.5. coarse	50% thru 35 onto 65 mesh	1.06	oil[b]		8 hr 200 C		15.0	19.9		
Graphite G.10 fine	90% thru 100 55% thru 200 mesh	1.06	oil[b]		8 hr 200 C		15.1	20.1		
Graphite B.B.5. coarse	50% thru 35 onto 65 mesh	1.06	CO_2 10%				14.5	18.8		
Graphite G.10 fine	90% thru 100 55% thru 200 mesh	1.06	CO_2 10%				14.8	19.2		
Silicon carbide	Medium	1.90	2% clay	2	vitrified 1300		16.9	mottled		
Silicon carbide	Coarse	1.90	2% clay	2	vitrified 1300		16.8	mottled		
Silicon carbide	Medium	1.90	2% clay	2	8 hr 200 C		15.9	mottled		
Silicon carbide	Coarse	1.90	2% clay	2	8 hr 200 C		15.8	mottled		

[a]1% linseed oil, 1½% starch. [b]4% linseed oil, 2% starch. All values the mean of at least two results.

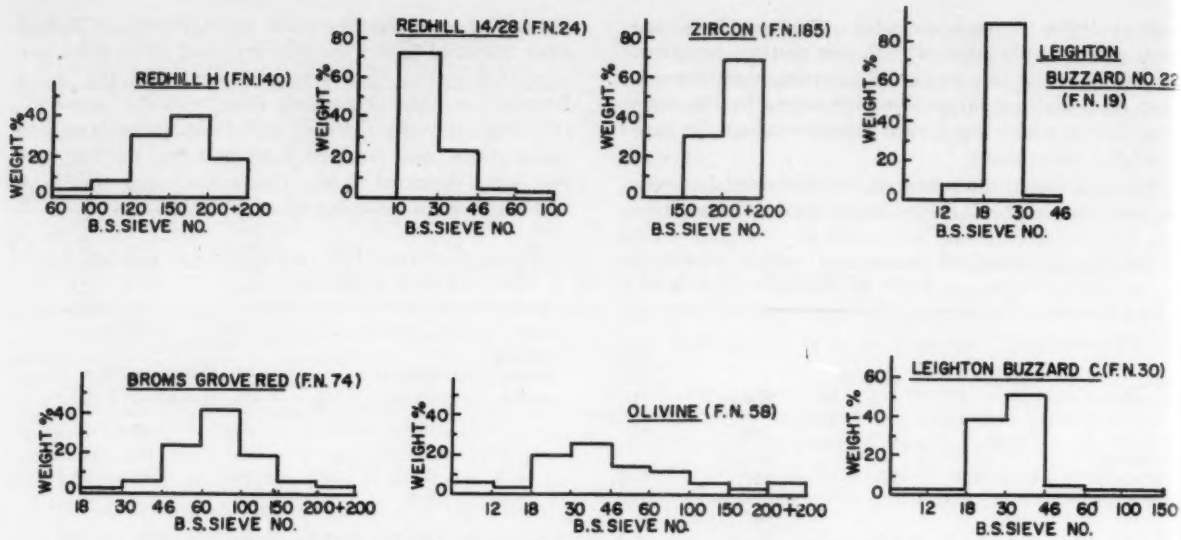


Fig. 3 — Gradings and AFS fineness numbers of the various sands used for work recorded in Table 2 (typical sand analyses are given in Table 3 of the first report[1]).

Piston Rings and Valve Guides

A series of tests was also made by a member of the sub-committee using valve guides and piston rings as test castings to determine the effect of various mold materials upon the physical properties and microstructure of the castings. The tests were carried out under standard production conditions and standard ramming densities.

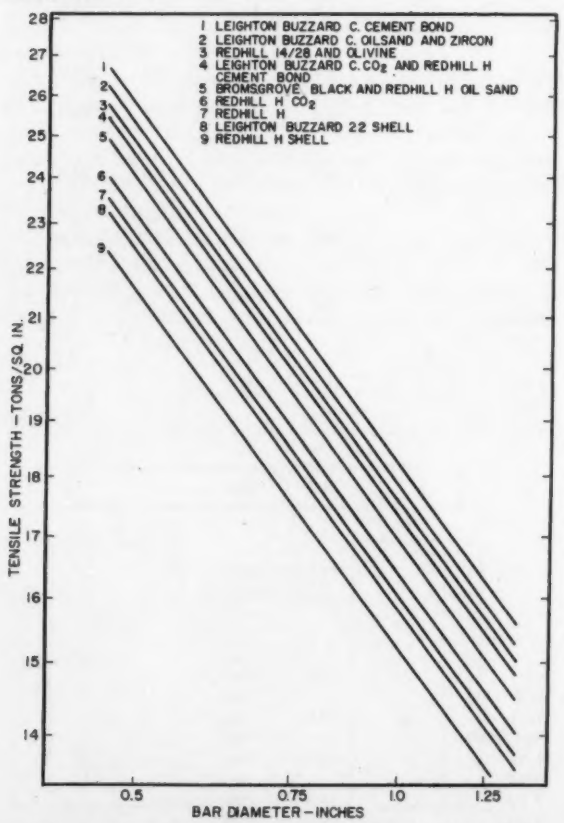


Fig. 4 — Tensile results of iron J (C.E. = 4.2) plotted against bar diameter.

EXPERIMENTAL RESULTS

Program of Work at the University

Sand molds. The results are set out in tabular form in Table 2 with details of the various sands and other molding materials used. The gradings of the sands are given in Fig. 3. The variation of tensile strength with mold material of the iron 'J' of C.E. 4.2 is shown in Fig. 4. The results are plotted in logarithmic coordinates and a common slope selected such that when lines of this slope were passed through the experimental points there was the least deviation.

The actual tensile values obtained are given in Table 2. The minimum tensile properties were obtained in the bars cast into Redhill 'H' shell molds, the maximum properties, approximately 20 per cent greater, in the bars cast into cement-bonded Leighton Buzzard 'C' sand. The bars produced were gray with the exception of those of ½-in. dia. cast into Leighton Buzzard cement- and oil-bonded sands. Figure 5 sets out the Brinell numbers obtained from zircon sand, Bromsgrove black and Redhill 'H' in graph form, and Fig. 6 shows the tensile stength and Brinell hard-

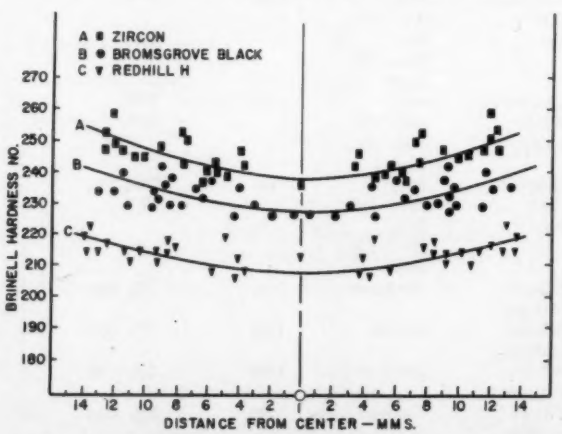


Fig. 5 — Brinell hardness numbers of iron J (C.E. = 4.2) bar diameter 1¼-in. As cast.

ness of the bars cast into Redhill 'H' and zircon sands bonded with bentonite.

Silicon Carbide and Graphite

The molds used in this series of experiments were poured vertically. Two bar diameters, 1¼ and 1 in., were chosen and the bar length was 8 in. Two grades of silicon carbide, medium and coarse, were used as mold materials. One series of molds was vitrified at 1300 C and another baked at 200 C for 8 hr. Both coarse and fine grades of graphite were studied. The graphite was bonded with either sodium silicate or an oil bond. Details of both the silicon carbide and graphite molds are given in Table 2. The iron 'J' in these experiments was of nominal C.E. 4.2 and the iron, superheated to 1450 C, was poured vertically into the molds at 1350 C. The tensile results obtained from the bars was given in Table 2.

Moisture

The Redhill 'H' and Redhill 1 4/28 bentonite-bonded and the Bromsgrove clay-bonded sands were studied in both green and dry conditions. The molds were dried by baking at 200 C for 8 hr. It was found that the moisture did not exert an appreciable effect upon the tensile strength of the iron of C.E. 4.2 cast into these molds.

Coal Dust

The Redhill 'H' and Redhill 1 4/28 bentonite-bonded and Bromsgrove clay-bonded sands were studied in the green condition both with and without the addition of 8 per cent coal dust. It was found that the coal dust showed no appreciable effect upon the tensile strength of the iron of C.E. 4.2 cast into these molds.

Inoculation

The effect of inoculation was studied only in Bromsgrove black sand and in ½-in., ¾-in., 1-in., and 1¼-in. diameter bars. The inoculants were introduced into the ladle prior to tapping, the degree of superheat being 1450 C and pouring temperature 1375 C. Figure 8 shows the effect of calcium-silicate in a C.E. 3.71 iron.

Irons of Various Carbon Equivalent

The results of the tensile strength tests on bars cast in irons 'J', 'K', and 'L' are plotted against bar diameter and given in Fig. 9. All three irons were cast into molds made of Redhill 'H', Redhill 1 4/28 and Bromsgrove black, respectively.

Square Bars

The iron used was one with a C.E. of 4.2, and the results are given in Table 3, which gives actual results obtained in tons/in.^{2} on the four bars from the three sands to enable other forms of graph to be plotted as required. In Fig. 10 tensile strength is plotted against volume-to-surface area ratio for both round and square bars.

Effect of Superheating the Iron

The effect upon the mechanical properties of the iron 'J' C.E. 4.2 was studied. Bars were cast in bentonite-bonded Redhill 'H' and Redhill 1 4/28 green sand molds. The superheat temperatures were

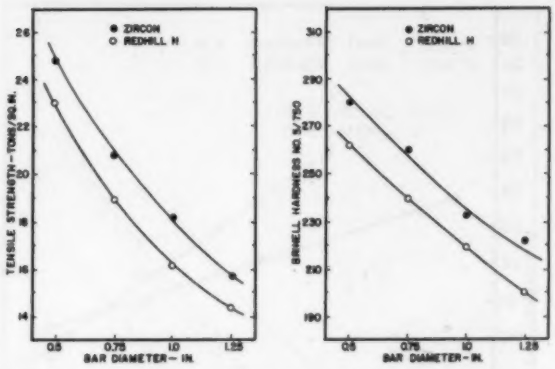
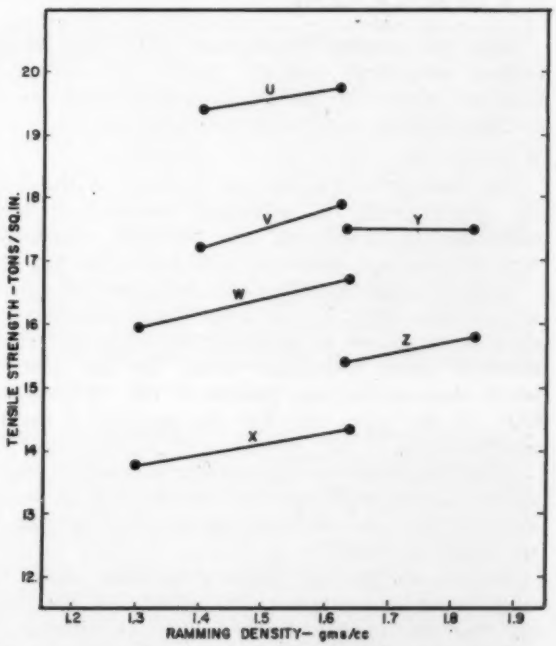


Fig. 6 — Tensile strengths and Brinell hardness numbers against bar diameters for iron J (C.E. = 4.2).

1550 C and 1350 C and the pouring temperature was 1340 C in both cases. The effect on the tensile strength is shown in Fig. 11. An increase in superheat from 1350 C to 1550 C with a constant pouring temperature of 1340 C, increases the strength of the 1-in. dia. bar by approximately 10 per cent in both sands.

Effect of Various Mold Materials and Section Size on Microstructures of Iron 'J' (C.E. 4.2)

The effect of section size and mold material upon the structure of the test bars of the iron of carbon equivalent of 4.2 only are considered as similar changes were observed in the other irons.



KEY:		Iron	Mold Material
		F	Bromsgrove
U	½ in.		
V	1 in.		
W	½ in.	G	Congleton
X	1 in.		
Y	½ in.	H	Bromsgrove
Z	1 in.		

Fig. 7 — Ramming density effect on tensile strength of test bars from step bar castings.

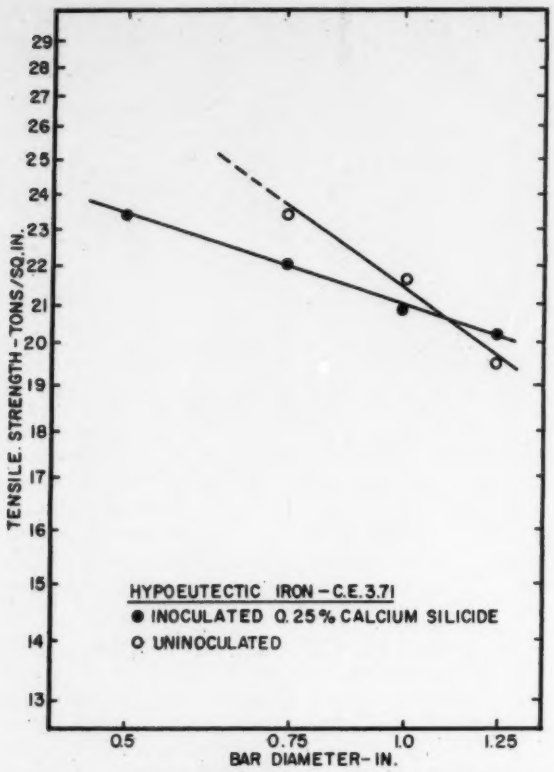


Fig. 8 — Inoculation by calcium-silicide effect on irons L and M (C.E. = 3.71).

Since the pouring temperature and degree of superheat were kept constant, any changes in structure are related to changes in cooling rate due to changes in either section size or mold material.

1) Section size.

The microstructures at the centers of the bars cast into Redhill 'H' green-sand molds have been considered in detail, and the unetched microstructures of these specimens are shown in Figs. 12a-12d.

Both the graphite type and flake size vary with section size, the 1¼-, 1-, and ¾-in. dia. bars all show types A and B graphite, the main difference between these specimens being in the rosettes, which show a dendritic pattern in the ¾- and 1-in. bars. In the ½-in. dia. bar the graphite is mainly type D.

At each section size the matrix structure is essentially pearlitic, although free ferrite tends to be associated with the undercooled graphite present in the ½-in. dia. bar.

Section size has an appreciable effect upon the distribution of the eutectic phosphide. In the ½-in. dia. bar this eutectic forms a network around the eutectic cells. With a slower cooling rate (¾-in. dia. bar) the eutectic phosphide tends towards a more random distribution, and in the 1- and 1¼-in. dia. bars it has a completely random distribution. This change from the random to a network distribution is shown in Fig. 12e-12f, and where the distributions of eutectic phosphorus in 1¼- and ½-in. dia. bars are shown.

The variation of eutectic cell number with bar diameter is shown in Fig. 12g, where the number of eutectic cells per cm at the center of a bar is plotted against bar diameter. The eutectic cell size decreases as the section size decreases, i.e., as the cooling rate increases.

2) Mold material.

The effect of mold material upon the tensile strength of a gray iron may be related to changes in structure. Considering the iron of carbon equivalent 4.2 the microstructures at the centers of the 1- and ½-in. dia. bars cast into the molds giving the maximum and minimum tensile results, Leighton Buzzard cement and Redhill 'H' shell, are shown in Figs. 12h-12k. These show that mold material can exert an appreciable effect upon both the types and flake size of the graphite, and in the case of the ½-in. dia. bars whether the iron is gray or mottled.

Work on Stepped Bar Castings

The details and results of the tests to determine the effect of mold material and ramming density upon the tensile strength of a rectangular stepped casting are shown in the following tables and figures:

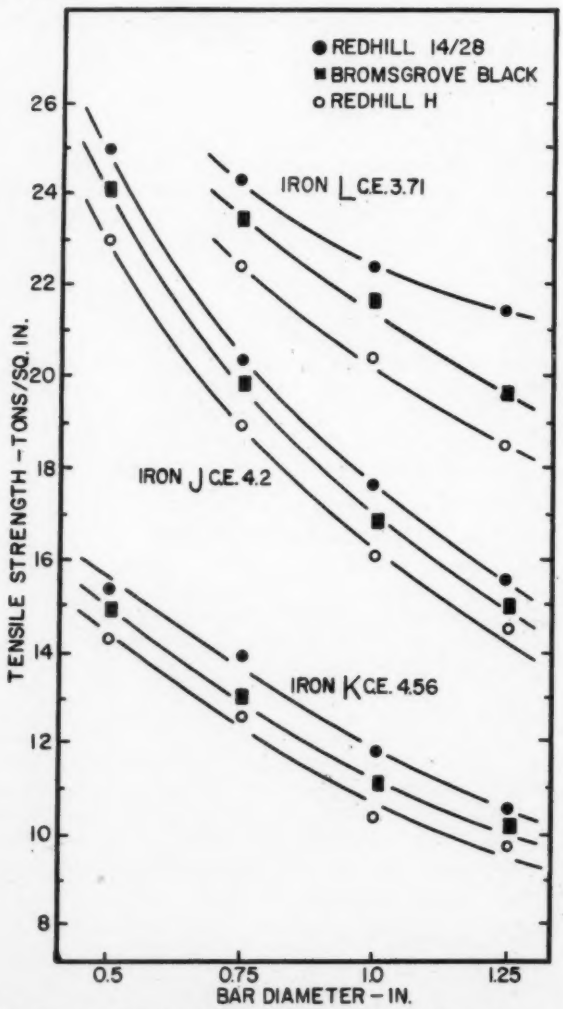


Fig. 9 — Relationship between mold material, carbon equivalent value, tensile strength and bar diameter.

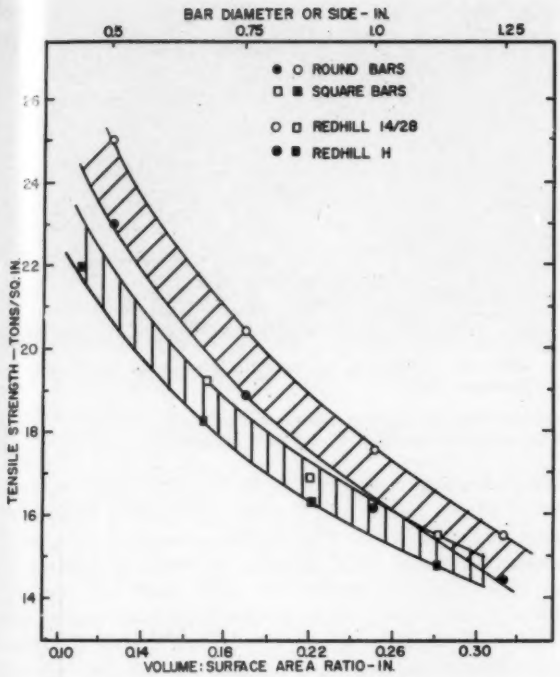


Fig. 10 — Tensile strength of round and square bars plotted against volume/surface area ratio.

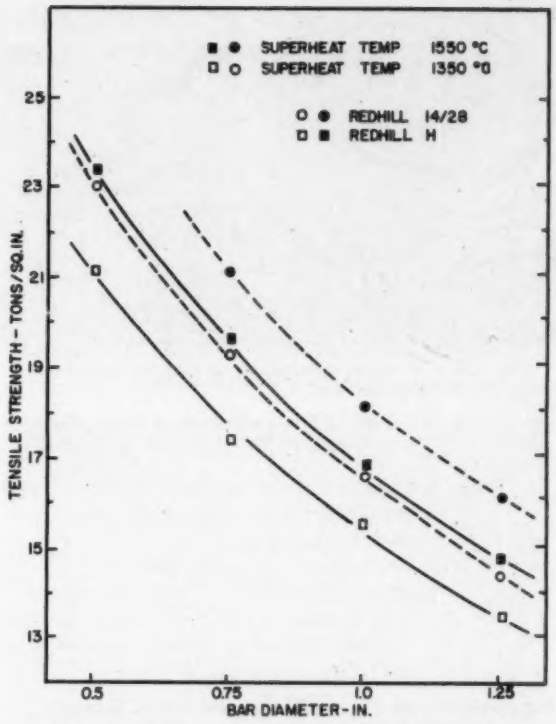


Fig. 11 — Degree of superheat effect on the tensile strength of iron J (C.E. = 4.2) cast into Redhill 14/28 and Redhill H green sand molds.

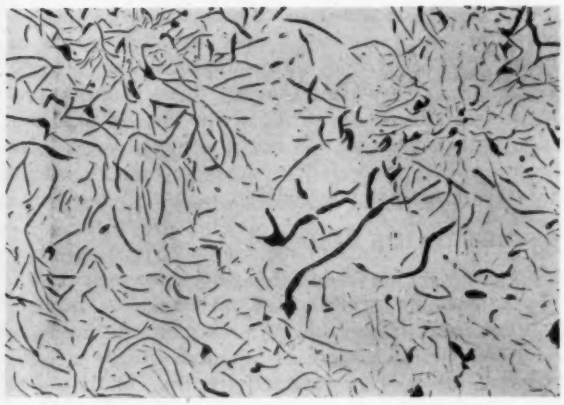


Fig. 12a — 1¼-in. diameter bar.

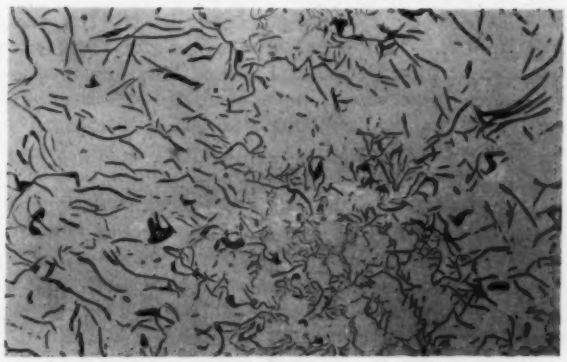


Fig. 12b — 1 in. diameter bar.

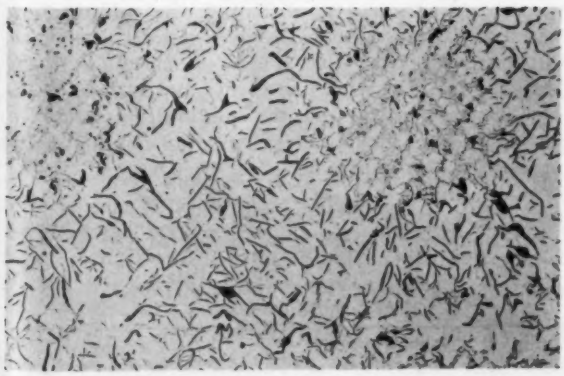


Fig. 12c — ¾-in. diameter bar.

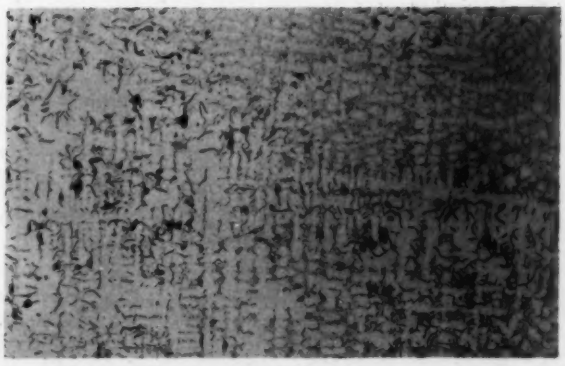


Fig. 12d — ½-in. diameter bar.

Fig. 12 a-d — Microstructures at the centers of 1¼, 1, ¾ and ½-in. diameter bars of iron J (C.E. = 4.2), cast in Redhill H bentonite bond green-sand molds. Unetched. 60 ×.



Fig. 12e — Random distribution of eutectic phosphide in 1¼-in. diameter bar.

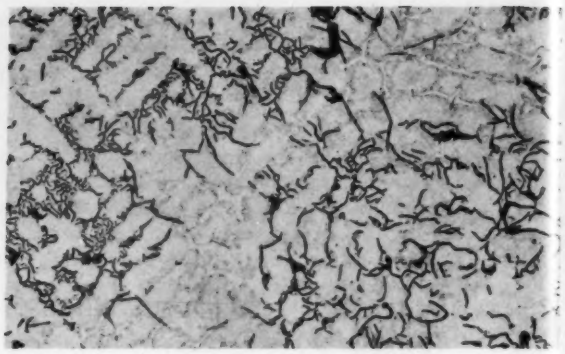


Fig. 12f — Network of eutectic phosphide in ½-in. diameter bar.

Fig. 12 e-f — Influence of cooling rate on the distribution of eutectic phosphide in bars of iron J (C.E. = 4.2) cast in Redhill H green-sand molds. 4 per cent Picral etch. 150 ×.

Table 4. Details of molds, mold material and tensile strengths obtained from 1-in. and ½-in. sections.

Table 5. Analysis of the irons used.

Figs. 13, 14 and 15 show the mean of the tensile values obtained plotted against ramming density for the 1-in. and ½-in. sections.

Experiments on Piston Rings and Valve Guides

This work was carried out to determine the effects of mold material at constant ramming density upon the physical properties and microstructure of cast iron piston rings and valve guides.

It was decided that as far as possible all test castings should be made by standard production methods, except that special care should be taken in presenting a given weight of sand to each box to ensure mold uniformity. Each box was subjected to a standard ramming or squeeze in an effort to obtain uniformity of mold density.

The analysis of the iron used in the experiments and degree of superheat and casting temperature are given in Table 6. Molds nos. 1 and 2 were made on a jolt-squeeze-draw machine, and nos. 3, 4, and 5 were made on an electromagnetic-squeeze-draw machine. Melting for molds nos. 1, 2 and 3 was carried out in a 600-lb indirect rocking arc furnace. Molds nos. 4 and 5 were cast in cupola-melted iron superheated in a mains-frequency furnace. Mechanical testing was carried out by machining a 0.178-in. dia. tensometer test piece from the bottom face of the valve guides. Piston rings were tested for En* and tensile strength by British Standard methods. Hardness tests were carried out on the bottom face of the valve guides and at 90° intervals round the diameter. Molding sands consisted of four different sands, details of which are given in Table 7.

The results obtained from the first series of tests with uniform ramming density are also given in Table 7. The second series in which an attempt was made to vary the ramming density under production conditions is set out at the bottom of Table 7. As will be seen, it was not possible to get a wide variation in ramming density, as soft molds would not stay in the shallow molding box.

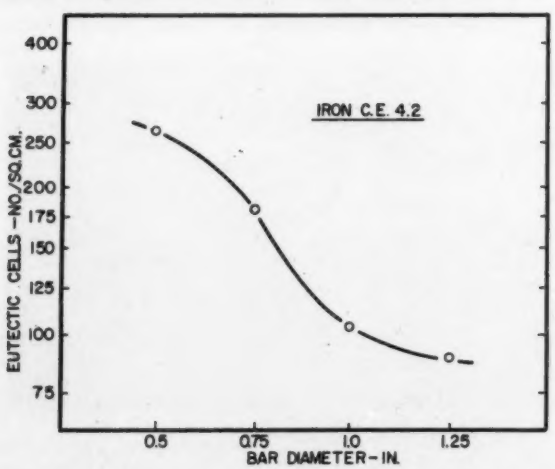


Fig. 12g — Variation of eutectic cell size with bar diameter. Iron J, Redhill H green-sand molds.

DISCUSSION OF RESULTS

Effect of Various Mold Materials on Mechanical Properties of Cast Bars

When melting variables are constant the mechanical properties of a given cast iron should be affected only by the cooling rate. Under the conditions of the experiments, the cooling rate is controlled by 1) thermal properties of the mold material, and 2) the size of bars. The difference in solidification times of castings due to these two factors may be calculated. Consider first the mold materials. It has been shown[2] that when the corner and surface curvature effects are neglected, the solidification time of a casting in a particular metal or alloy should be approximately proportional to $1/b^2$ where 'b' is the heat diffusivity.

*Modulus of elasticity determined on ring test piece, B.S. 5004.

TABLE 3 — TENSILE STRENGTH OF SQUARE BARS IN TONS/SQ. IN.

Length of side, in.	Volume-surface area ratio, in.	Redhill 14/28	Congleton	Redhill H
1.11	0.28	15.5	15.3	14.8
0.89	0.22	16.8	16.4	16.2
0.67	0.17	19.2	18.9	18.2
0.44	0.11	mottled	22.4	21.9

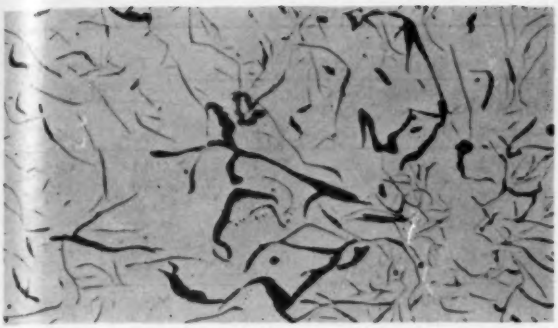


Fig. 12h — Redhill H shell mold. Unetched.

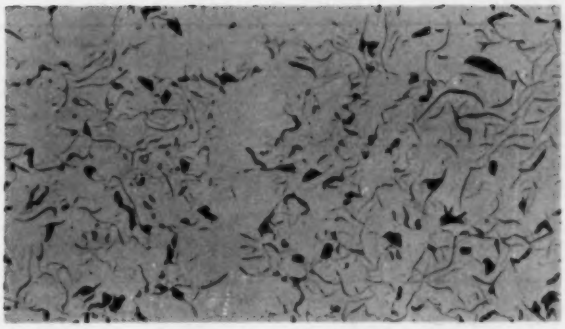


Fig. 12i — Leighton Buzzard C, cement bond. Unetched.

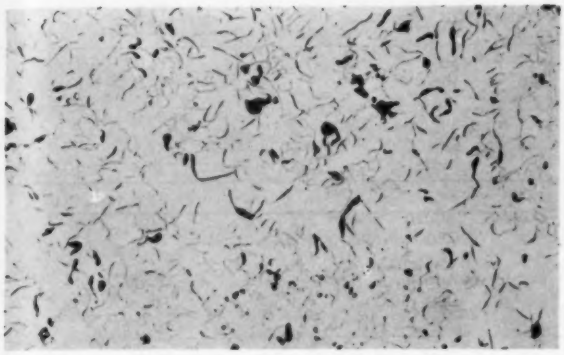


Fig. 12j — Redhill H shell mold. Unetched.

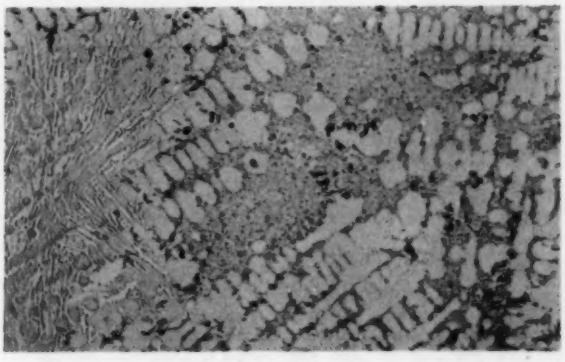


Fig. 12k — Leighton Buzzard C, cement bond. 4 per cent Picral etch.

Fig. 12 h-k — Mold material effect on structure of 1 in. diameter bar cast in iron J (h,i), and of ½-in. diameter bar cast in iron J (j,k). 100 ×.

Reference to the earlier work of the sub-committee[1] and the values of b[2] recorded there for two extreme sands, Redhill $1\frac{4}{28}$ and Redhill 'H', reveals that the solidification time of the fine sand rammed soft is approximately 1.8 times that of coarse sand rammed hard. In the case of bars of different size, Chvorinov's[10] rule may be used to calculate approximately the relative solidification times. Considering the ½-in. bar and the 1¼-in. bar, the solidification time of the 1¼-in. dia. bar is approximately 5½ times that of the ½-in. dia. bar.

From the results obtained in the experiments described, it will be seen that the effect of mold material on the physical properties of the cast iron is in line with the cooling rate of the iron which is mainly increased by an increase of 1) the grain size of the sand, and 2) the ramming density of the mold.

It has been shown[1,3] that the heat removed in a given time by a dry sand mold is proportional to the difference between the temperature of the metal/mold interface and the initial mold temperature, and the heat diffusivity 'b,' which may be expressed as

$$b = \sqrt{(kpc)}$$

where

k = apparent thermal conductivity of mold
p = density of mold
c = specific heat of mold

The term "apparent thermal conductivity" is used, as heat transfer in a sand mold may take place in several ways,[4,5,6] part of it being transferred by a true solid conduction mechanism, and part by radiation of heat between and through the sand grains.

In the present work the metal mold interface and initial mold temperature are nominally constant, and so to explain the different cooling powers of the different mold materials, differences in heat diffusivities of the materials must be considered. In this work there are five variables of which to take account, namely:

a) Physical and chemical properties of the mold material.
b) Grain size of the material.
c) Ramming density of the mold.
d) Bonding material.
e) Moisture content of the mold.

Physical and Chemical Properties of Mold Material

Both specific heat and density are terms in the equation expressing heat diffusivity, and so it is to be expected that the cooling power of a mold will depend upon these two properties of the mold material.

The thermal conductivity of the mold material would exert an influence upon the apparent thermal conductivity of a mold compact, as would the opacity of the material to heat radiation.

TABLE 4 – DETAILS OF MOLDS AND MOLD MATERIALS USED IN MAKING STEP CASTINGS, AND TENSILE STRENGTH OF TEST BARS OBTAINED FROM STEP CASTINGS

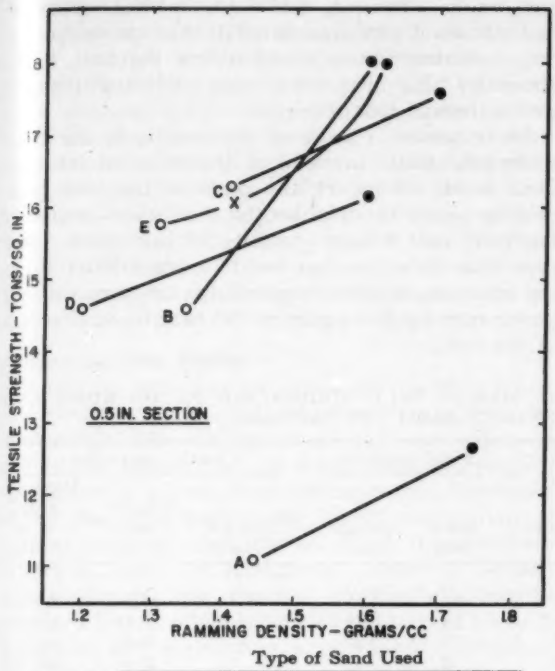
Iron and Casting Temp., C	Mold material	Sand grading, % on 60 mesh	Sand grading, % thru 60 mesh	Sand Properties Green Comp. lb/in.²	Sand Properties Green Perm.	Sand Properties Shatter No.	Mold details Moisture, %	Mold details Bond	Mold details Coal Dust	Mold details Density g/cc	Tensile strength tons/sq in. 1 in. Sect.	Tensile strength tons/sq in. ½-in. Sect.
A										1.75	10.1	12.6
1350	Bromsgrove	31	59	7.5	24	52	5.5	Natural Clay	yes		9.6	12.7
											8.8	11.3
										1.45	8.8	11.0
B								Added			15.5	18.2
1428	Hensall	93	7	9.6	68	83	7.6	Clay	yes	1.61	15.3	17.8
	Erith Loam and Ryarsh	1.5	98.5	13	23	61	3.0	Natural Clay	no	1.35	13.2	14.2
											13.1	15.0
C								3%			15.0	17.4
1385	Queslett	77.8	22.2	2.2	500	—	3.0	Bentonite	8%	1.70	15.6	
												17.8
	Kingscliffe	7.8	92.2	3.0	230	—	3.5	3% Bentonite	8%	1.41	14.4	16.6
											14.4	16.0
D								4%	8%		13.6	16.0
1390	Chelford	72.4	27.6	6.4	166	90	3.6	Added Clay	fine	1.6	13.8	16.2
	Congleton and Kings Lynn 4F	30.8	69.2	6.0	124	86	4.2	4½% Added Clay	8% fine	1.2	12.6	14.7
											12.4	14.5
E								3½%	8%		14.6	17.6
1350	Redhill 14/28	97.6	2.4	2.7	200		3.0	Bentonite	coarse	1.62	14.8	18.4
	Congleton	32.1	67.9	8.5	86	45	3.0	3½% Bentonite	8% medium	1.52	14.0	16.8
											14.0	16.6
	Redhill H	0.0	100	8.0	30	38	4.0	3½% Bentonite	8% fine	1.31	13.8	16.0
											13.6	15.5
Tests using one sand at extreme of ramming density												
F								Natural			17.9	19.7
1350	Bromsgrove	31	59	7.5	24	52	5.5	Clay	yes	1.62	18.0	19.8
										1.40	17.1	19.4
											17.3	19.4
G								3½%	8%		14.1	16.5
1350	Congleton	32.1	67.9	8.5	86	45	3.0	Bentonite	medium	1.62	14.5	16.9
										1.30	13.8	16.0
											13.7	15.9
H								Natural				
1380	Bromsgrove	31	59	7.5	24	52	3.0	Clay	yes	1.83	15.8	17.5
										1.63	15.4	17.5

TABLE 5 – ANALYSIS OF IRONS (STEPPED TEST CASTINGS) (Mold details given in Table 4)

Iron	Composition, % T.C.	Si	Mn	S	P	C.E.
Iron A	3.74	2.15	0.77	0.065	0.34	4.57
Iron B	3.43	1.87	1.13	0.057	0.066	4.07
Iron C	3.35	2.00	0.91	0.095	0.27	4.10
Iron D	3.42	2.11	0.50	0.111	0.59	4.32
Iron E	3.37	2.50	1.42	0.008	0.24	4.28
Iron F	3.11	1.66	0.42	0.10	0.26	3.75
Iron G	3.32	2.47	1.40	0.015	0.21	4.21
Iron H	3.42	1.80	0.82	0.087	0.18	4.08

TABLE 6 – ANALYSIS OF THE IRON FOR VALVE GUIDES AND PISTON RINGS WITH SUPERHEAT AND CASTING TEMPERATURE

Casting	Composition, % T.C.	Si	Mn	S	P	Temperature, C Superheat	Casting
1. Valve guide 3½ in. x ⅝ in. dia.	3.55	2.29	0.93	0.035	0.16	1440	1340
2. Valve guide 2 in. x 13/16 in. dia.	3.55	2.18	0.80	0.05	0.20	1430	1345
3. Valve seat	3.74	2.35	0.70	0.06	0.18	1430	1345
4. Piston ring	3.70	2.41	0.60	0.06	0.58	1640 + slot. bath 1500	1420
5. Valve plate	3.70	2.41	0.60	0.06	0.58	1640 + slot. bath 1500	1420



Ramming Density	Soft	Medium	Hard
Symbol	O	X	•
Iron Used			
A	Bromsgrove		Bromsgrove
B	Erith and Ryarsh		Hensall
C	Kingscliffe		Queslett
D	Congleton and Kings Lynn		Chelford
E	Redhill H	Congleton	Redhill 14/28

Fig. 13 — Mean tensile strength values for ½-in. diameter bars machined from step castings plotted against ramming density (results given in Table 4).

Grain Size of Material

Sand grain size affects the apparent thermal conductivity of a sand, an increase in the grain size increasing the apparent thermal conductivity. the heat diffusivity and so the cooling power of the sand. This increase in the thermal conductivity is attributable to three mechanisms. True thermal conduction across the junction between the sand grains, intergranular radiation, and transgranular radiation. Atterton[4] shows the importance of the last two at higher temperatures, which is in line with theoretical arguments.

Ramming Density of Mold

Mold density is one of the factors in the expression for the heat diffusivity, and so it is to be expected that an increase in the ramming density of a mold material produces an increase in cooling power of the material.

The true thermal conductivity of a sand aggregate will also increase as the compact density increases,[4] due to an increase in the area of contact between the sand grains.

Bonding Material

If ramming density is plotted against tensile strength for the bars of different diameters, cast in

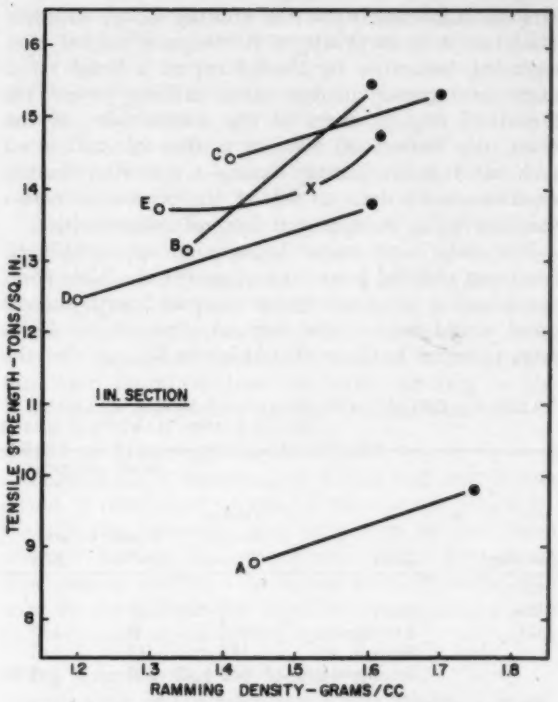


Fig. 14 — Mean tensile strength values for 1 in. diameter bars machined from step castings plotted against ramming density (see Fig. 13 for key, and Table 4 for results).

the iron of nominal C.E. 4.2, the graph shown in Fig. 7 is obtained for the 1¼-in. dia. bars. If chilling power of a mold is related to the tensile strength of a bar of a given diameter cast in that mold, it can be seen that 1) a coarse-grained silica sand has a greater chilling power than a fine-grained silica sand, 2) an increase in the ramming density increases the chilling power of both fine- and coarse-grained silica sands, and 3) the bond, with the exception of sodium silicate, has little effect upon the chilling power of silica sand aggregates.

The small effect of the bond, with the exception

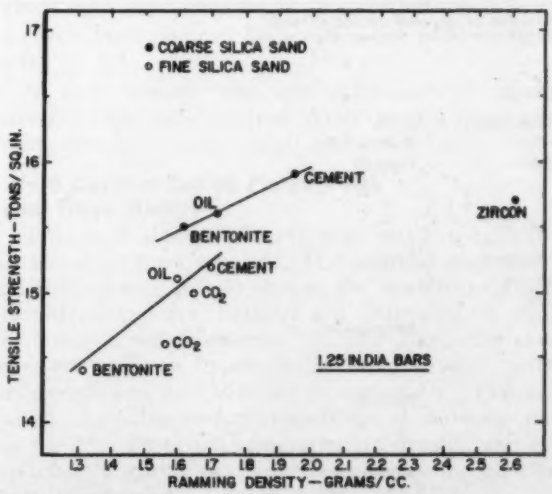


Fig. 15 — Mold variables effect on the tensile strength of 1¼-in. diameter bars of iron J (C.E. = 4.2).

of sodium silicate, upon the chilling power of silica sand molds is surprising. Pell-Walpole[8] found that replacing bentonite by linseed oil as a bond for a silica sand greatly increased the chilling power. He explained this in terms of the conductivity of the bond, the linseed oil forming a film of carbon on each sand grain during baking— this film having superior conductivity to that of the bentonite bond— thus increasing the apparent thermal conductivities.

The only bond found to have an appreciable effect upon chilling power is sodium silicate. This bond consists of a silica gel which compared with the oil bond would have a low thermal conductivity. However, it seems unlikely that this wholly explains the low results obtained, and it is perhaps possible that this bond produces materials that are opaque to heat radiation which would reduce the heat transferred by this mechanism, thus reducing the apparent thermal conductivity.

An important feature of the results is the relatively low tensile strengths of the bars cast into the shell molds. Morey et al.[9] consider that the poor chilling power of an unbacked shell mold is due to the poor heat storage capacity of the system, and show that it can be improved by using suitable backing materials. Ruddle[2] suggests that the poor chilling power may be due partly to the heat of combustion of the resin.

TABLE 7 — DETAILS OF MOLD AND MOLD MATERIAL USED IN MAKING VALVE GUIDES AND PISTON RINGS AND TENSILE AND HARDNESS RESULTS OBTAINED FROM THE CASTINGS

		Sand properties				Mold properties		Casting properties		
Casting	Sand	Green Comp., lb/in.	Green perm.	Density, g/cc	Moisture, %	Hard-ness	Density, g/cc	En. x 10^6 lb/sq. in.	Tensile. Tons/ sq. in.	Hardness, Brinell 5mm, 750 kg
No. 1 Valve guide	Erith	8.5	30	1.46	3.3	65	1.15		13.0	215/220
	Bromsgrove	12.5	36	1.47	3.4	75	1.49		13.63	213/219
	Hensall	13.5	105	1.47	3.8	71	1.40		10.83	210/213
	Southport	12.0	150	1.45	3.2	70	1.15		12.8	216/224
No. 2 Valve guide	Erith	8.5	30	1.46	3.3	70	1.10		13.2	197/229
	Bromsgrove	12.5	36	1.47	3.4	70	1.40		12.7	217/229
	Hensall	13.5	105	1.47	3.8	70	1.36		13.1	197/229
	Southport	12.0	150	1.45	3.2	65	1.00		13.1	207/229
No. 3 Valve seat	Erith	7.8	30	1.46	3.6	85	1.20			99/102
	Bromsgrove	12.6	40	1.46	3.5	85	1.40			100/103
	Hensall	13.5	105	1.47	3.8	84	1.27			99/103
	Southport	12.0	150	1.45	3.2	80	1.10			100/104
No. 4 Piston ring	Erith	8.5	30	1.46	3.3	78	1.00	15.6	22.4	102/104
	Bromsgrove	12.8	36	1.47	3.4	78	0.89	15.5	22.3	102/104
	Hensall	12.0	90	1.45	4.0	76	1.10	15.5	23.2	102/104
	Southport	10.0	140	1.48	3.2	74	1.02	15.4	22.3	101/103
No. 5 Valve plate	Erith	7.8	30	1.46	3.6	78	1.10		Rockwell 'B'	102/104
	Bromsgrove	12.6	40	1.47	3.5	78	0.90		Rockwell 'B'	102/104
	Hensall	12.0	90	1.45	4.0	76	1.20		Rockwell 'B'	101/103
	Southport	10.0	140	1.48	3.2	74	1.05		Rockwell 'B'	102/104

Results of second series of tests.

			Max. stress Hounsfield	
No. 1				
Repeated valve guide	Erith	1.19	10.2	170/182
	Bromsgrove	1.42	11.8	187/197
	Hensall	1.30	11.0	187/191

			En. x 10^6	Tensile	Rockwell 'B'
No. 4					
Repeated piston ring	Erith	1.27	14.3	21.4	98/99
		1.29	14.7	22.9	98/100
		1.31	13.9	22.1	99/100
	Bromsgrove	1.34	14.5	21.7	98/100
		1.35	14.5	21.6	99/101
		1.35	14.5	21.7	100/102
	Hensall	1.19	14.2	22.2	99/100
		1.25	14.6	23.2	98/99
		1.27	14.6	22.0	98/99
	Southport	1.28	14.4	20.8	99/101
		1.31	14.9	21.5	99/101
		1.32	14.0	20.1	100/101

Moisture Content of Molds

The moisture content of a mold, in the case of small castings and low casting temperatures, increases the cooling power of the mold.[3,6,7] Directly after pouring, the moisture adjacent to the metal/mold interface vaporizes, resulting in a large increase in the rate of heat absorption, and the subsequent movement of this steam away from the metal/mold interface results in a substantial transfer of heat outward. Ruddle and Mincher[3] consider that the effect of moisture upon the cooling capacity of mold materials decreases with increasing temperature, and has only a small effect even at copper casting temperatures.

Silicon-carbide Molds

The 1-in. dia. bars were mottled in both coarse and medium grades, and the effect of grade in silicon-carbide does not appear to have any effect on the tensile strength of the iron. However, vitrified molds gave a higher tensile strength than the baked molds, due apparently to the higher conductivity of the closer-bound vitrified mold. When silicon-carbide is used as a sand, the mechanical strength of the bars is not greater than the best mechanical strengths obtained in any other mold material used. This is in agreement with expectation, since it is not the thermal conductivity of the sand grains which exerts most influence over the cooling rate. On the other hand, when a vitrified bond is used then a much higher cooling rate is obtained and mottled bars result.

Effect of Various Mold Materials on Different Types of Cast Iron and of Inoculation

The influence of mold material and bar diameter upon the tensile strength of gray iron of various carbon equivalents is shown in Fig. 9. This indicates that tensile strength increases with decreasing carbon equivalent. All the ½-in. dia. bars cast in iron 'D', nominal C.E. 3.7, were mottled.

The results from the inoculated series are shown in Fig. 8. The base iron had the same nominal composition as iron 'J' (Fig. 9), and it can be seen that the variation in tensile strength with bar diameter is very much reduced and that the ½-in. dia. bars, which without inoculation were mottled, solidified gray. Inoculation reduced the coefficient of section sensitivity, i.e., the slope in logarithmic co-ordinates, from 0.37 to 0.17.

The effect of inoculation is greatest when the carbon equivalent value of the iron is low, as in the case of the iron under consideration, that is when the composition is such that the iron would very nearly solidify mottled and the eutectic cell size is large. In this case inoculation causes the graphite, which would be under-cooled in the absence of inoculation, to be of the flake type and the cell size to be reduced. It is therefore to be expected that the effect of inoculation would be to reduce the effect of variations of cooling rate, and therefore the variation of tensile strength with bar diameter.

Tests Using Square Bars

The first tests with square bars had shown no correlation whatever, an anomaly which was traced to be due to feeding. The round bars had a slight thickness at the end which acted as a feeder, and this had been omitted from the first series of square bars. When the square bar tests were repeated with the slight thickness on the end the results were reproducible and show a correlation similar to the round bars. The three sands had the same effect on square bars, but there was a most unusual effect in that the square bars had lower properties than the round bars.

Furthermore, the tests had been repeated in some light alloys and copper-base alloys, and in each case the properties of square bars were lower. Consideration was given to the effect of feeding as the thickness at the end of the bar is not a feeder in the true sense. Various degrees of feeding were tried as a variable, but were found to have no effect so long as the thickness at the end of the bar was not omitted.

Effect of Superheating the Metal

The degree of superheat to which cast iron is subjected is recognized as one of the factors which influence the structure and properties of gray iron castings. Within the temperature range considered, superheating increases the tensile strength of the iron used in the experiments, and the superheating temperature must be controlled.

Work Carried Out by Members

This work was carried out under practical conditions in a number of foundries operating under widely differing conditions, and the results indicated graphically in Figs. 13, 14 and 15, and in tabular form in Table 4, show that the effect of sand grain size and ramming density is maintained in all foundries.

In the case of iron 'A' where only one type of sand was used (Bromsgrove) the difference in tensile strength appears to be of almost the same magnitude as obtained in other foundries using coarse and fine sands as well as varying ramming density. To investigate this further the second program of work was carried out, and from the results shown at the bottom of Table 4 it will be noticed that the average difference in tensile strength when using one sand at extremes of moldable ramming density is about half that which could be expected when using a coarse sand rammed hard and a fine sand rammed soft.

It now appears that the difference in tensile strength obtained with iron 'A' in the first series and using only one sand is exceptional.

Work Carried Out on Piston Rings and Valve Guides

This work shows little evidence which is relevant to the other work reported. The physical results and microstructures are all within the usually expected variations, and are therefore not illustrated. It must be borne in mind, however, that the piston-ring castings are made in hypereutectic iron. Another point to emerge was that with the exceptionally great care taken to achieve uniform conditions of molding, and in the first series uniform ramming density, the En and tensile scatter was reduced to about one quarter of its usual range.

It was considered that it was unlikely that any

real effect would be observed with the small differences in density, and as the sands were all fine the differences in both grain size and mold density were such that no great difference could be expected.

CONCLUSIONS

The following main conclusions may be drawn from these investigations:

1) The work at the University and the field tests carried out in production foundries using standard materials in everyday use demonstrated that the tensile strength of an iron is influenced by the mold material, the tensile strength being higher when the mold material is coarse-grained and rammed hard and lowest when the mold material is fine-grained and rammed soft.
2) The influence the mold exerts on the tensile strength is due to the mold material or mold density or both affecting the cooling rate of the molten metal.
3) The maximum change in tensile strength brought about by the widest range in sand mold material and mold density studied is approximately 20 per cent. This is from half to one third of that which could be expected by a change in section size (bar diameter) from 1¼-in. to ½-in. diameter.
4) The results of the tests using cast test bars are confirmed by the results from test bars machined from the step castings made under commercial conditions in four foundries using iron with widely different carbon equivalent values.

ACKNOWLEDGMENTS

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REFERENCES

1. Report of Sub-committees T. S. 21, 33 and 46 of the Technical Council, *Proc. Inst. Brit. Found.*, 1955, vol. 48, p. A218.
2. R. W. Ruddle, *Solidification of Castings*, London Inst. Metals, monograph No. 7, London, 1950.
3. R. W. Ruddle and A. L. Mincher, *J. Inst. Metals*, 1949-50, vol. 76, p. 43.
4. D. V. Atterton, *J.I.S.I.*, 1953, vol. 174, p. 201.
5. R. P. Dunphy and W. S. Pellini, AFS TRANSACTIONS, 1951, vol. 59, p. 425.
6. H. F. Bishop, F. A. Brant and W. S. Pellini, AFS TRANSACTIONS, 1951, vol. 59, p. 425.
7. V. Paschkis, AFS TRANSACTIONS, 1952, vol. 60, p. 163.
8. W. T. Pell-Walpole, Foundry Congress, Chareville, 1956.
9. R. E. Morey, H. F. Bishop and W. S. Pellini, AFS TRANSACTIONS, 1954, vol. 62, p. 448.
10. N. Chvorinov, *Proc. Inst. Brit. Found.*, 1938-9, vol. 32, p. 229, and *Giesserei*, 1940, vol. 27, p. 177.

INTERRELATION BETWEEN STRESS CONCENTRATION AND CASTABILITY

By J. B. Caine

ABSTRACT

There is a close relationship between shapes designed to minimum stress concentration, maximum load carrying ability and the castability of such shapes. The designer and foundryman are in basic agreement on this subject to the point that most, if not all, shapes designed for minimum stress concentration are shaped for maximum castability.

INTRODUCTION

For years the foundryman, individually and collectively as an industry, has been trying to interest the designer and engineer in designing for castability. The results of this campaign have not nearly equalled the efforts expended upon it. Perhaps one reason is that the approach is psychologically wrong. The designer and engineer are not particularly interested in the problems of the foundryman. These individuals are interested in load-carrying ability, and should be interested in the mutual problem of load-carrying ability and castability.

The examples discussed in this paper are only a few that illustrate this mutual problem. The designer's interest in stress concentration can almost always be translated into foundry terminology by substituting "strain" for "stress". The shape that plagues the foundryman with defects due to contraction strain concentration during solidification and subsequent cooling is prone to plague the designer due to stress concentration in service.

THE PROBLEM

In an effort to illustrate this mutual problem, the writer has attempted to summarize the known information on stress concentration of a few simple shapes, and correlate this information with heat transfer data on similar shapes. Only the simplest shapes are discussed in this paper. The simpler shapes must be discussed and understood first to properly understand the stress pattern of more complicated shapes. Then too, as the shape becomes more complicated to the point of most castings, generalization based theoretical and photoelastic studies of stress concentration, thermal analysis of strain concentration, becomes less precise. Each shape, or small groups of shapes must be studied as individuals. The problem becomes one of individual stress analysis by brittle laquer, or strain gage techniques.

Specifically, this discussion compares stress analysis data, determined mostly by photoelastic methods, with thermal analyses determined by thermo-couple and electric analogue techniques. As none of the work was done with the problem under discussion in mind, there must be some extrapolation and interpretation of the original data. This extrapolation introduces errors, but at the present state of knowledge the errors should not be serious. More important is the question of limits to the conclusions.

The photoelastic data given seem reliable in that the stress concentration factors (K_t), derived from photoelastic work, check closely experimentally determined decreases in fatigue strengths of metals (K_f). The stress concentration factors (K_t) apply to all metals. A major point here is the particular metal's sensitivity to stress concentration. This is not only a function of each metal, but also of the hardness and microstructure of a particular metal.

The limits to the correlative strain concentration data must be more strongly exphasized. Practically all quoted data are specific only to steel. Most are specific only to steel cast in 4-in. sections. For all practical purposes for the present, it does not seem necessary to restrict the correlation only to steel poured in heavy sections. Thermal gradients in shapes cast of metals with temperatures and solidification characteristics other than those of steel will differ from those shown, but the relative gradients should be similar.

A similar situation exists for section sizes other than the 4-in. section that will be referred to repeatedly. Thermal gradients will vary due to section size, but again the relative gradients should be similar. Therefore, all dimensions will be given in relation to section thickness (t).

A major exception to this discussion is gray cast iron. This most castable of all metals is not prone to the defects associated with solidification shrinkage common to most other cast metals. Its different strengths in tension and compression make other shapes more attractive to the designer than those best suited for the so-called ductile metals. The shapes discussed in this report must be amended for gray iron design.

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The first shape to be considered is the simple corner. As can almost be expected, this most basic shape has been taken for granted and overlooked. There are heat transfer data on this shape, but to the writer's knowledge, no basic stress analysis data. Therefore, a discussion of this basic problem must be deferred, to a later portion of this paper.

BASIC SHAPES

The "L" Junction

The next simplest shape is the "L" junction of Fig. 1. Stress concentration studies of this shape, as influenced by the radii of the fillets, have been made by Roark, Hartenberg and Williams.[1] The K_t values reported in reference (1) have been plotted in Fig. 1. Before delving more deeply into this subject, perhaps an explanation of the significance of the K_t (stress concentration) factors is in order.

The value K_t denotes the increase in overall stress due to geometry. The geometry is established basically by design, but may be modified or accentuated by other factors, macro or micro discontinuities caused by fabrication or inherent in the metal, type of load application, accidental overloads and misalignment. For example, a K_t factor of 2.0, in Fig. 1 and those to follow, indicates that for every 1000 lb of overall load, the metal at the surfaces of the junctions of sections (usually the fillets) is subjected to twice the overall stress, or to a stress of 2000 lb. A stress concentration factor of unity (1.0) denotes no stress concentration, and the load is distributed evenly throughout the shape.

These values, 1.0, 1.3, 1.7, 3.0 seem rather innocuous, but they are not. A K_t factor of 3.0 means that the load in pounds, or in psi in certain areas is tripled; K_t of 1.5 means a 50 per cent increase in load. The importance of these factors is accented when they are compared with possible changes in the metal's ability to withstand the load, tensile, yield and fatigue strength.

Consider Fig. 1 as the first specific example. It shows — as the radius of the fillet is increased from $r/t = 0.2$ to $r/t = 1.0$, K_t in tension decreases from 2.03 to 1.20. This means that the load carrying ability under these conditions of shape and loading is increased 70 per cent regardless of the type of metal. It should be emphasized that a r/t ratio of 0.2 is about equivalent to the normal 1/4-in. radius with a 1-in. section.

Castability follows, as shown by the scale drawings of Fig. 1. Brandt, Bishop and Pellini[2] have studied the thermal gradients of "L" junctions and found a freezing pattern (simplified), as shown in Fig. 1. The shaded areas of Fig. 1 represent metal containing liquid when the connecting members are completely solid. These sketches have been constructed from the 50-sec isochromes of Brandt, Bishop and Pellini's work.[2]

It must be re-emphasized that the isochromes are specific only for steel cast in 4-in. sections. The data should be qualitatively relative for other metals, cast in other sections. Another qualification must be made. Brandt, Bishop and Pellini did not investigate the specific problem of size of fillet on the freezing pattern. It is assumed constant in regard to fillet radii. This assumption should not introduce a great error, and if anything should make Fig. 2 conservative, for the larger radii should cause thicker casting skins at the fillet.

With radii under $r/t = 0.5$ in Fig. 1 there is liquid metal at the sand-metal interface of the fillet when the connecting members are completely solid. The probability of shrinkage defects at the surface of the highly stressed fillet is self evident. With radii over 0.5 r/t, and especially when the ratio approaches $r/t = 1.0$, there is appreciable solid metal at the fillets, and the thermal isochromes follow the geometric outlines of this junction of sections.

The dotted lines of the sketches of Fig. 1 denote the 10-sec isochrome for the 4-in. section. Metal between the dotted lines and the sharp outer corner is cooling and freezing over five times as fast as the metal in the hot spot. The chances for trouble due to steep temperature gardients is self evident. Cooling cracks in hardenable alloys and chilled edges in iron are just two possible defects that are caused by these steep temperature gradients. The analogous stress concentration picture does not show up with the tensile loading of Fig. 1, for the sharp outer corner is in compression.

Lyse and Johnson[3] have partially investigated stress concentration in the same "L" junction, but stressed in torsion. Their results are plotted in Fig. 2, in comparison with those of Fig. 1 in tension. There is a significant increase in stress concentration when this shape is stressed in torsion. A prime suspect is the sharp outer corner that is now stressed in torsion and not in compression. It can at least be inferred that if the outer corner be rounded as shown in Fig. 2b to a 2t radius, stress concentration in torsion would approach that in tension and perhaps both would be decreased. At the same time, rounding the outer

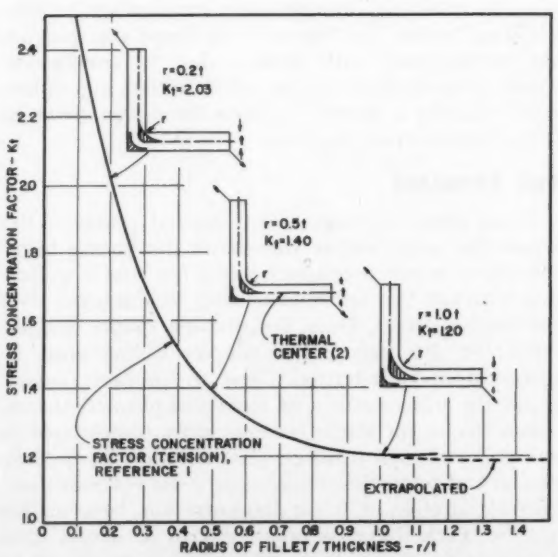


Fig. 1 — Relationship between stress concentration of an "L" section in tension and thermal gradients near the end of solidification.

corner removes the temperature gradient that causes cooling rates to vary within a fraction of an inch by a factor over five. Need it be said that such streamlining is easily accomplished by casting, is difficult or impossible by any other forming method?

Figures 1 and 2 illustrate the writer's thesis, the common problem of stress concentration and castability. A rather simple change from Fig. 2a to Fig. 2b has increased the load carrying ability of the common "L" junction, at least in fatigue, by 70 per cent, an appreciable increase. The same change has increased castability by moving the thermal center to the geometric center, and has eliminated extreme temperature gradients. It is no longer necessary to sell castability. If streamlining is done properly castability takes care of itself.

Another qualification must be made. A thermal analysis of the 1t-2t rounded corner has not been made. However, it can be safely assumed that the outer edge of the hot spot of Fig. 2b does not extend beyond the geometric centerline. The surface area-volume ratio of the outer surface of the 2t radius is greater than the straight sides of the legs. Therefore, the cooling rate at the geometric center of the corner must be greater than in the straight legs. There is no reason to believe that the heat flow across the 1t inner radius of Fig. 2b is less than across the same radius of Fig. 2a. Therefore, the size of the hot spot in Fig. 2b must be materially decreased. However, this reasoning must be checked before being accepted without question.

For example, the "L" junction of Fig. 2a is quite resistant to hot tearing even though it is prone to shrinkage defects in the fillet. The reason is that the sharp corner is cooling so fast that it holds the junction together, possibly to the point of subjecting the fillet to compression. Removing this temperature gradient can conceivably cause trouble. Perhaps a compromise between Figs. 2a and 2b is necessary. Only experimental work will settle this point.

The "T" Junction

Figure 3 compares stress concentration and castability as measured by heat transfer for the next more complicated junction, the "T" junction. Stress concentration values in relation to fillet radii are quite similar to those of the "L" junction. They also level off at fillet radii of about 1.0t, and there is little to be gained with larger radii as far as stress concentration is concerned.

Just a casual inspection of the thermal gradients as determined by Brandt, Bishop and Pellini[2] shows why such junctions are so prone to hot tearing. There is a path of liquid metal extending through the junction of the horizontal arm of Fig. 3a when the uniform arms are completely solid. Contraction strain is not only concentrated at this junction, but also the partially liquid metal can offer no resistance to this strain. Increasing the radius as in Figs. 3b and 3c gives a solid envelope around the hot spot as well as decreasing stress concentration in service. It is evident that a combination of a sharp corner as in Fig. 3a, plus a hot tear or shrink crack, can increase stress concentration to values approaching five.

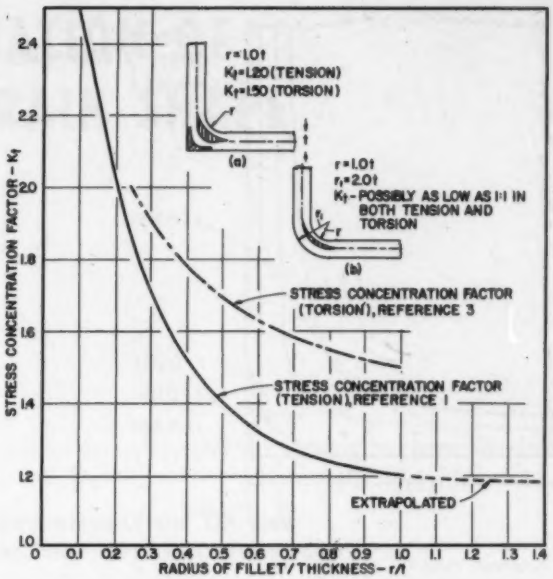


Fig. 2 — Relationship between stress concentration in tension and torsion and thermal gradients during solidification.

The larger radii of Fig. 3c will not guarantee freedom from hot tears. They can only be expected to reduce their severity. So much is dependent on so many other factors, particularly size, that further discussion in a paper of this type is of no value. If "T" junctions must be used in shapes that show shrinkage defects the foundry must carry on from the point of 1t radii with chills, brackets, collapsible molding and core sands.

An interesting point of conjecture is shown by Fig. 3d. Several publications on casting design recommend "dimpling" a "T" section to obtain uniform geometric sections. When this is done, as in Fig. 3d, it is possible that the "dimple" will protrude into the hot spot and make matters worse. There

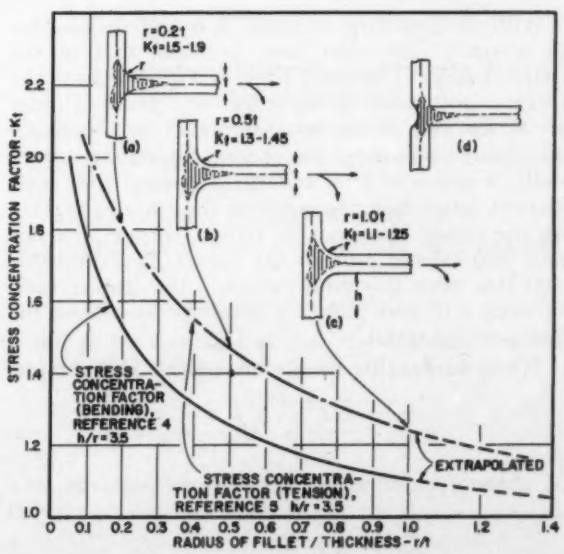


Fig. 3— Relationship between stress concentration and thermal gradients of "T" junctions.

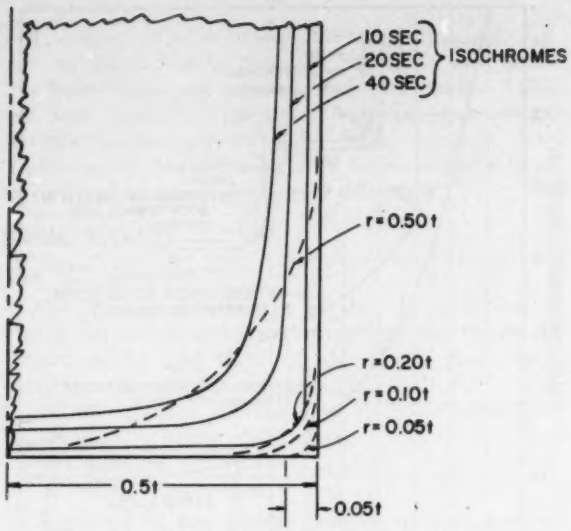


Fig. 4 — Relationship between thermal gradients and corner radii.[6]

may now be three rather than one semi-liquid cross-sections. Unfortunately, the literature on casting design is full of recommendations such as this, not based on factual evidence but on an idea. Most suggestions are good, but there are too many that possibly, or probably, are not.

It is possible for the reader to extrapolate the data for "T" sections to the more complicated "K" and "X" sections. No stress or heat transfer data are available for these more complicated junctions and they can not be discussed in detail. However, it does not seem to be too much in error to recommend the 1t radius for all cast junctions of sections as a minimum. "K" and "X" junctions should be simplified to staggered "T" junctions wherever possible, as has been recommended in practically all publications on casting design.

The Simple Corner

With the preceding in mind, it is perhaps possible to return to the most basic problem, that of the simple corner. The solid lines of Fig. 4 denote the thermal isochromes of the corner of a finite cylinder at 10, 20 and 40 sec, as determined by Paschkis.[6] The dotted arcs show the outlines of various corner radii. A radius of 0.10t removes the metal that cools at rates faster than the metal of the surfaces making up the corner. As the 0.10t radius increases the surface area-volume ratio of the corner, it is probable that this value is a minimum, and that corner radii between 0.10 and 0.20t are necessary to remove the highly cooled metal.

When hardenable metals are cooled, the elimination of these high temperature gradients will do more than any other single step to eliminate corner cracks. This is true for any hardenable metal, regardless of the forming method. In any other forming process than casting, rounded corners usually require an additional chamfering operation. Most rounded cast corners can be made on the pattern with a simple templet and a piece of sandpaper. The two exceptions, sharp corners formed at partings and where cores leave the casting cavity can be remedied by simple pattern construction if they are specified before the pattern is made.

Rounding external corners to 0.20t radius seems even more important when casting gray iron. Here, the problem is not only corner cracks, but also chilled edges. Chilled edges many times limit the strength of iron castings in that without expensive alloying, or heat treatment, a particular casting cannot be machined if cast with an iron with any higher strength. Would not the elimination of the corners simplify the problem?

No correlative data are available on service stress concentration of simple corners. However, the correlation between thermal and stress concentration data seems so close that it does not seem too hazardous to assume the same decrease in stress concentration as is obtained in thermal gradients.

SUMMARY

Three general rules can be advanced from the preceeding correlations between stress and strain concentration. All dimensions can be expressed as fractions of the section thickness, t.

1) Corners should be rounded to 0.1 to 0.2t radius.
2) Fillet radii of 1.0t seem generally applicable to most junctions of sections.
3) Outer radii of 2t when used in conjunction with 1t fillet radii seem the best compromise between stress concentration, castability and moment of inertia of "L" and "K" junctions.

REFERENCES

1. Roark, Hartenberg and Williams, "Influence of Form and Scale on Strength," Engineering Experiment Station Bulletin, University of Wisconsin, 1938.
2. Brandt, Bishop and Pellini, "Solidification at Corner and Core Positions," AFS Transactions, *vol. 61*, p. 451, 1953.
3. Lyse and Johnson, "Structural Beams in Torsion," Lehigh University Publications, *vol. 9*, 1935.
4. Leven and Hartman, "Factors of Stress Concentration for Flat Bars and Shafts with Centrally Enlarged Section," *Proc., Society of Experimental Stress Analysis, vol. 9* no. 1, 1951.
5. Frocht and Landsberg, "Factors of Stress Concentration in Bars with Deep, Sharp Grooves and Fillets in Tension," *Proc. S.E.S.A., vol. 9* no. 2, 1951.
6. V. Paschkis, *"Solidification of Finite Cylinders,"* AFS Transactions, *vol. 62*, p. 48, 1954.

EXPERIMENTAL DETERMINATION OF SPECIFIC SURFACE AND GRAIN SHAPE OF FOUNDRY SANDS

By Franz Hofmann

ABSTRACT

The methods for determining the grain shape of foundry sands described in the literature are based almost exclusively on microscopical observation. They are subject to the disadvantage of being rather inaccurate and to personal interpretation. With the new testing method described the average grain shape of sand grain distributions as well as of individual grain size fractions can be determined experimentally and reproducibly.

TESTING PRINCIPLE

With the permeability method described herein it is possible to determine the actual specific surface of sand and other granular material, i.e., the surface in sq cm of all sand grains contained in one gram of sand. On the other hand, the theoretical specific surface of the sand may be calculated from its sieve analysis assuming that all grains are spherical. The value of the theoretical specific surface is always smaller than that of the actual specific surface, due to the shape of the grains which are not spherical, and which for this reason have a larger surface area with identical weight.

The more angular the sand grains are, the larger the difference is between actual and measured specific surface area and theoretical specific surface area. Dividing the value of the actual specific surface area of sand by the value of the theoretical specific surface area yields a coefficient for the angularity of a grain. This is a unit for the average grain shape:

$$E = \frac{S_w}{S_{th}}$$

where E = Coefficient of angularity.
S_w = Actual measured specific surface in sq cm/g.
S_{th} = Theoretical specific surface assuming that all grains are of spherical shape, in sq cm/g.

Only in the ideal case of a sand consisting entirely of spherical grains S_w would be equal to S_{th} and E would be 1.

This principle has been applied by Robertson and Emödi[1] for the first time to characterize the rugosity of granular solids, and the method has been adapted to test foundry sands by Davies and Rees.[2,3]

DETERMINATION OF THE ACTUAL SPECIFIC SURFACE

The measurement of the actual specific surface of sand bases on the flow of air through a long cylindrical sand bed. The actual specific surface S_w is a function of porosity and permeability of this bed.

$$S_w = \frac{\text{Constant}}{\text{Bulk Density}} \sqrt{\frac{\text{Porosity}^3}{\text{Permeability}}}$$

Basing on a formula of Kozeny[4], Carman[5] has suggested the following, semi-empirical formula:

$$S_w = \frac{1}{\delta(1-\epsilon)} \sqrt{\frac{\epsilon^3 A p g}{K \eta Q L}}$$

where

δ = specific weight of material investigated (silica sand 2.65).
ϵ = porosity of the sand bed (pore space volume per volume unit of the sand bed).
A = cross-sectional area of the sand bed (sq cm).
p = pressure difference between the ends of the sand bed (dyn/sq cm).
g = gravitational acceleration (981 cm/sec^2).
K = constant (= 5, not affected by apparatus and grain size).
η = dynamic viscosity of air (g/cm/s).
Q = rate of flow of air through sand bed (cc/sec).
L = length of sand bed (cm).

g and K are constants so that the equation may be simplified as follows:

$$S_w = \frac{14}{\delta(1-\epsilon)} \sqrt{\frac{\epsilon^3 A p}{\eta Q L}} \qquad (1)$$

This equation contains the permeability formula of Darcy:

$$Q = PE \frac{A p}{\eta L} \quad \text{or}$$

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$$\text{Permeability} = PE = \frac{Q\,\eta\,L}{A\,p}$$

This permeability formula shows the same relation as the gas permeability formula for foundry sands, but it also considers the viscosity of air. The value of the dynamic viscosity of air (η) is considered to be constant at room temperature and is equal to $181 \cdot 10^6$.

Davies and Rees[2,3] have applied the measuring principle of Lea and Nurse[6] for foundry sands. The method was first developed for measuring the specific surface of fine powders, basing on Carman's equation. Air is passed by water pressure through the sand arranged as a cylindrical bed in a 50 cc glass burette, and then through a fine orifice (actually a capillary tube). Pressure is measured before and after the sand bed.

Rigden[7] simplified the principle of Lea and Nurse by using a U-tube, and drawing air through the powder sample by means of the liquid column in a U-tube, first lifted by suction to a certain height. Blaine[8] further developed the U-tube measuring principle. He simplified the testing device to its present state as it is used for evaluating the specific surface of fine powders, particularly of cements. It has never been applied for granular material.

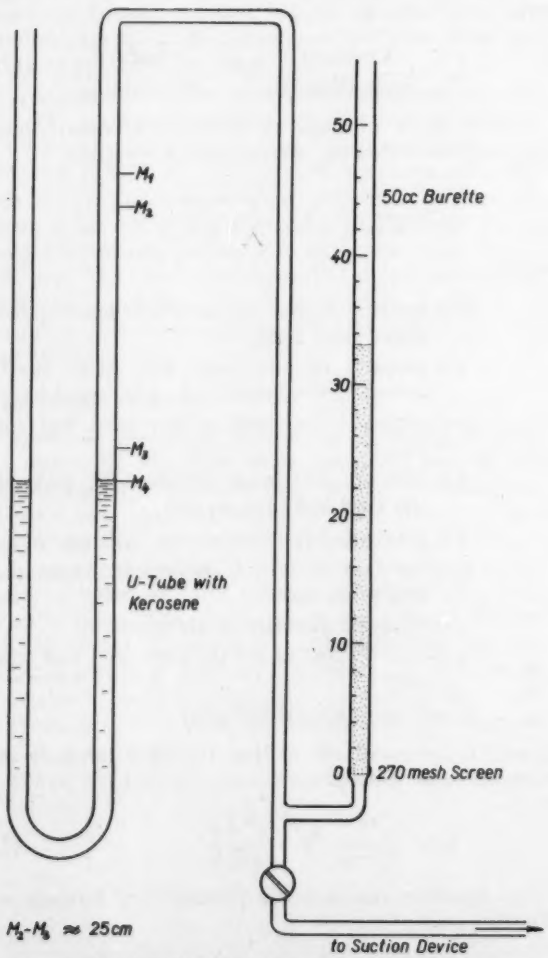


Figure — Schematical sketch of the equipment for the determination of the actual specific surface of sands.

The measuring method shown in the figure has been developed by the author and bases essentially on the principle of Blaine, adapted to permit the measurement of the specific surface of granular materials, mainly sands.

Air is drawn through the sand bed in a measuring pipette by means of the liquid column in the U-tube, which has been lifted by suction up to mark M_1. It is now allowed to drop, drawing air through the sand bed. The time used by the liquid column to drop from mark M_2 to mark M_3 is measured. Equation (1) may now be adapted as follows to this measuring method:

The value Q, i.e., the (average) amount of air drawn through the sand bed per time unit is

$$Q = \frac{V}{T}$$

where V = total volume of air (ccs) flowing through the specimen equalling the volume of the length of U-tube between marks M_2 and M_3.

T = time required by the liquid column for dropping from M_2 to M_3 (sec).

Using equation (1), the specific surface will then be

$$S_w = \frac{14}{\delta(1-\epsilon)} \sqrt{\frac{\epsilon^3\,A\,p\,T}{V\,\eta\,L}} \qquad (2)$$

According to Rigden[7] and Blaine[8] the pressure (p) may be expressed as follows for the U-tube method:

$$p = \frac{0.8686\ h_1\ \gamma}{\log\,(1+2\ h_1/h_2)}$$

where

h_1 = height between M_2 and M_3 = measuring interval.

h_2 = double height between M_3 and M_4.

M_4 = equilibrium height of the liquid column.

γ = specific weight of the liquid in the U-tube (kerosene or any suitable mineral oil of low viscosity).

With equation (2):

$$S_w = \frac{14}{\delta(1-\epsilon)} \sqrt{\frac{\epsilon^3\,A \cdot 0.8686\ h_1\ \gamma}{V\,L \log\,(1+2\ h_1/h_2)\ \eta}} \sqrt{T}$$

or with $\delta(1-\epsilon) = D$ = bulk density of the sand bed, the formula may be expressed as follows:

$$S_w = \underbrace{\frac{1}{D}\sqrt{\frac{\epsilon^3}{L}}}_{X} \cdot \underbrace{14\sqrt{\frac{A \cdot 0.8686\ h_1\gamma}{V \log\,(1+2\ h_1/h_2)\eta}}}_{Y} \cdot \underbrace{\sqrt{T}}_{Z}$$

or

$$S_w = X \cdot Y \cdot Z \qquad (3)$$

The value for X results from the volume and the length of the sand bed (of known weight) yielding also D and ϵ.

The value for Y is a constant of a particular apparatus, and will have to be determined only once. The measurement is therefore greatly simplified, and is reduced to the determination of the volume of the sand bed, its length for a particular apparatus, and the time measurement for the drop of the liquid column between M_2 and M_3, respectively. Basing on these figures and the specific weight of the material being tested, the actual specific surface is easily obtained.

As a preliminary to testing, the sand is boiled with an addition of sodium-pyrophosphate, and subsequently stirred up in the same manner as used for the determination of the clay content. It is then washed through a 270 mesh screen by a gentle stream of water, and then dried. The initial sample of the sand must be big enough to produce a sample of exactly 50 grams of washed and dried sand. The same sample is subsequently used for the determination of the theoretical specific surface by means of the sieve analysis.

The dry sand is filled as uniformly as possible into the 50 cc burette which stands upside-down, and is closed by a 270-mesh screen at the zero mark. The burette is slightly tapped by hand or with a pencil until the sand volume cannot be reduced any further. From the volume results the length of the sand bed.

Various sands have different bed volumes. The finer the sand, the larger the bed volume. At the same time, the grain size distribution has an effect; a narrow distribution giving higher volumes than a wide distribution. Together with the gaging of the apparatus, it is possible to draw curves for various bed volumes and flow times T. Using these curves, the determination of actual specific surface is simple and fast.

DETERMINATION OF THEORETICAL SPECIFIC SURFACES

Assuming that all grains are spherical, the theoretical specific surface S_{thi} for a particular grain size fraction (sieve interval), i is

$$S_{thi} = a_i \pi d_i^2$$

where

a_i = the number of sand grains of the particular screen interval (i.e., of the particular grain size fraction).

d_i = the average diameter of the grains equaling the average diameter of the particular screen interval opening (medium grain size diameter of the grain size fraction) in cm.

The total theoretical specific surface is

$$S_{th} = \Sigma a_i \pi d_i^2 \qquad (4)$$

The number a_i of grains per fraction is not known, however the weight P_i of a particular sieve size fraction (grams):

$$P_i = a_i \frac{1}{6} \pi d_i^3 \delta$$

where δ is the specific weight of the grains, i.e., 2.65 for silica.

Hence

$$a_i = \frac{6 P_i}{\pi d_i^3 \delta}$$

and with equation (4)

$$S_{th} = \Sigma \frac{6 P_i}{d_i \delta}$$

or per gram of sand:

$$S_{th} = 2.26 \frac{\Sigma \frac{P_i}{d_i}}{P} \qquad (5)$$

where P = the total weight of the sand sample (usually 50 grams).

For determining the theoretical specific surface quickly, the factors for the standard grain size fractions (sieve intervals) for foundry sands listed in Table 1 may be used.

The factors are the average specific surface of 1 gram of sand for each particular sieve interval, assuming that the grains are spheres.

For determining the theoretical specific surface of a particular sand, the sand weight on each screen is multiplied by the pertinent factor. The products of the multiplications of all intervals are added to give the total product, which is then divided by the total weight of the sample (50 grams). It is also possible to multiply the percentage values of each particular fraction by the respective factor and to divide the total product by 100.

Example

a) Actual Specific Surface

Sample weight = 50 grams.

Volume of sand bed = 31 cc.

$$\text{Bulk density } D = \frac{\text{sample weight}}{\text{sand bed volume}} = \frac{50}{31} = 1.61 \text{ grams/cc.}$$

$$\text{Porosity} = \epsilon = \frac{\gamma - D}{\gamma} = \frac{2.65 - 1.61}{2.65} = 0.39$$

Flow time T for a particular apparatus used = 28 sec.

From the above figures, X can be calculated, whereas the figure for L is obtained for the appar-

TABLE 1 – FACTORS FOR STANDARD GRAIN SIZE FRACTIONS

Screen Interval (U.S. Series Equivalent Number)	Factor
12	9.8
20	18.2
30	31.8
40	45.1
50	63.7
70	89.9
100	127.3
140	180.1
200	254.0
270	355.9
Pan	619.2

atus in question by multiplying the volume of the sand bed by 0.8767.

Log Y is a constant for the apparatus in question and is 2.998. Z results from T determined by means of a stop watch.

From the graphical data for a particular apparatus, the specific surface value may be read directly for the sand bed volume for 50 grams and the flow time.

In the case mentioned above, the actual specific surface is

$$S_w = 154 \text{ cm}^2/\text{g}$$

b) Theoretical Specific Surface

Sieve Interval (U.S. Series Equivalent Number)	Weight retained on screen, grams	Factor	Product
30	0.38	31.8	12.1
40	9.35	45.1	421.7
50	17.16	63.7	1093.1
70	14.30	89.9	1285.5
100	2.78	127.3	353.9
140	2.46	180.1	443.0
200	1.67	254.0	424.2
270	1.86	355.9	661.9
Pan	0.23	619.2	142.4
Total	50.00		4837.8

$$S_{th} = 4837.8 : 50 = 96.8 \text{ cm}^2/\text{g}$$

c) Coefficient of Angularity (E) (Table 2)

$$E = \frac{S_w}{S_{th}} = \frac{154}{96.8} = 1.59$$

POSSIBILITIES OF APPLICATION OF SPECIFIC SURFACE EQUIPMENT

The determination of specific surface alone can be of interest for the preparation of core sands. Pure silica sands require more binder with increasing fineness. The quantity of binder required depends on the specific surface. It must be noted, however, that the relationship exists only for pure silica sand where each grain is a single, rolled, compact crystal of quartz. There are, however, certain silica sands containing a large quantity of porous grains which require more binder than the specific surface would indicate.

In this case, the determination of specific surface, and comparison with pure and compact silica sands with identical specific surface, permits the determination of the requirements for additional binder, due to porosity of the grains and which does not act as a binder.

TABLE 2 – EXAMPLES OF ANGULARITIES OF SOME SILICA SANDS

Calibrated Glass Spheres (for comparison)	1.04
F – Ottawa Silica Sand, very well rounded	1.12
F – Lake Sand, Lake Michigan, well rounded	1.18
F – Belgian Silica Sand, subangular	1.30
F – Swiss Silica Sand, subangular	1.31
F – Swiss Silica Sand, angular	1.53
Crushed Quartz Rock, classified, 50 AFS fineness	1.58
Swiss Alpine River Sand, very angular	1.62

F = commercial foundry silica sands

The determination of the coefficient of angularity, that is of the grain shape, may be useful for the investigations of new sand deposits, or of new sands offered commercially.

Of particular interest is the combination of different testing methods with specific surface and grain shape determination, such as grain shape vs. flowability, rammability, strength, permeability, density and baked strength of core sands.

The described method of specific surface and grain shape determination can also be applied for purposes outside of foundry sand control, such as geology, sedimentary petrography, petroleum exploration, ceramics, etc.

REFERENCES

1. R. H. S. Robertson and B. S. Emödi, "Rugosity of Granular Solids," *Nature, no. 152,* Nov., 1943.
2. W. Davies and W. J. Rees, "The Effect of Grain Shape on the Molding Properties of Synthetic Molding Sands," *Journal of the Iron and Steel Institute, no. 2,* 1944.
3. W. Davies, *Foundry Sand Control,* The United Steel Co., Ltd., Sheffield, 1950.
4. J. Kozeny, "Ueber kapillare Leitung des Wassers im Boden," *Wien. Akad. Wiss.,* Sitz.-Ber. (A) 136, 1927.
5. P. C. Carman, "The Determination of the Specific Surface of Fine Powders," *Journal of Society of Chemical Industry, vol. 57,* 1938.
6. F. M. Lea and R. W. Nurse, "The Specific Surface of Fine Powders," *Journal of Society of Chemical Industry, vol. 58,* 1939.
7. P. J. Rigden, "The Specific Surface of Powders—A Modification of the Air-Permeability Method for Rapid Routine Testing," *Journal of Society of Chemical Industry, vol. 62,* 1943.
8. R. L. Blaine, "A Simplified Air Permeability Apparatus," A.S.T.M.-Bulletin, August, 1943.

AID FOR THE DESIGN ENGINEER

foundrymen! "accentuate the positive"

By Jules J. Henry

ABSTRACT

The cast process has many advantages over other methods of manufacturing. Calling the designer's attention to these "positive factors" of the cast process will aid him in producing structures and machinery that are stronger, safer and more economical. References are quoted to show the influence of design on the strength of structures and why castings can produce stronger components than other methods. Examples are shown to illustrate the superiority of the cast process, and suggestions are given on how the foundryman can help the design engineer.

INTRODUCTION

For many years, foundrymen have been fighting a battle with wrought metal manufacturers in attempting to prove that cast metals can be as good or even better than wrought metals. Although they must admit that wrought metal test bars can show better physical properties than cast metal test bars of comparable analysis, foundrymen have tried to match test bar values in the attempt to advance the cause of cast metals. It is also true that test bars can be used to evaluate the quality of the metal in a test bar; but here the apparent superiority ends.

Industry does not use test bars to carry loads or build machinery, it uses structures and components. And in these engineered shapes, cast metals can be superior to wrought metals. There has been a growing feeling in industry that test bars do not accurately predict the performance of a structure or component made of the same metal. R. J. Roark has stated "because of the many factors that influence fatigue properties . . . experimental results obtained from conventional tests on small prepared specimens do not provide an altogether satisfactory basis for engineering design."[1]

Caine has also shown that the tension test can overevaluate wrought metals.[2] Even in testing for impact strength, investigators have shown that the standard test specimen is not reliable for evaluating the resistance of structures to shock loads. It is recommended that a notched bar that approximates the size and shape of the component or structure be used, even though it is one developed specifically for the one application.[3] There is a definite trend toward testing full-sized parts under actual stress conditions of service for accurate evaluation of the component itself, as shown in Fig. 1.

Instead of fighting the battle of test bars, foundrymen should band together in a campaign to "accentuate the positive" factors of the cast process. The foundry industry must bring to manufacturers, and particularly to their design engineers, the important fact that the casting process has tremendous advantages of economic importance to give to industry that no other process can duplicate. Some of these advantages are:

1) Design freedom.
2) Reduction of stress concentrations.
3) Low end cost.
4) Uniform strength.
5) Good appearance.
6) Choice of properties.

DESIGN FREEDOM

To the design engineer, this first positive factor can be the most important. It is in this design freedom of the cast process that the design engineer can utilize

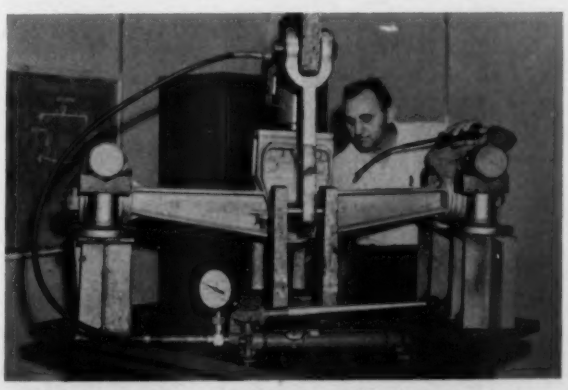


Fig. 1 — Testing of "walking beam" casting with loads simulating actual service conditions.

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Fig. 2 — (Above) Spring equalizer casting of unusual shape.

Fig. 3 — (Right) Redesign to provide maximum strength at highly stressed areas by using variable sections.

the proper distribution of metal to obtain the optimum weight-strength ratio. In cases where a non-uniform load is applied, it is wasteful to use a uniform standard section just because it is available. An example of this is shown in Fig. 1 which shows the testing of a "walking beam" used on a tandem trailer.

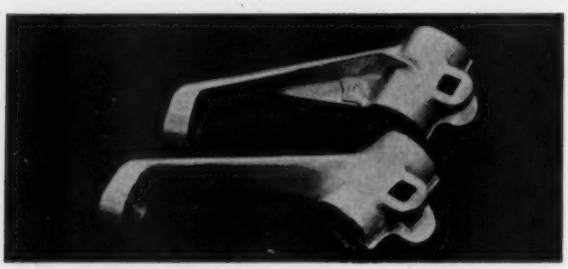


Fig. 4 — Original arm (top) and redesign (bottom).

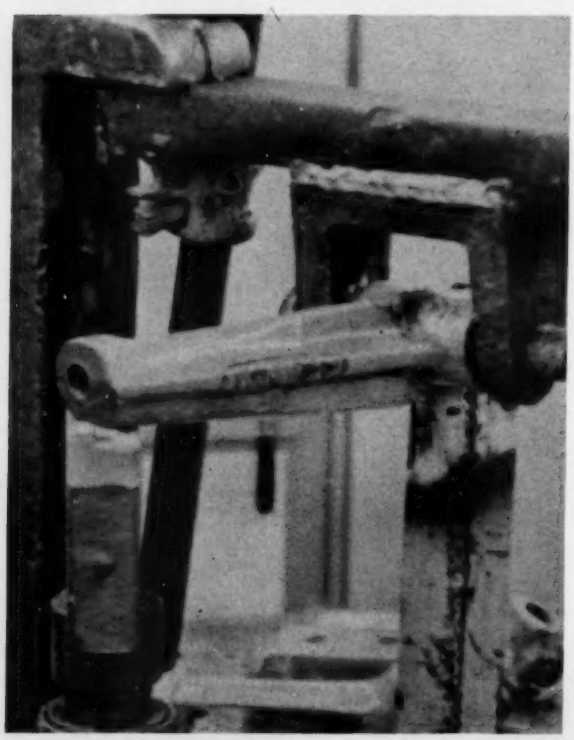
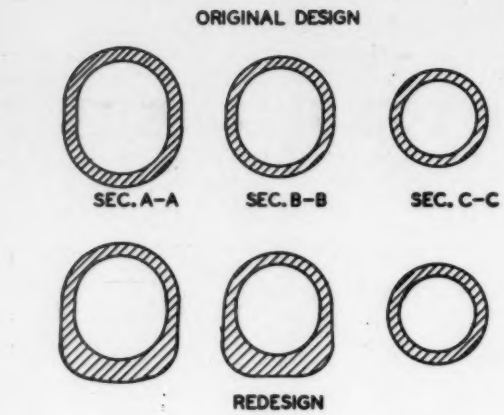
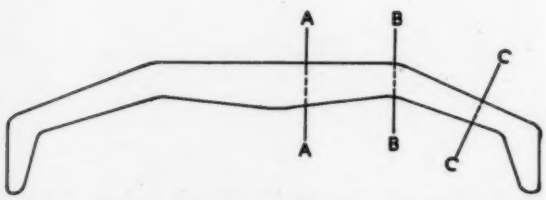


Fig. 5 — Method of loading for stress analysis of original design (Fig. 4).





Loads are applied only at the ends of the bar and at the center. It can easily be seen that maximum strength is needed at the center, so here is where the designer placed the most metal. If the beam were made of a standard rolled shape, it necessarily would be the same size as the center throughout its entire length. The casting process has permitted the designer to place the maximum amount of metal at the place it is most needed. Another example of designing to provide maximum strength at critical stress areas is given in Fig. 2, a spring equalizer used in the trucking industry.

Here again the load is applied only at the ends and center. In this case however, there may be some torsional loads applied, so the structure was made in tubular form. Because of clearance requirements, it was found necessary to bend the two arms. After stress analysis, this structure showed the presence of high stresses at the bends, so it was necessary to redesign these areas to provide the necessary strength. The design freedom of the casting process permitted the designer to use a tube of variable cross-section, as shown in Fig. 3.

The option of the individual in providing thicker sections at stressed areas, and thinner sections at unstressed areas, is a distinct advantage of the casting process that should be brought to the attention of all design engineers.

The casting process by its design freedom, can also allow the designer to utilize special shapes that will offer maximum strength and rigidity per pound of metal. This is illustrated in Fig. 4, which shows the original and the redesign of a machine part that had failed in service. Stress analysis of the casting, as shown in Fig. 5, revealed the maximum stress areas, and the component was redesigned to reduce stresses at these areas and place the metal where it was

needed. Weight of the redesign is the same as the original, but the strength is 50 per cent greater.

There are many references to this factor of design freedom in the literature and many examples have been shown from time to time.[2,4,5,6] This positive factor of the casting process has not been overlooked entirely by design engineers. Some will readily admit that "no other method of manufacture has flexibility in this respect and foundries may well exploit this."[7]

REDUCTION OF STRESS CONCENTRATIONS

This second positive factor is partially dependent on the first one. Indeed, the design freedom inherent in the cast process makes it possible for foundrymen to produce castings with reduced stress concentration areas. By making slight changes in the pattern, one can make larger radii, heavier sections, remove objectional ribs or redesign bosses. These features can all help reduce stress concentrations that are due to design.

Another aspect of the cast process that results in lower stress concentration, is casting a structure in one piece so that joining by welding is not necessary. Welding can induce residual stresses that are sufficient to cause failure, and certainly can contribute to failures at loads much lower than the safe calculated load of the structure.[8] Other causes of residual stresses in fabricated structures or components are forming operations, bending, cold working, permanent or plastic deformation, restraint of deformation, and many others. These are all absent from the casting process.

The fatigue strength of a structure can be influenced by stress concentrations due to design.[9] Sharp corners are known to be areas of high stress concentrations. Eliminations of these high stresses can be difficult in fabricated structures, but can be easily corrected in castings. The theory is illustrated in Fig. 6, which shows the change in stress concentration for a shaft with different sizes of shoulder fillets.[10] It indicates that by changing from a small fillet ($r/d = 0.06$) to a larger fillet ($r/d = 0.20$), the stresses can be reduced up to 50 per cent.

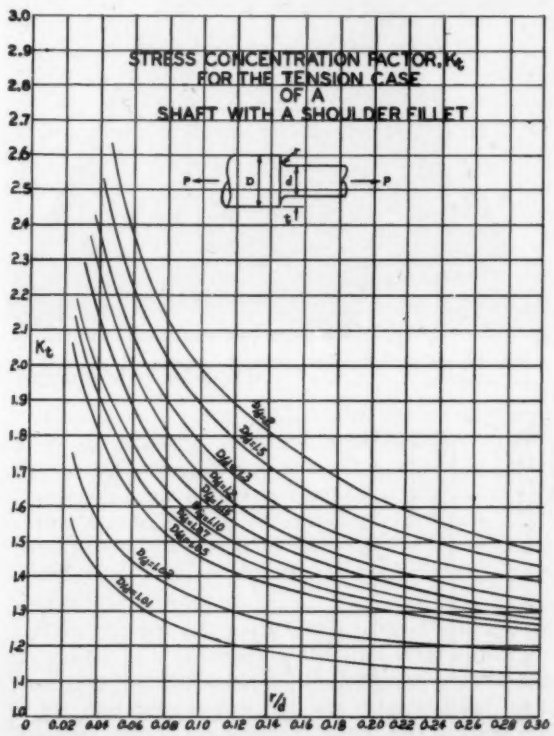


Fig. 6 — Stress concentrations as a function of fillet size (ref. 10).

In a practical case, illustrated in Figs. 7 and 8, the strength of a tractor rail was increased by 30 per cent through reduction of stress concentration by enlarging two fillets. Large radii and streamlined junctions that have low stresses are quite easy to produce in the cast process, but can be difficult if not impossible to fabricate. The optimum junction of two sections for low stresses, shown in Fig. 9, can be produced quite easily by casting, but machining such a junction would be costly and require special tools.

One factor in reduction of stress concentrations that is present in all castings, is the beneficial effect of shot blasting. Investigators have shown significant increases in fatigue strength by shot peening of fabricated surfaces. Comparison of "as received" surfaces



Fig. 7 — Tractor rail casting.

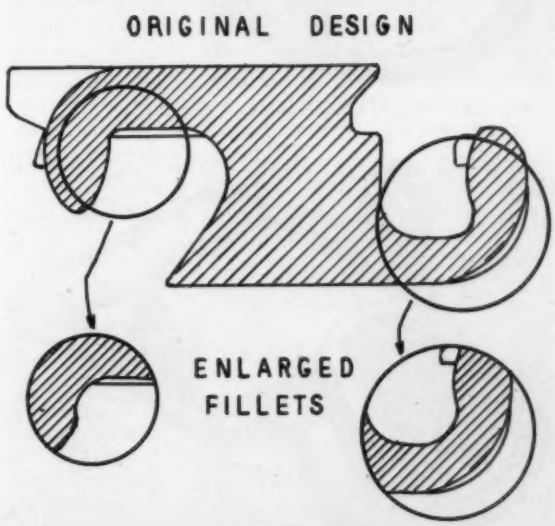
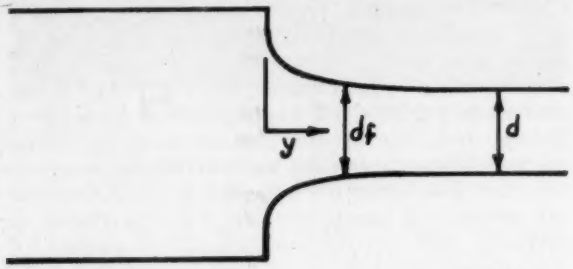


Fig. 8 — Section through tractor rail (Fig. 7) to show areas of stress concentration due to small original fillets.

and polished surfaces, showed that polishing increased the fatigue limit about 20 per cent. Shot peening an as-received finish increased the fatigue limit to the same value as for polished surfaces.[11]

Investigation of residual stresses in structures showed that shot peening increased the endurance limit of steel shafts from 26,000 psi in the polished condition, to 40,000 psi by shot blasting.[12] And this benefit the foundrymen is giving away without cost! Foundrymen: Let us accentuate this positive factor of the casting process to the design engineers.



y/d	df/d for tension	df/d for Bending or torsion	y/d	df/d for tension	df/d for Bending or torsion
0.0	1.636	1.475	0.3	1.187	1.052
0.01	1.572	1.336	0.4	1.134	1.035
0.06	1.440	1.193	0.6	1.070	1.021
0.10	1.374	1.145	0.8	1.037	1.015
0.2	1.260	1.082	1.0	1.019	1.010
			1.6	1.004	1.003

Fig. 9 — Dimensions for optimum shape of radii to provide stress free junctions (ref. 10).

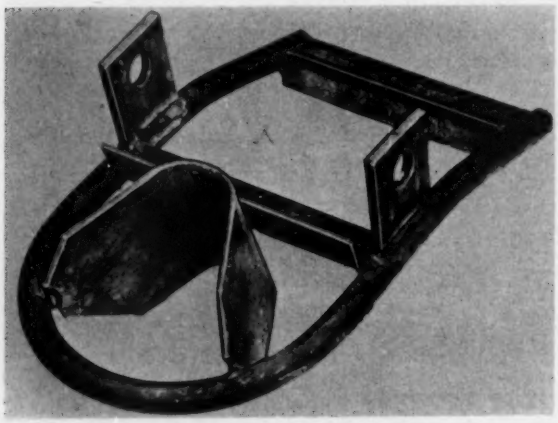


Fig. 10 — Trailer hitch weldment (ref. 4).

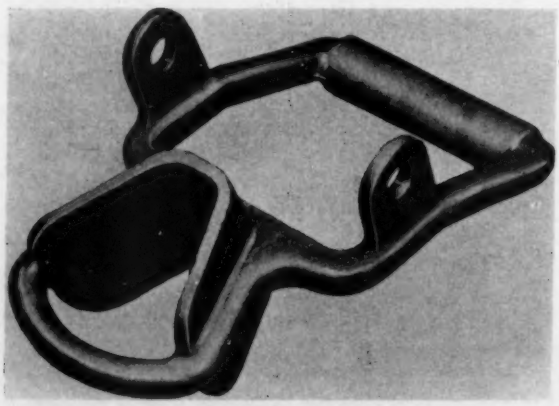


Fig. 11 — Trailer hitch casting (ref. 4).

LOW END COST

Many design engineers have been using castings for one reason only, because they are cheap. Foundrymen would rather have design engineers specify castings because they are a quality product that can perform the function desired, at a lower end cost. There is a big difference between cheap castings, and engineered cast structures that are low in end cost. Design engineers are aware of many instances where castings, when substituted for other methods of fabrication, give better performance and yet show savings.

Some examples of these are shown in Figs. 10, 11, 12 and 13, which are taken from Product Development Contests sponsored by the Steel Founders' Society of America. Figures 10 and 11 show replacement of a difficult weldment by a casting with considerable man-hour savings as well as improvement in operation. The casting shown in Fig. 12 showed a unit cost saving of 12 per cent over the weldment shown in Fig.

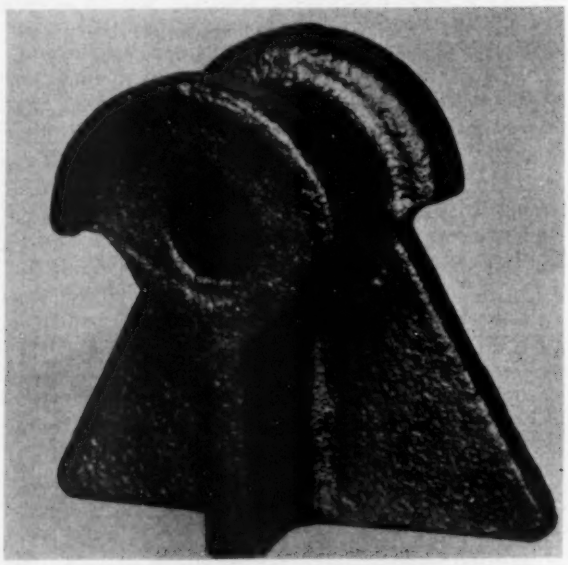


Fig. 12 — Crawler roller bracket redesigned as casting (ref. 4).

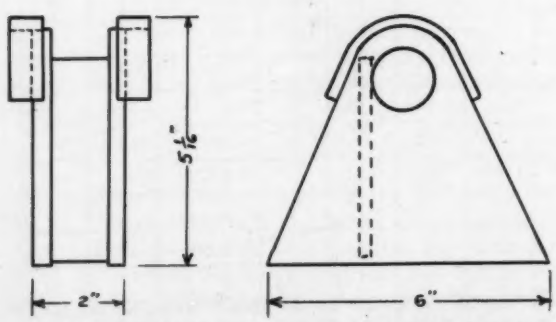


Fig. 13 — Sketch of crawler (Fig. 12) roller bracket as weldment (ref. 4).

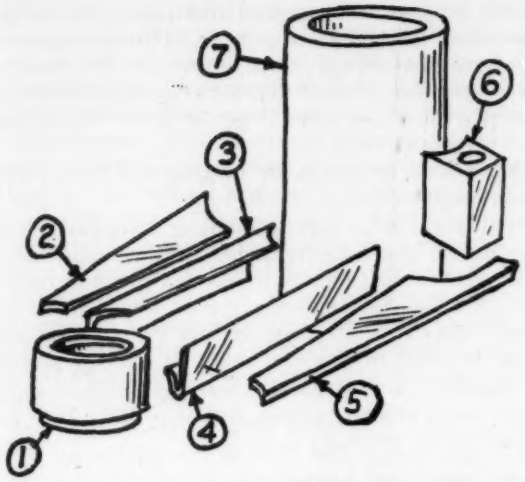


Fig. 14 — Pieces required for weldment of steering arm assembly (ref. 4).

13. In evaluating the end cost of a casting, all items must come under consideration.

The cost of the weldment in Fig. 14 was given as

Total material cost	$19.28
Welding cost	5.80
Fabricating charges	3.10
Total estimated cost	$28.18

While the cost of the casting shown in Fig. 15 was

Cast steel casting	$27.20
Handling costs	2.72
	$29.92

which appears to be higher than the weldment. But a further break down of other savings possible with a casting shows the elimination of:

16 purchasing operations.
20 receiving operations.
16 stores operations.
42 materials handling operations.
10 fabricating operations.
1 machining operation.
22 inspection operations.

Total 128 operations eliminated.

Which do you now think had the lower end cost? This then is another positive factor of the cast process: elimination of indirect labor.

UNIFORM STRENGTH

This positive factor available in the casting process at no extra cost, is one that is frequently overlooked by the design engineers. All too frequently they are influenced by the higher test bar results of other fabricating methods, without realizing that there can be considerable difference in some of the physical properties depending on the direction of rolling or forging, as the case may be.

This difference is particularly true in tensile ductility values. It is possible for the transverse direction to have one-half the value of the longitudinal direction. It has also been shown that the directionality, or anisotropic effect, for wrought steels can influence the toughness in the impact test as shown in Table 1.[13]

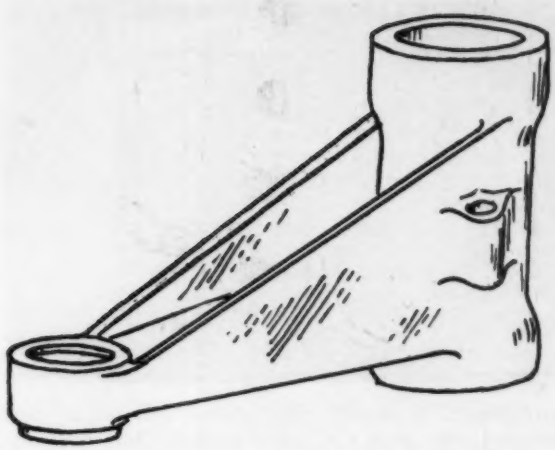


Fig. 15 — Steering arm assembly as a cast steel kingpin (ref. 4).

Foundrymen have not emphasized to the design engineer this beneficial crystal structure of the cast process. Multi-directional properties can be beneficial in structures such as gears, where equal strength in all directions is a design necessity. Fabricating gears out of a rolled plate is certain to produce teeth of varying abilities to withstand shock, impact, fatigue, etc., because of the anisotropic effect of wrought metals.

In the case of steel castings, and probably in some nonferrous castings, heat treatments are normally part of the process and so insure the absence of residual stresses that are present in most wrought metals. Here again foundrymen have been giving to design engineers stress relieving as a part of the cast process, which other processes include as an extra charge.

GOOD APPEARANCE

Streamlining. Need more be said for good appearance? Structures certainly look stronger, actually are stronger, have more eye appeal and are easier to sell when streamlined. Foundrymen should realize that full utilization of streamlining will not only please the design engineer, but also make his own job easier. Smooth flowing lines, gradual section changes, absence of abrupt corners or angles, all result in castings that are easier to produce. So, in selling this positive factor to the design engineer, the foundryman is actually helping himself.

Also, it must be brought to the attention of the design engineer that the casting process can make

TABLE 1 — TENSILE, HARDNESS AND FATIGUE TEST RESULTS FOR WROUGHT 4140 STEEL

Direction	Tensile Strength, psi	Yield Strength, psi	Red. of Area, %	Elongation, %
Transverse	110,700	81,000	29.5	13.0
Longitudinal	110,100	80,400	60.0	24.5

	Endurance Limit, psi		Endurance Ratio	
	Unnotched	Notched	Unnotched	Notched
Transverse	53,000	26,000	0.48	0.24
Longitudinal	61,000	28,000	0.55	0.25

TABLE 2 — RANGE OF PROPERTIES OF CAST METALS

	Tensile Strength, psi	Modulus of Elasticity	Hardness, Bhn
Gray iron	15,000 — 75,000	9 — 20 × 10^6	170 — 400
Nodular iron	58,000 — 150,000	18 — 25 × 10^6	140 — 400
Malleable iron	50,000 — 90,000	25 — 28 × 10^6	110 — 285
Steel	55,000 — 250,000	29 — 30 × 10^6	120 — 550
Aluminum and alloys	13,000 — 50,000	10 — 11 × 10^6	40 — 140
Copper and alloys	30,000 — 165,000	9 — 10 × 10^6	40 — 235
Magnesium and alloys	26,000 — 49,000	6 × 10^6	50 — 80
Nickel and alloys	55,000 — 285,000	21 — 31 × 10^6	90 — 350

streamlined structures much easier and much more economically than any other process. Only the forging process, of all the other manufacturing methods, can approach the cast process in ease of producing well streamlined structures. Reference to most of the preceeding photographs certainly emphasizes the better appearance of castings.

CHOICE OF PHYSICAL PROPERTIES

This factor is not one of the major advantages of the cast process, since other methods can offer a great variety of physical properties at an extra cost, if the quantities are sufficiently large. The cast process, however, can offer a much greater choice even in small quantities if needed. A great variety of chemical analyses can be procured from producers of castings in all types of metals.

Mechanical properties in the ranges given in Table 2 are available.

This list is by no means complete, for added to this can be the metals specially developed to resist corrosion, heat shock, abrasion, creep, to mention only a few. Also, design engineers should know about the many special metals that can not be wrought, and must be made in the final shape by the cast process.

Foundrymen: you have so many positives to accentuate about the cast process that it is difficult to put an end to the imposing list.

AIDS FOR THE DESIGNER

Let us discuss a few specific ways that the foundryman can offer positive help to the design engineer. In the matter of design freedom, foundrymen can help the design engineer by suggesting ways to improve the design while it is in the development stage. This is not always easy, since many design engineers are behind an iron curtain created by purchasing agents. But it is possible to have meetings with engineers on design problems, in fact there are some foundries who have scheduled casting design clinics to which they have invited their customers' engineers.

This is a potent method of demonstrating the design freedom inherent in the cast process to the persons who will actually be using it. Figures 16, 17 and 18 illustrate some of the many complex designs that can be shown to design engineers as examples of design freedom.

Caine has stated that one of the greatest increases in casting usage can be created by supplying the design engineer with information on the properties of irregular sections.[15] There are many tables giving properties of regular sections, published by wrought metal manufacturers, and these are within reach of every design engineer. So it is too convenient for



Fig. 16 — (Above) Cast steel compressor cylinder (ref. 4).

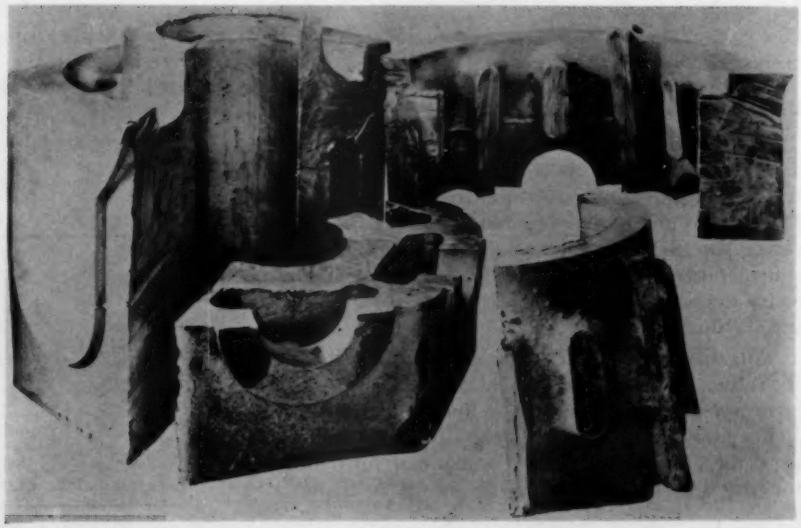


Fig. 17 — (Right) Compressor sectioned to show complex internal structure (ref. 4).

him to pick up a handbook that contains all the information that he needs, in designing a wrought structure. But if he can use irregular sections to better advantage, unfortunately it becomes necessary for him to calculate all the properties in question.

Foundrymen: you should provide the design engineer with this information. Perhaps handbooks prepared by the various national foundry organizations may be the answer. This means of course, that foundrymen must become familiar with the terms used by design engineers. While it is not necessary to be a mechanical engineer in order to understand the meaning of Youngs Modulus, one must become familiar with commonly used design terms, such as central axis, fiber stress, neutral axis, moment of inertia, unit stress, etc.

Just understanding their terms and language will help appreciably in dealing with design engineers. Therefore, one important item in aiding him is to provide technical information showing properties of irregular shapes that can be used to obtain the best weight-strength ratio in his design.

Another rather new area for the foundryman to probe in helping the design engineer, is to investigate stress concentrations. Establishment of a stress analysis laboratory by casting producers may seem to be a little beyond the normal scope of foundry operations. However, closer examination of this new field will show that foundries are the most logical location for stress analysis laboratories. Foundries are normally dealing with hundreds of different designs each year, while a manufacturer of machinery or other equipment may have a new design only once a year.

For a small manufacturer a stress analysis laboratory is practically out of the question, yet he can gain appreciable benefits from a stress analysis of his designs. Of course, that does not mean that foundries should do the original design work, but they ought to be in a position where they are able to check their customers' designs for unforseen stress concentrations. This is exemplified in Fig. 19, which pointed out stress concentration areas to the design engineer. These stresses were substantially reduced, as shown in Fig. 20.

Lower end cost of castings can also be an area where foundrymen can offer valuable assistance to the designer. Here again, the use of a stress analysis laboratory can be of great value in revealing areas where weight can be safely reduced. Also the use of cores and uneven partings (which the foundryman would rather not see in castings) can, in many cases, reduce the end cost by decreasing weight or eliminating machining. Some foundrymen have been guilty of forcing simplification of the design in order to obtain as much weight as possible, without regard to the final end cost to the customer.

This is a short-sighted attitude, since simple masses of metal are easier and cheaper to produce by other methods and will eventually be lost by the foundry industry anyway. The intelligent approach is to help the design engineer reduce his end cost by making the structure in the most economical way, and in a shape that no other method can successfully duplicate. Remember that complex shapes, with cores and variable sections, can only be produced by the cast process. The designs, illustrated in Figs. 2, 7 and 16, show castings that can not be produced in any other way without excessive cost.



Fig. 18 — Suction nozzle end head for barrel pump (ref. 4).

Uniform strength that is inherent in the cast process, is dependent on the foundryman's individual concept of quality. Consistent analyses and heat treating, together with the absence of surface defects in stressed areas, will assure the design engineer of obtaining the anticipated strength in his structure.

One of the most outspoken adherents of better appearance in castings has been Clyde Sanders, who has



Fig. 19 — Spring equalizer, showing stress concentration areas.

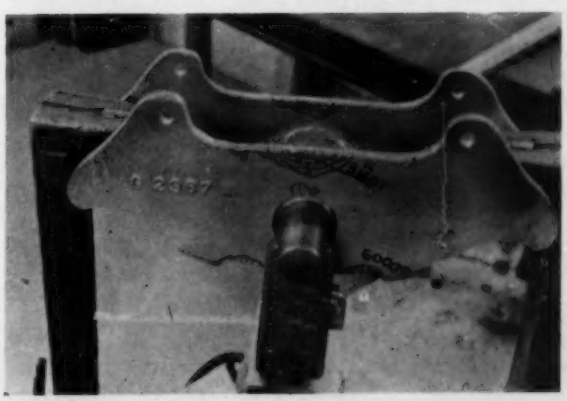


Fig. 20 — Spring equalizer, suggested redesign to reduce stress concentrations.

repeatedly urged foundrymen to use finer sands and closer sand control to improve surface finishes.[16] One of the best ways to turn a design engineer away from castings is to have a roughly finished, poorly cleaned casting in a prominent place on his machinery. Foundrymen can help the design engineer by finding out which surfaces are important in regard to appearance, and then take more effort to make the engineer proud of his accomplishment.

Data sheets of the various alloys and physical properties of the metals produced by each foundry can be of considerable aid to the design engineer. Quite frequently the design engineer may not specify the best alloy for a particular design, simply because he was not aware of the great variety of physical properties the foundry can produce. Cooperation with the designer in selecting the proper alloy or heat treatment is beneficial to the foundry, as well as to the design engineer.

As a final suggestion, think of the wonderful mutual benefits that could result from the foundryman receiving information on static and dynamic stresses, their direction and magnitude, on all drawings for all customers' designs.

SUMMARY

The foundryman has much in the cast process to recommend to the design engineer. Calling his attention to the positive factors of the cast process will aid him in producing structures and machinery that are safer, stronger and more economical. Some of the beneficial factors of the cast process and methods that the foundryman can use to help the design engineers are:

1) *Design freedom.* Provide tables containing information on the properties of irregular sections that can be most economically produced by the cast process.
2) *Reduction of stress concentrations.* Install stress laboratories in order to evaluate designs for possible high stress areas.
3) *Low end cost.* Demonstrate how castings can provide the best weight-strength ratio and eliminate some machining, joining and handling.
4) *Uniform strength.* Install quality control methods, and give particular attention to highly stressed areas.
5) *Improved appearance.* Use best materials and methods available to produce smooth castings with streamlined features.
6) *Choice of properties.* Provide design engineers with data on types of metal produced, and methods of obtaining variations in physical properties.

REFERENCES

1. R. J. Roark, *Formulas for Stress and Strain,* McGraw-Hill, New York, p. 42, 1954.
2. J. B. Caine, "The Case Against the Tension Test," *Foundry,* Nov. 1957 and Sept. 1958.
3. S. L. Hoyt, "Notched-Bar Testing, Theory and Practice," *Symposium on Impact Testing,* A. S. T. M., p. 5, 1955.
4. Product Development Contest Bulletins 1954, 1956, Steel Founders' Society of America.
5. S. C. Massari, "Marketing Your Product," AFS TRANSACTIONS, *vol. 64,* p. 1, 1956.
6. J. M. Tarnow, "What the Designer Should Know About Production — Part 3," *Machinery,* Nov. 1958.
7. T. Davidson, "What Design Engineers Look for in Castings," AFS TRANSACTIONS, *vol. 63,* p. 282, 1955.
8. L. E. Benson, "Some Considerations Regarding the Generation and Importance of Residual Welding Stresses," *Residual Stresses in Metals and Metal Construction,* Reinhold Publishing Corp., New York, p. 76, 1954.
9. W. T. Bean, Jr., "Endurance — A Criterion of Design," *Symposium on Testing of Parts and Assemblies,* A. S. T. M. Technical Publication 72.
10. R. E. Peterson, *Stress Concentration Design Factors,* John Wiley & Sons, Inc., New York, p. 65-67, 1953.
11. R. P. Felgar, "Effect of Shot Peening on Fatigue Strength," A. S. M. E. Paper No. 58-SA-46, p. 1, 1958.
12. O. J. Horger and H. R. Neifert, "Correlation of Residual Stresses with Fatigue Strength of Machine Elements and Related Phenomena," *Residual Stresses in Metals and Metal Construction,* Reinhold Publishing Corp., New York, p. 248, 1954.
13. E. B. Evans, S. J. Ebert, C. W. Briggs, "Fatigue Properties of Cast and Comparable Wrought Steels," A. S. T. M. *Proceedings,* p. 14, 1956.
14. "Properties of Materials," *Materials in Design Engineering,* p. 30-123, Oct. 1958.
15. J. B. Caine, "How To Deal With Customers Engineers", Talk given at Div. VII, Steel Founders' Society of America Meeting, Sept. 5, 1958.
16. C. A. Sanders, "Casting Finish — Tolerance — Precision," AFS TRANSACTIONS, *v. 64,* p. 318, 1956.

HOW AIRCRAFT DESIGNERS LOOK AT LIGHT METAL CASTINGS

By Albert R. Mead

ABSTRACT

Light-metal castings could be used in modern aircraft design more than they are at the present time if they can be developed to the point where they meet strength requirements of 60 ksi ultimate strength, 50 ksi yield strength and 5 per cent elongation, in the author's opinion.

Flight loads which must be considered in aircraft design are:

1) Maneuver loads.
2) Gust loads.
3) Flutter.
4) Pressurization loads.
5) Power plant loads.
6) Sonic fatigue.

Besides these flight loads, nonflight loads must also be considered. Those which are considered are for carrier-based aircraft.

The problems with various light-metal alloys in the manufacture of aircraft are considered.

INTRODUCTION

As a prime contractor for the U.S. Navy, the author's company is responsible for the design and manufacture of Naval aircraft. The company is awarded a contract by the contracting agency after a review of a design proposal, which is essentially a detailed specification and basic drawings. The contract or letter of intent will include the items to be produced and the information required, such as analyses, drawings and tests.

The detailed specifications prepared according to the latest issue of a General Specification for the design and construction of airplanes for the U.S. Navy. When a contract or letter of intent has been received the engineering team goes to work calculating the strength required for catapulting, arrested landings, flight loads, nonflight loads, hoisting, tanks, fasteners, materials, etc.

Now that the general outline of the aircraft is fixed, weight, proper strength, ease of manufacture, reliability, ease of servicing and many other factors too numerous to mention become of importance.

An example of how weight is of prime importance is exemplified by this information.

The approximate effect of an additional 100 lb over design weight on performance for a swept wing Naval fighter is:

Max. speed	= −1 knot
Combat radius	= −2 miles
Combat ceiling (500 ft/min R/climb)	= −120 ft
Rate of climb	= −160 ft/min

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Ease of manufacture and cost of production come into play early in design, and here is where the designer considers light metal castings. However, before light metal castings are discussed in detail some of the loads which must be considered in the design of an aircraft will be presented.

LOAD DESIGN CONSIDERATIONS

Maneuver Loads. When an airplane is flying level, the intensity of the vertical pressure on the wings is just sufficient to balance the weight of the airplane. If the aircraft deviates from level flight and takes a curvilinear course, the balance is interrupted and an acceleration results. The resultant accelerations cause loads which must be considered in design. The strains induced on the structure due to other than level flight are related in a linear manner to the accelerations of the aircraft as measured at the center of gravity.

Gust Loads. Along with maneuver loads, gust loads must be evaluated. A simple way of presenting gust statistics is to indicate the number of times a particular gust velocity is exceeded in a certain flight path. High altitude gust loads are much less severe than ground level gust loads.

As many as 10 gusts may be encountered in each mile of flight. Flutter must be considered, but here a basic definition of flutter is in order because of the complexities of this phenomenon, the author believes.

Flutter. This is a self-excited oscillation of a structural part resulting from a combination of inertia, elastic, aerodynamic, damping and temperature forces. In combination these forces can result in unstable motion of the structural part, not the airplane, (i.e. flutter) which leads to mild or extremely severe structural failures. Flutter is usually restricted to control surfaces, etc.

Pressurization Loads. These must be considered due to the rather large loads produced in cockpit and fuselage areas from pressure differentials (between pressurized cockpit and outside air). Pressurization loads have caused some much publicized serious crashes in foreign aircraft.

Power Plant Loads. These are important especially when any vibration is present, and are thoroughly checked by Engineering personnel.

Sonic Fatigue. This is another type load which has

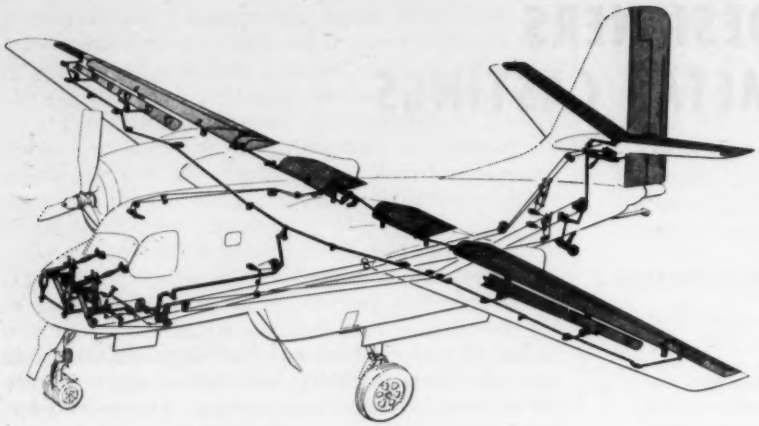


Fig. 1 — An example of one system in a modern aircraft where castings were selected for use. Most of the castings are used in the controls system.

caused serious problems, especially in jet aircraft. High frequency noise impinging on aircraft structure must be either eliminated or the structure must be sufficiently strong to withstand the loads created by the noise.

The flight loads mentioned are of prime importance, but we also should mention the nonflight loads which cause the designer and stress analyst considerable concern. Since the author's company builds a large number of carrier-based aircraft, these will be concentrated on. They include:

1) High sink speeds associated with arrested landings (over 23 ft/sec).
2) Towing loads which often times are much higher than those of land-based planes.
3) Securing loads. High loads can result due to high wind-over-deck velocities.
4) Hoisting requirements for frequent and rapid hoisting cause high loads on structure and hoisting attachment areas.
5) Catapulting. Large loads due to rapid accelerations in catapult take-offs place much importance on this load criteria (approx. 4G loads).
6) Arresting. Large loads, which the airplane is subjected to in arrested landings, play a major role in designing carrier-based aircraft. Consideration must be given to arrested landings at maximum gross weight due to emergency arrestations.
7) Wing folding loads are not severe when the folding takes place in a hangar, but on board the carrier with high wind-over-deck velocities, wing folding loads are high.
8) Crash loads. High loads which are developed when an arresting hook or cable breaks and the airplane goes into the barricade must be accounted for in design.

A few other loading conditions that exist for carrier-based aircraft are considered, but they will not be considered in detail here.

After due consideration is given to the loads previously mentioned, the designer goes to work to see where he might combine bits and pieces into one piece of structure (if this proves economical). Here is where casting is considered as a possible process for making parts. An excellent example of one system in a modern aircraft where castings were selected is in one of the company's airplanes. Most of these castings are used in the controls system as bellcranks, as illustrated in Fig. 1.

When this airplane was first considered, a serious lack of machine time was evident on the eastern coast of the U.S. because of the mobilization in progress due to the Korean War. The airplane had many bellcranks, arms, etc., which could be made into castings, so a detailed engineering investigation was started to make full use of castings.

Castings have been used for many years in all industries for three major reasons:

1) Low cost—tooling costs (dies vs. patterns) are much lower, and small quantities can be made economically.
2) Intricate shapes can be made which are not possible by any other process.
3) No wide differential between transverse and longitudinal properties exists in castings as can exist with forgings, extrusions and rolled shapes.

AIRCRAFT CASTING USE

Control system castings are often designed on a stiffness basis, and as a result the stresses in these parts were comparatively low. This was a point in favor of the castings since the design allowables for castings available in the early 1950's were:

	220-T4	356-T6
F_{tu}	42,000 psi	30,000
% Elong.	12	3

220-T4 was first selected because of the 40 per cent higher ultimate tensile strength. After the many parts were properly designed, the production lots of 220-T4 castings were received and serious inconsistencies in quality were found.

In order to insure a fair x-ray standard for acceptance of production parts, the following procedure was used to arrive at a realistic quality control.

1) Select the three poorest looking x-ray prints from the first production lot of castings which pass the minimum x-ray soundness standards based on experience and specifications.

2) Machine the three parts selected under (1) to finished dimensions, then re-x-ray and statically test to failure. The loading duplicates service conditions.
3) If the static failing loads are satisfactory after engineering evaluation, the x-rays for the three castings tested would then be used as a minimum x-ray acceptance standard.

This inspection standard served initially to accept castings for production airplanes. As production rates increased, the number of rejected parts increased considerably. A thorough cost evaluation was made and the high cost of rejections, etc., caused management concern. As production increased, quality control by the producing foundries decreased, so it became necessary to re-evaluate the company's position.

About this time American-Brake Shoe Co. of Mahwah, N.J., requested that the company work with them on a joint high integrity casting program in order to develop high strength reliable aircraft castings. This work is described in the February 17, 1958 issue of *Product Engineering*.

For general information, the stress at 100 per cent limit load (the highest load expected in the service life of the aircraft) was calculated as follows: the critical stress at 100 per cent limit load was defined as 14,400 psi. This value was arrived at by dividing a factor of 1.3 into 75 per cent of yield stress of the alloy being considered. (In the original design, the values of 220-T4 were used.) The specified minimum tensile strength for 220-T4 was 42,000 psi with a typical yield strength of 25,000 psi.

A few shortcomings of specifications were resolved due to the work of many foundry and aircraft materials people, who found that most aircraft designers were reluctant to use castings in large numbers because wide variation in strength and soundness from casting to casting was evident, and a higher level of strength could be obtained with wrought materials. In addition, loose specifications permitting a large differential between test bar and casting properties make the use of a large casting factor necessary. All three factors tend to increase the weight of the part. Increased use of castings in the aircraft field would occur if uniform and consistent high strength castings were available.

Two factors in existing specifications which cause concern are:

1) Separately cast test bars vs. test bars cut from castings.
2) Lack of correlation between x-ray properties and mechanical properties.

Both problems have an important bearing on the thinking of design engineers with regard to the use of castings.

CUT TEST BARS FROM CASTINGS

Aluminum aircraft castings are commonly produced in conformance with Federal Specification QQ-A-601 and MIL-C-6021B. These specifications dictate that, after heat treatment, a separately cast standard 0.505 in. diameter test bar of 356-T6 shall possess mechanical properties of 20,000 psi yield strength, 30,000 psi tensile strength and 3 per cent elongation. However, these specifications permit the average of three test bars cut from the castings be as low as 20,000 psi yield strength, 22,500 psi tensile and ¾ per cent elongation. Under these specifications, one or more bars could have zero per cent elongation, so long as the average is ¾ per cent.

The designer, unfortunately, is using the casting and not the test bar. Now however, the new specification for high strength aluminum alloys (MIL-C-21180A) is a major step forward in overcoming some of these problems.

On one aircraft the author's company uses several light metal castings as well as a few ferrous castings. The most complex castings are made from magnesium alloy AZ63, and make up the slat core (a slat is used to create a flow over the airfoil which in turn lowers the stall speed of the aircraft). This is important for high speed, carrier-based aircraft. Other planes in the same speed range of this airplane require long runways or a tail chute when landing. An aluminum alloy skin is put on the aerodynamic side of this part, and a magnesium sheet on the inside (where magnesium is used 5056 rivets are put in), and care taken to properly insulate all joints involving dissimilar metals. The casting method of manufacture saved a large sum of money for this particular assembly.

Alternate methods of manufacture were:

1) Bits and pieces.
2) Complete casting (core and skins).
3) Honey comb.

Figure 2 shows two of the alternate types of manufacture.

The honeycomb structure was the lightest, but was expensive to manufacture. The complete casting was also expensive, and it was difficult to protect against rain erosion.

Castings are often considered as replacements for welded structure. Figure 3 shows a welded steel jato rack (for an amphibian) which was replaced by a casting. The jato system was incorporated in this aircraft to help shorten take-off runs with large loads



Fig. 2 — Two of the alternate types of manufacture for a slat core. Top — the honeycomb structure, which although is lightest is expensive to manufacture. Bottom — the complete casting (core and skins) which is difficult to protect against rain erosion.

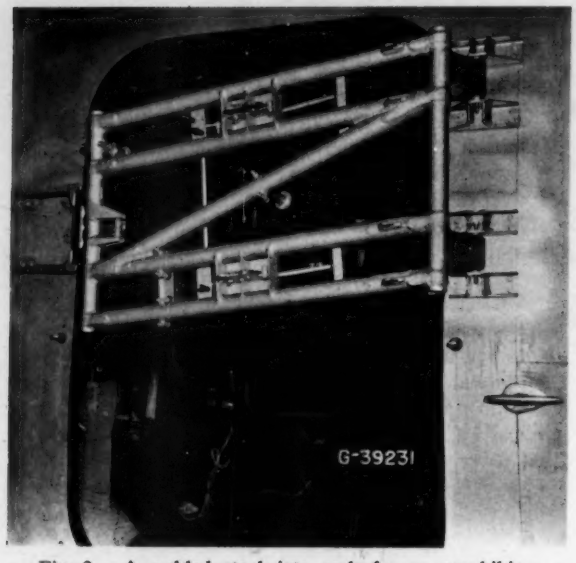


Fig. 3 — A welded steel jato rack for an amphibian which was replaced by a casting.

aboard, help take-off in rough seas, help take-off from ice covered runways (some of these airplanes are triphibians) and to help steer the airplane in high winds.

The addition of the jato system has helped immeasurably in the performance of the main mission of this aircraft, i.e., air sea rescue.

The initial weldments of normalized 4130 steel were satisfactory from a strength standpoint. However, manufacture was difficult due to warpage, so a casting was designed for a magnesium alloy. Then the magnesium alloy casting was changed to 356-T6 before the first production castings were made. There has been no service problems with this part to date.

On one of the company's airplanes, the trailer was an application for a rather heavy set of load distributing links which were originally considered as either a steel weldment or a steel casting. Due to the design concept of this light weight trailer, aircraft thinking in regard to weight and high efficiency of construction materials were paramount. The two castings, shown in Fig. 5, were first manufactured from 356-T6 alloy, but due to the close design philosophy (i.e., small casting factor) the static tests on these parts were so marginal that an alloy change was made to Al Mag-35 (used in the as-cast condition). The extra strength and better ductility of this alloy enabled the static test to be successfully completed. An interesting point should be made here which might prove helpful in the future. If cooling from the casting temperature of Al Mag-35 is improperly controlled, a two-phase microstructure will result which will be responsible for a loss in yield strength and ductility.

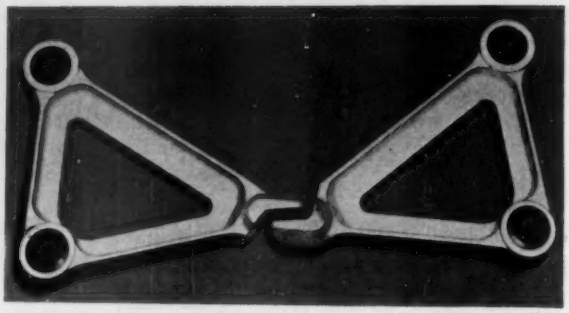


Fig. 5 — Trailer castings.

Recent aircraft industry problems relating to aluminum-magnesium Alloy 220 have resulted in the issuance of official letter from the Bureau of Aeronautics requesting the number and position of all 220 alloy castings in airplanes produced during the last 5 years. This does not mean to give the 220 casting alloy a black eye, for this alloy has potential use, but careful consideration must be given to any sustained stress applications. In 220 casting alloys the aluminum-magnesium constituent is highly anodic to both aluminum and the aluminum-magnesium solid solution. Attempts to overcome this shortcoming required additions of manganese and chromium in small amounts. Most of the cases of partial and complete failure reported to date were caused by two major and one minor factors. They are:

1) Applications which made possible the over-stressing of parts: To be specific, 95 per cent of the trouble occurred when bolts were used to attach rod end bearings to ears on the castings. Bolts were inserted and a specified torque was given to personnel putting together the assemblies. If care was not taken even with the use of torque wrenches, it was easy to over-torque the bolts, and thus, high sustained stresses caused failure by stress corrosion. (These overstressed areas were also aggravated by fatigue and operating loads.)
2) Pressed in bushings and bearings: Due to manufacturing tolerances, etc., high interference fits caused a few cases of this type of high sustained stress which ultimately led to failure.



Fig. 4 — A cast jato rack after failure from load static test.

3) Straightening or force fitting of 220 castings caused a few cases of stress-corrosion cracking that the Bureau of Aeronautics became aware of. Straightening of the casting caused localized precipitation along planes of slip, and these localized areas of highly anodic material led to early failure.

It is important to be sure that any 220 sand castings designed for possible use in a marine atmosphere be of a size and complexity which will allow as rapid a quench as possible. Susceptibility to stress-corrosion cracking of 220 sand castings is decreased considerably as the rate of quench is increased.

In 1950 the author's company received a contract to prototype and produce a fighter aircraft which would have the ability to sweep its wings in flight. The fittings which were designed to allow the loads to be properly transferred to a central point required very large 2014 forgings.

Figure 6 is shown to give an idea of the large volume of metal which had to be used to produce a fitting of approx. 225 lb. Castings were originally considered, but at that time unreliability, low guaranteed mechanical properties and space considerations were deterrents to the use of a casting.

If the casting allowables which are available today were available in 1950, the part shown in Fig. 6 might have been a casting. A 2014 hand forging was selected at that time because of the superior mechanical properties.

Hand Forged – 2014		
Longitudinal	60 ksi Ult.	48 ksi yield, 5% elong.
Long trans.	..57 ksi	48 ksi yield, 2% elong.
Short trans.	..54 ksi	48 ksi yield, 1% elong.

Then hand forging was completely machined into a fitting weighing approx. 225 lb. To make this part from bits and pieces, 39 large individual parts would have had to be used. The cost of the material in 1950 was: $1.75 per lb, or $6,125.00 for the billet. Since only a small portion of the metal remained, approx. $5,700.00 worth of chips were generated per fitting exclusive of the expensive machining costs.

Of course the author's company could not continually make these parts from hand forgings, so a die was cut (costing $62,000), and die forgings were made which still needed 100 per cent machining (Fig. 7).

Due to low short transverse mechanical properties it was necessary to put a good margin on any attachments which loaded this part in the short transverse direction. This point is brought up because many engineers feel that castings are the only type parts which contain defects. As we know here this is not true.

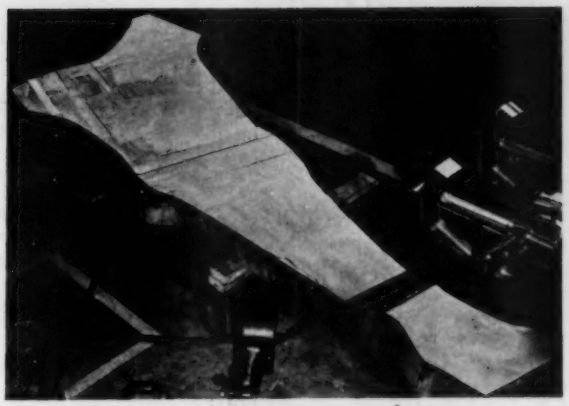


Fig. 6 — Rough cut forging.

A high percentage of wrought products used at the author's company are inspected by ultrasonics. In order to acquaint those not familiar with the reflectoscope, Fig. 8 will help to understand this valuable inspection tool.

The reflectoscope is used by the author's company to inspect steel and aluminum products. For contact testing, a smooth surface is machined on one side of a large billet so that the short transverse grain direction is parallel to this surface. A light machine oil is then applied to the surface to provide intimate contact between it and the quartz searching crystal. The average crystal size is 1 sq in., and has applied to it a frequency of 5 megacycles. A short, high frequency electrical pulse is transformed by this crystal into mechanical vibrations which are transmitted into the billet being tested.

Mechanical echoes come back through the billet and are converted again to electrical pulses, which are then amplified and applied to a cathode ray tube. A pattern is formed on the face of the tube, and any defect which interrupts the echo appears as a blip. The bigger the defect the higher the height. Both contact and immersion sonic testing are used.

The author's company has spent a large sum of money developing a proprietary set of curves which show the reduction in strength due to the defects found by the reflectoscope. Several aluminum produc-

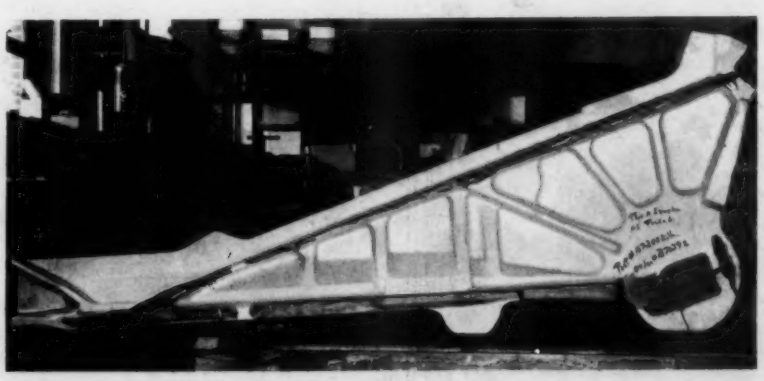


Fig. 7 — A die forging. These parts still needed 100 per cent machining.

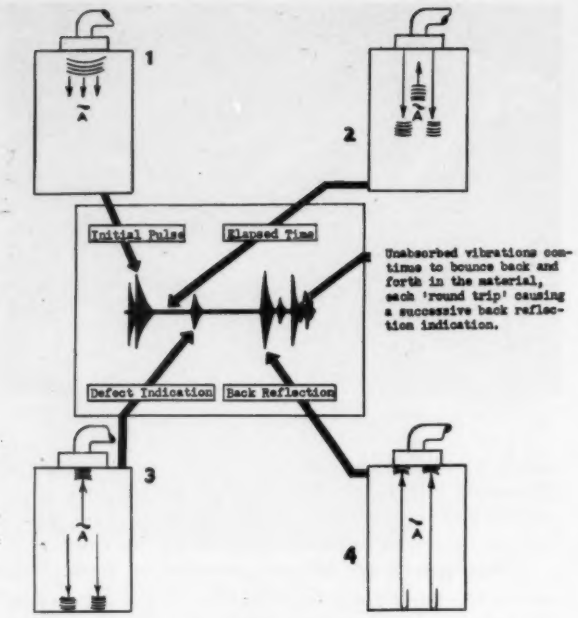


Fig. 8 — Principles of reflectoscope operation for ultrasonic testing of wrought parts. 1) Electrical pulses are transformed into ultrasonic vibrations and transmitted to material. The portion of the electrical pulse delivered to the cathode ray tube causes an initial pulse indication. 2) A portion of the vibrations traveling through the material strikes the defect (A). The sweep travel indicates the time elapsed since the vibrations left the searching unit. 3) The portion of the vibrations which struck the defect (A) bounces back to the searching unit and is transformed back into electrical pulses causing appearance of a vertical indication on screen. 4) The remainder of the vibrations travel to the bottom of the part and are reflected back to the searching unit. The indication of their return is known as the first back reflection indication.

ing companies feel that the company is too conservative in the analysis of ultrasonically located defects vs. strength reduction, but experience has taught the company to be so. Serious trouble in 1948 with die forged 2014-T6 landing gear outer cylinders required the limiting of the maximum operating stress at the parting line to 20 ksi, although it should be mentioned that the last few years have brought about tremendous quality improvements in wrought aluminum products.

A high strength aluminum casting was considered for the pylon stores fittings for the airplane. This fitting serves the purpose of holding a weapon or a 150 gallon fuel tank on the bottom side of the wing. A 2014-T6 forging was first considered and found to be rather heavy and bulky, so a hypothetical casting was next analyzed. The casting actually proved to be lighter due to the cored construction. As the company was preparing to proceed further with an aluminum 356-T6 casting, the analysis indicated two areas on the casting which would be marginal, so an aluminum casting had to be discarded for this part. A flash welding method of fabrication was next considered but the tooling costs were prohibitive. The part finally ended up as a 4330 steel casting which is working out well in production (Fig. 9).

In this case aluminum casting alloys did not have sufficient strength. There are many parts on an airplane which would be considered as castings if more strength could be obtained in light metal castings. These include the many bulkhead fittings which weigh 30-40 lb, which are currently being made from either 7075-T6 or 2014-T6 wrought alloys.

The ultimate possibility in the potential use of castings would be in areas such as the under side of wings. These parts are presently being made from thick plate material and machined all over. As an example, here are a few statistics on the manufacturing details of the lower wing skins of one airplane. For this airplane 1⅝-in. stretched 7075-T6 aluminum plate measuring approximately 16 ft x 4 ft and weighing approx. 1500 lb is used. The plate is ultrasonically tested, then milled into an integrally stiffened part weighing approx. 200 lb. The machining takes about 30 hr on a large skin miller.

The 30 hr would not be so bad, but the availability of these machines is limited. If, some day it would be possible to make parts like these from castings, huge savings in machinery, manpower and dollars would result. As a point of interest concerning these types of parts, a large Army transport aircraft made by milling plate generated 15 tons of scrap (floors and skins).

In conclusion, it is felt that castings, if developed to the point where they can meet strength requirements of 60 ksi ultimate strength, 50 ksi yield strength and 5 per cent elongation, will find wide use in future aircraft.

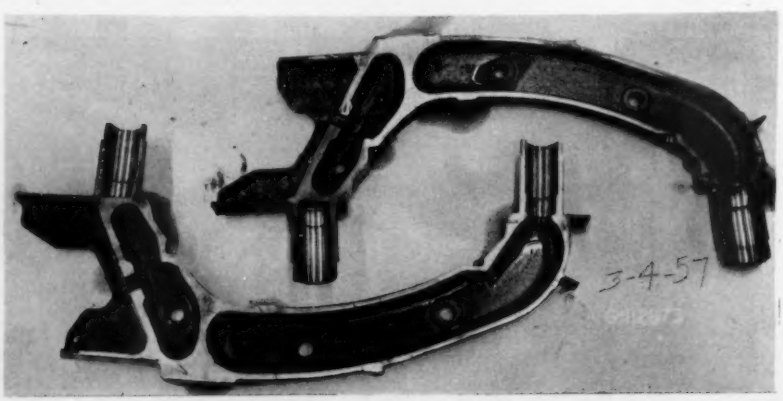


Fig. 9 — Sectioned pylon casting.

DESIGN ENGINEERING AS RELATED TO MAGNESIUM CASTINGS

By George H. Found

ABSTRACT

This is a conceptual treatment of how to go about designing in magnesium castings. It treats the subjects of structural and casting design separately, and then attempts objectively to relate the two. The results when applied to magnesium can be a rather unique situation, where the best features from both structural and casting standpoints become indistinguishable.

The optimum circumstance is one where the service performance can be described and related to available mechanical property data. The resulting structures can then be evolved into configurations which quite coincidentally are preferred for best foundry practices.

INTRODUCTION

The purpose of this paper is to aid in the appraisal and designing of magnesium castings for the aircraft and the broadening missile and electronics fields.

Several methods might be used to treat this subject.

a) A simple statistical comparison of the abundant mechanical property data of the various magnesium and competitive casting materials.

b) A more involved statistical comparison of these data. In this approach the sensitivity of these data to foundry and other quality variables would be considered together with putting emphasis on those data most directly related to the type of design loads that are most critical, whether they are sustained (or creep) loads, dynamic (or fatigue) loads, impact loads, elevated temperatures, corrosive environments, etc.

c) A treatment of certain concepts involved in designing where the characteristic features of magnesium, particularly its physical properties, may be employed to advantage, thus complementing mechanical property comparisons.

Data describing (a) and (b) are readily obtainable from several sources, including publications of The Dow Chemical Co., and will not be considered here.[1,2,3] This paper will contemplate, as in (c), the design opportunities and design practices suggested by the physical properties of magnesium, particularly its density and amenability to highly simplified design.

Since one of the objects of this paper is to recommend best design practices for magnesium, it is helpful to reflect upon what is meant by "good design."

WHAT IS "GOOD DESIGN"?

Good design implies a structure which gives best service performance with least weight, while still retaining a capability for being cast at good production rates.

The use of both good service performance and castability as factors in good design may seem routine. In actual practice, however, it frequently is not. This unfortunate situation arises out of the usual separation of the engineers responsible for the function of the part from those charged with its fabrication. A further influence in this direction is the fact that in many materials there is but limited opportunity for matching best design for serviceability with best design for castability.

WHAT CONSTITUTES GOOD "STRUCTURAL" DESIGN AND HOW DO YOU APPROACH IT?

Good structural design is often described in terms which convey the following:

a) The material is chosen whose mechanical properties are adequate to sustain the critical loads taking into account their mode, direction and rate of application under the expected environmental conditions. It is desirable also that the material be selected to match the design with respect to the presence of stress raisers or notches. In this regard, a structure with unavoidable ribs, unavoidably sharp radii and abrupt changes of section should be matched with a material or alloy or heat treatment that gives the plasticity to permit as much load redistribution as possible.

b) The design is one which comes as close as possible to allowing all portions of the structure equally to share in supporting or distributing load.

c) The structure has no excess metal, but was designed from the low side toward the condition of adequacy, including prescribed safety factors.

Of great assistance to designers in the light metals is the relatively extensive supply of mechanical property data available. The accumulation of such data during and since World War II for both magnesium and aluminum alloys permits certain approaches to designing in the light metals which are not possible for the other structural materials.

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When light alloy castings are to be used, the designer frequently may start with a completely new design, as is usually the case for aircraft and missile components; or start on the basis of a previous design in another metal for which the light alloy casting is to be substituted. The stepwise procedure followed for either approach is outlined in the Table.

NEW DESIGN

These procedures preclude the use of ordinary cut-and-try methods since such methods offer little chance for arriving in a short time at a design of as high structural efficiency as by the procedure outlined in Table and described hereafter. Cut-and-try methods are being modified as the demands for higher structural efficiency become more intense, and as metal strength data for use in predicting service and the facilities for experimental and theoretical stress and thermal analysis are employed.[4,5,6]

Structural uses for which the designer must develop a new design frequently involve complex design requirements where loading and other factors defy predictability. In these cases, a prototype casting is made based on matching theoretical loading requirements against the known safe static, fatigue, impact and creep strength property values given for the metal.

At this point, two problems must be considered. The first is the method for accurately predicting service loading requirements. The accuracy of theoretical design analysis for predicting service conditions in castings where complicated loading is involved is inadequate. Such calculations usually must be predicted on simple static loading and do not properly consider complex dynamic loading, fatigue and impact problems and stress concentrations that may be critical regardless of the thickness of wall sections, or other design details in the immediate proximity of the stress raisers. Theoretical calculations, consequently, serve only as a starting point in the design project.

The second problem involves predicting serviceability from the available mechanical property data. The only data that can be used are those which can be interpreted, to the proper degree, in terms of the foundry and other practical variables that are likely to be present.

These variables include the effects of porosity, grain size, perfection of heat treatment, surface smoothness (as cast surfaces vs. machined surfaces) and stress concentration. If data are available and experience has been obtained to demonstrate how the data can be used directly for design purposes, the first design can then be established.

TABLE — BASIC STEPS IN DEVELOPING A NEW OR SUBSTITUTE DESIGN

I. Developing A New Design—Sequence of Steps for Establishing Final Design	
a. Theoretical Determination Of Loads	
b. Matching Above Loads Against Known Metal Strength Information To Establish Initial Design	
c. Creation Of Prototype Casting	
d. Static And Dynamic Experimental Stress Analysis During Service Or Simulated Service Loading	
e. Matching Results Of Stress Analysis Against Known Metal Strength Information To Establish Final Design	
f. Creation Of Casting To Final Design	
g. Final Proof Testing	
II. Developing A Substitution Design—Sequence of Steps for Establishing Final Design	
a. Assumption That Design In Former Metal Is Adequate — Or Revisions To Allow For Structural Inadequacies Or Changes In Fabrication Procedure	
b. Matching Sections Of Previous Metal With Proposed Light Metal Designs, Taking Into Account Known Metal Strength Information For *Both* Previous Metal and Proposed Light Metal To Establish Initial Design.	
c. Same As c Above	These steps may eventually be excluded when experience with the procedure in (a) and (b) is perfected and when strength data in (b) become adequate for forecasting serviceability for all structural metals.
d. Same As d Above	
e. Same As e Above	
f. Same As f Above	
g. Same As g Above	

SUBSTITUTION DESIGN

It is often advisable, when changing materials, to develop a new design from scratch, thereby permitting freedom for attacking the problem as a whole. In some cases, however, when materials are being changed to magnesium or aluminum, it is imperative that the starting point for designing be the design in the metal previously employed. If it is assumed that the period of service of the part in the previous metal resulted in the evolution of a successful design in the previous material, the problem is then simplified for the designer. The usefulness of complete strength data again becomes apparent, however.

In this case, the designer may substitute magnesium or aluminum, taking into consideration the respective strength properties of the previous and the proposed new metal. Essential for this procedure is complete strength information for the previously used material. This, however, is frequently not as complete as the information available describing the light metals, and, therefore, poses another problem in simple substitution techniques.

If it is known that the part to be converted to light metal alloys is not adequately designed, alterations can be planned at this time based on the previous service information on the part. Frequently certain features of the previous design were necessary to enable given fabrication methods to be employed. Such would be the case where pressed or formed steel parts are involved.

When a casting is designed to replace a formed and assembled part, sections should be changed to accommodate the casting process. Certain section thicknesses and design details were incorporated in the previous design only to facilitate forming and should be identified and altered to favor the casting process and structural efficiency requirements.

After the initial design is established, based on theoretical load calculations or material substitution criteria involving accurate material strength data, an initial casting is made. The efficiency of this first design is then evaluated by experimental stress checks. Metal strength data, if available and sufficiently extensive, are again used for comparison with stresses measured in the experimental stress analysis.

On the basis of this comparison, superfluous metal can be removed where working stresses are judged

to be far below the critical strength values for the metal, or metal added where stresses are above or too close to the critical strength values. A new, and perhaps final, casting design is then made.

WHAT CONSTITUTES GOOD "CASTING" DESIGN AND HOW DO YOU APPROACH IT?

Good casting design is that which gives the foundryman the best chance to achieve uniformly high properties throughout the structure of the casting, while still maintaining the rates of productivity that his type of operation requires.

A study of factors regarding wall configurations is fundamental to good casting design. For castability reasons, the minimum wall thickness desired in magnesium castings should not be under $\frac{3}{16}$-in. For small areas, however, $\frac{5}{32}$-in. or even $\frac{1}{8}$-in. can be maintained. Standard casting design, as it has developed over the years with the heavier metals, has tended toward the use of as thin sections as castability will allow, rigidized and strengthened by the addition of ribbing systems.

So long as the fluidity requirements are satisfied, thin sections do yield mechanical properties superior to thick sections or variable sections.[7] Such thin structures, however, do show lower property values at the intersections of ribs and in sections heavied-up for added strength. These thick-and-thin features impose the need for provisions to feed and otherwise specially to favor the heavier sections, in order to prevent and to control porosity and the lower properties that develop in adjoining areas.

Extensive studies started during World War II and which have been continued to date measure the effect on mechanical properties of foundry-induced variables of this type. The mechanical properties of the uniform thin areas approach, more closely than do variable sections, the properties of cast test bars, and, thus, show better promise to the design engineer by giving average strength properties higher than the specified 75 per cent of test bar minimums.[8] Indeed, the disparity in mechanical properties that exists between test bars and coupons cut from castings demonstrates how much improvement can result if more uniform cast structures as in test bars could be accomplished in castings.

The best structure from a casting standpoint thus can be visualized to be one that is not so thin in the overall wall areas as to require ribbing and rigidizing, but rather one that is somewhat thicker and uniform, and not requiring such irregularities as ribbing systems. Recent studies show that such uniform structures in magnesium yield better properties than the thick-and-thin type configurations which result from thin webs and thick intersections between ribs and highly filleted wall areas.[9]

WHERE DOES MAGNESIUM FIT IN?

Mechanical property data reveal to design engineers why magnesium is of interest. Magnesium may be used in many applications on an equal volume basis, with a full weight saving in proportion to relative densities.

If a small amount of magnesium is added to the general wall thickness of the previous heavy metal design, the ribbing can be eliminated. The magnesium structure resulting is only a few per cent heavier than the ultimate weight saving offered by magnesium on equal volume basis substitutions. Thus, the rigidity and strength of the increased thickness, which increase by the cube and square, respectively, become quite adequate for most applications without the complications of rigidizing ribs and other complicated stress raisers. In structural analysis studies, many designers have come to the conclusion that ribs do not materially strengthen castings, proof of which is provided in both practical and theoretical treatments. Instead of strengthening the structure, such ribbing systems provide stress concentration zones, poor quality zones and crack propagation centers, especially in dynamically loaded parts, that more than offset the reasons for introducing them in the first place. The literature has become quite well documented with experimental stress analysis and theoretical proofs of the fact.[9,10]

Most magnesium sand castings are sufficiently massive after the above treatment that the lower modulus of elasticity of magnesium becomes a much reduced problem. The avoidance of stress concentrations by the adding of wall thickness rather than ribs is in accordance with the recommended practices for generous filleting. The use of a large number of fasteners modestly loaded rather than a few fasteners heavily loaded, the use of longer studs or through bolts, the avoidance of sharp corners particularly around bosses and the use of alloys of magnesium in the heat treatments that provide maximum toughness rather than only highest strength and maximum hardness are in accordance with this practice.

The lower modulus of elasticity of magnesium alloys may contribute to better serviceability in some cases since it effects an improved distribution of load to the principal load carrying members, thus contributing to the efficient use of metal in the structure.

One of magnesium's most favorable features is its high dimensional stability at elevated temperatures providing recommended heat treatments are applied.

The coefficient of thermal expansion of magnesium is somewhat higher than for the heavier structural metals. This should be taken into account to avoid the development of high stresses which would lead to creep and distortion.

SPECIFIC DESIGN APPLICATIONS FOR MAGNESIUM CASTINGS

An original aluminum crankcase for a 12 cylinder aircraft engine was of conventional design with heavily ribbed sections, as shown by the section in the upper left view of Fig. 1. The part did not pass qualifying tests and was overweight. The part was changed to magnesium. The ribs were removed from the sidewalls and from the bearing diaphragms, as shown in the section in the lower right view of this Fig. 1.

Stress concentrations were thus reduced and wall thicknesses were increased slightly, so that the section

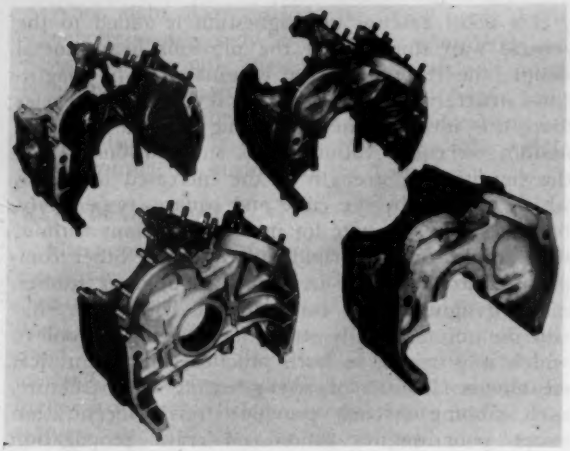


Fig. 1 — Evolution of a clean design in a crankcase; bearing support section is shown. Initial design is upper left. Sequence follows to upper right, lower left to final at lower right. Ref. (9).

modulus at highly stressed sections was increased and the lower modulus of elasticity was offset. The resulting magnesium crankcase did pass the qualifying tests with more than twice the endurance of the initial ribbed aluminum structure, but with a 25 per cent weight saving.

The procedure above is shown as it progressed through the stages of design simplification; upper left to upper right, to lower left hand to final structure lower right hand. Note the absence of ribs in the clean final design.

An aircraft camshaft housing, Fig. 2, was converted from aluminum to magnesium and the design simplified in similar fashion. The old version in various areas is at the left. The changed structure is at the right of the figure. The removal of ribs and the slight addition of wall thickness reduced stress levels 55 per cent with no increase in weight or flexibility.

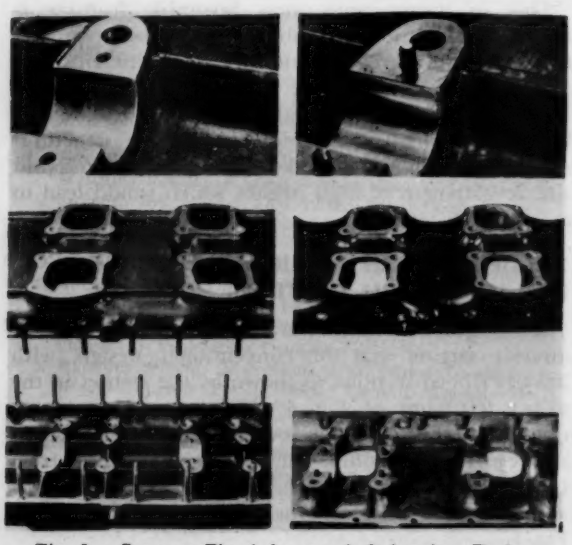


Fig. 2 — Same as Fig. 1 for camshaft housing. Design features before simplification are at left, and after is at right. Ref. (9).

Similar simplifications in aircraft wheel design have taken place. The intricate, highly ribbed wheel structures, Fig. 3, are giving way to the simpler design, Fig. 4, with reduction in cost and increases in service life under critical fatigue loading. These structures have become sufficiently clean that some are now being either permanent mold cast or forged, rather than sand cast.

A final series of views show two other areas of magnesium casting application of interest to aircraft and missile designers. Figures 5 and 6 show a sand cast magnesium missile wing, an application warranted by the development of strength and rigidity in bending by slight increases in wall thickness, but without weight penalty when compared to the heavier structural materials.

Figures 7, 8, 9 and 10 show various die cast and sand cast magnesium electronic chassis components. These are good magnesium applications in today's



Fig. 3 — Aircraft wheel of previous ribbed design.

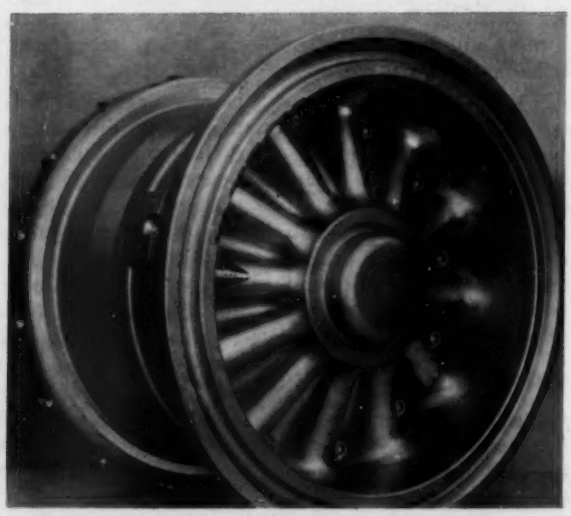


Fig. 4 — Aircraft wheel of simplified ribless design.

highly instrumented missile and aircraft applications, because of the high degree of rigidity and structural stability that can be developed without prohibitive weight penalties.

CONCLUSIONS

A critical view of the structural design and casting design possibilities for magnesium reveals what can be termed a break-through in design opportunities. Alloys of cast magnesium are being designed in configurations to approach not only the foundryman's ideal in design, but the designer's ideal as well.

The designer's ideal, as already described, is one where stress concentrating features can be avoided without the penalties of excessive weight from built-up wall thicknesses, and where simple functional design without re-entrant sections in complicated load distributing patterns are eliminated.

For the foundryman, the ideal design is one of uniform, relatively thin sections where feeding prob-

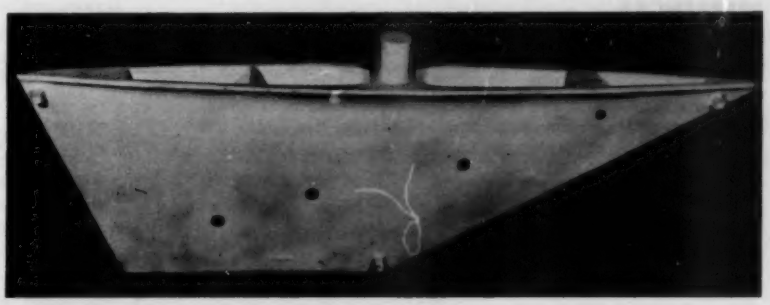


Fig. 5 — Sand cast magnesium missile wing before finishing.

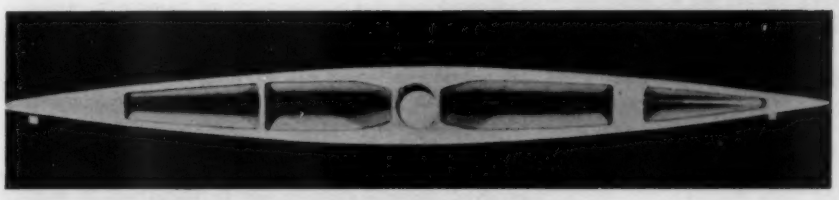
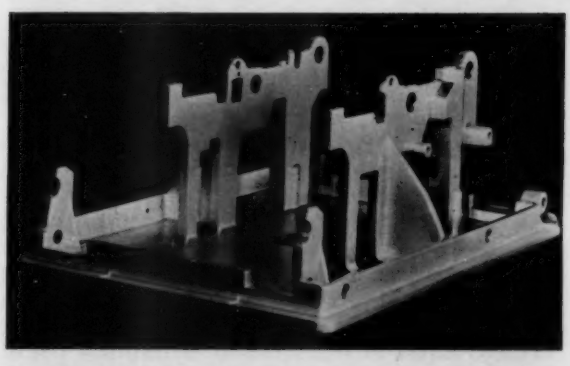
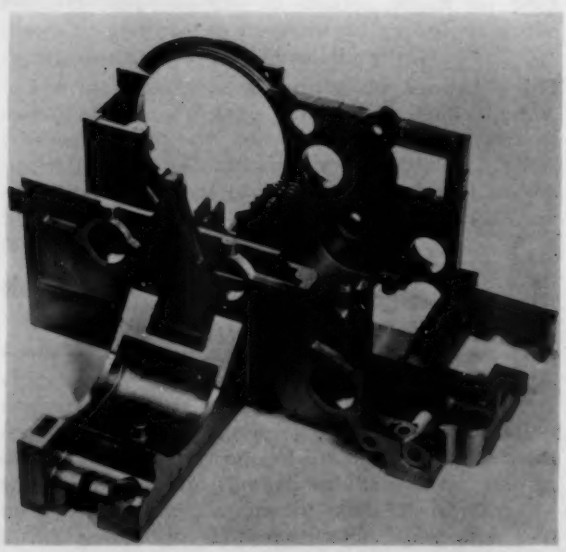
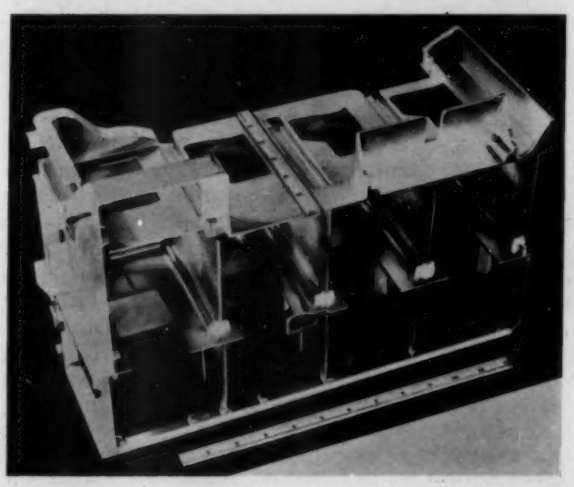


Fig. 6 — Same as Fig. 5. End view.







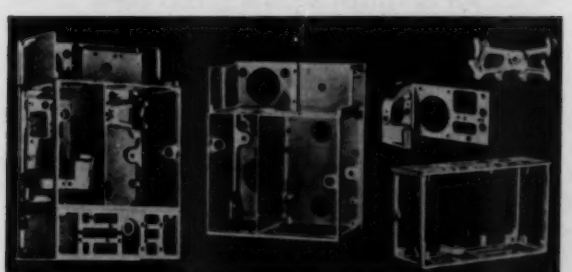


Fig. 7, 8, 9, 10 — Various cast magnesium chassis components for electronics units.

lems are simple, walls are without abrupt change in thickness or directionality and metal flow and solidification can be controlled without the complications of elaborate feeding and risering systems.

The so-called simple, clean "forging-like" design is the result of these procedures.

ACKNOWLEDGMENT

All figures and tabular matter of this paper are courtesy of The Dow Chemical Co., Midland, Mich.

REFERENCES

1. John C. McDonald, "Tensile, Creep and Fatigue Properties at Elevated Temperatures of Some Magnesium-Base Alloys," A.S.T.M., *vol. 48,* (1948).
2. George H. Found, "The Notch Sensitivity in Fatigue Loading Of Some Magnesium-Base And Aluminum-Base Alloys," A.S.T.M., *vol. 46,* (1946).
3. "Room and Elevated Temperature Properties Of Magnesium Casting Alloys," Bulletin No. 141-176, The Dow Chemical Co., Midland, Mich.
4. George H. Found, "Fatigue Characteristics of Magnesium Castings" A.S.T.M., (1946).
5. George H. Found, "The Relative Significance of Design and Metallurgical Factors To The Serviceability of Magnesium Castings," AFS TRANSACTIONS, *vol. 54,* p. 93, (1946).
6. George H. Found, "Efficient Magnesium Castings – Their Design and Production," *Metal Progress,* p. 833, Dec., 1954.
7. M. C. Flemmings, R. W. Strachan, E. J. Poirier, H. F. Taylor, "Performance of Chills on High Strength Magnesium Alloy Sand Castings of Various Section Thickness," AFS TRANSACTIONS, *vol. 66,* p. 336, 1958.
8. Military Specification QQ-M-56.
9. W. T. Bean Jr., "Simplification of Light Metal Casting Design and Its Effect Upon Serviceability," AFS TRANSACTIONS, *vol. 55,* p. 430, (1947).
10. Anonymous, "Minimum Ribbing For Maximum Strength," *Die Casting,* Feb., 1949.

FACTORS INFLUENCING SOUNDNESS OF GRAY IRON CASTINGS

a review of some recent British work

By I. C. H. Hughes, K. E. L. Nicholas, A. G. Fuller, and T. J. Szajda

ABSTRACT

The constitution of nearly all gray irons leads to both contraction and expansion stages during their solidification, and soundness is to a large extent dependent upon the detailed way in which these volume changes act. In many castings, general expansion of the casting during solidification is associated with the formation of shrinkage defects. Casting expansion is promoted by mold cavity volume changes, which are more pronounced in softer molds and are dependent upon pouring temperature and molding sand type and properties.

Major composition of the iron also affects casting expansion and soundness by influencing the eutectic graphite amount. Phosphorus leads to soundness problems because of the presence of a low melting point phase. An important finding is that increasing the degree of nucleation of the graphite eutectic leads to unsoundness. This is a common foundry variable which can be measured, and is amenable to some control.

NATURE OF GRAY IRON SHRINKAGE

Shrinkage defects in light and medium sectioned gray iron castings are comparatively rare, unless the composition is strongly hypoeutectic or contains more than about 0.2 per cent phosphorus. Many British castings, particularly those of light section, contain over 1 per cent phosphorus, and considerable technical skill is necessary to make sound the wide range of castings produced from such irons. Early research work on soundness was, not unnaturally, largely confined to investigations of major compositional variables, in particular carbon, silicon and phosphorus.[1]

Such investigations clearly demonstrated the important influence of composition on soundness, but were limited in their practical application. This is because it was apparent that within a given foundry melting an iron of fairly constant composition under repetitive conditions, there were other undetermined variables operating. These were causing greater variations in the incidence of shrinkage defects than could be explained by the most careful analysis of the effect of the minor changes in composition.

Research over the past decade has led to the belief that apart from liquid shrinkage prior to solidification, the actual solidification of the commonly used gray irons does not involve a net shrinkage, except in irons of hypoeutectic composition or in noneutectic high phosphorus irons. However, the solidification sequences of gray irons involve two or three stages in which either shrinkage or expansion are occurring. It is the way in which the associated volume changes act which leads to the occurrence of shrinkage defects.

These volume changes result either in the reduction of shrinkage cavities, or in an overall expansion of the casting creating further unsoundness. This fact is highlighted by the spectacular increase in the incidence of unsoundness in nodular graphite irons, compared with flake graphite irons of the same major composition, in which the same quantities of graphite and austenite form during solidification.

The precipitation of graphite is accompanied by a relatively large increase in volume, which has been shown to give rise to proportional expansion of the casting during this stage of solidification.[2]

The main variables governing the formation and magnitude of shrinkage defects are mold rigidity, pouring temperature, constitution and the degree of nucleation of the austenite-graphite eutectic. These variables have been extensively investigated at the BCIRA, and some of the more important results are reviewed in this paper.

HOW GRAY CAST IRONS SOLIDIFY

In order to understand the fundamental types of shrinkage unsoundness characteristic of the main kinds of gray irons, a knowledge of their solidification sequences is necessary. This subject has received considerable attention.[3,4,5] The solidification sequence of an iron controls the distribution of solid and liquid eutectic throughout the casting, this in turn controlling the nature and distribution of shrinkage defects. The magnitude, time and uniformity of the casting dilation relative to the net solidification volume changes govern the severity of the shrinkage defect.

In hypoeutectic irons the initial process is the precipitation of austenite dendrites, a stage of solidification involving a shrinkage. Because of the presence of eutectic liquid dispersed throughout the casting, it is unlikely that any internal cavity formation can occur at this stage of solidification. Shrinkage is therefore accommodated either by top surface or hot spot piping or, alternatively, by the supply of feed metal through the gates or from an external feeder.

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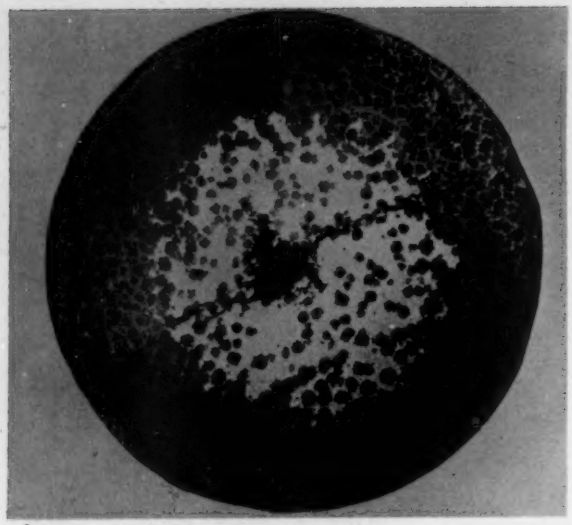


Fig. 1 — Macrostructure of quenched gray iron bar, showing region of solidified eutectic at edge of casting and simultaneous dispersed solidification of eutectic cells throughout the interior.

Subsequently the binary graphite/austenite eutectic solidifies with an expansion which will tend to force the last remnants of liquid out of the top of the casting to appear as plugs in previously formed pipes, or out of hot spots as exudation defects. Alternatively, a general increase in the size of the casting may occur.

The graphite/austenite eutectic solidifies throughout the casting in discreet cells. Within each cell the graphite is interconnected, as has been demonstrated by deep etching techniques. The cells are roughly spherical, and solidification proceeds fastest at the edge of the casting to form a continuously thickening shell of solid eutectic, although at the same time cells are growing throughout the interior of the casting. This type of solidification, which does not represent true directionality, is illustrated in Fig. 1 (from the work of Gittus[6]). Figure 1 shows the macrostructure of a gray iron casting quenched during eutectic solidification.

The casting skin and the semi-solid casting interior are sufficiently plastic to permit deformation of the casting due to either mold wall movements or expansion accompanying the precipitation of graphite.

Figure 2 represents schematically an intermediate stage in the solidification of a low phosphorus hypoeutectic flake graphite iron, and is based upon the results of quenching experiments and thermal analysis. In irons of this type it is rare to find internal porosity, but piping may occur at hot spots and junctions between sections of differing thickness.

Solidification sequences of hypereutectic irons differ from those of hypoeutectic irons only in the nature of the primary phase.[5] Precipitation of primary graphite occurs over a range of temperatures, and is accompanied by liquid shrinkage. This liquid shrinkage almost certainly more than compensates for the expansion accompanying the primary graphite precipitation.

Increasing the phosphorus content of a flake graphite iron above 0.2 per cent considerably changes the shrinkage characteristics of the iron, and may lead to the formation of either interdentritic porosity, or alternatively localized shrinkage cavities. An explanation for this is given by a study of the solidification sequence of a 1.1 per cent phosphorus flake graphite iron.

In irons of this type flake graphite, eutectic solidification is substantially completed while the eutectic cells are still surrounded by a network of phosphide eutectic liquid, as shown in Fig. 3. This liquid is thought to form a continuous network even at quite low phosphorus contents, although only isolated particles appear in sections under the microscope. By deeply etching nodular irons of various phosphorus contents, it was possible to remove the matrix completely and to shake out the graphite nodules, revealing the phosphide network which surrounded the eutectic cells. A representation of the resulting structure, in a 0.4 per cent phosphorus iron, as seen under a binocular microscope, is shown in Fig. 4.

The melting point of the phosphide eutectic is about 950 C (1742 F), some 200 C (392 F) lower than that of the graphite eutectic. Hence, a stage of solidification is reached when the casting is solid except for a network of phosphorus-rich liquid.[4] This liquid remains throughout the casting in spite of quite steep temperature gradients which may exist in the latter stages of solidification in some castings. Finally, the phosphide eutectic solidifies progressively throughout the casting.

The solidification sequence of nodular iron differs from that of flake graphite iron mainly in two respects. The solidification of the austenite/graphite eutectic is more dispersed (Fig. 5), the last liquid being widely dispersed in fine interconnected channels throughout the casting. Unlike the last liquid in phosphoric irons, it consists of ordinary liquid graph-

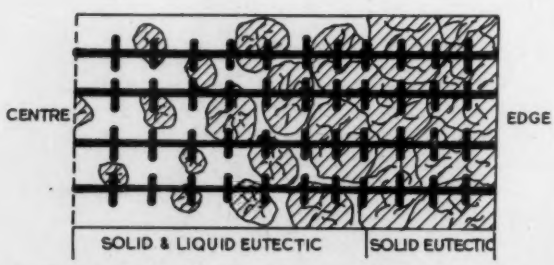


Fig. 2 — Low phosphorus hypoeutectic flake graphite iron, intermediate stage of solidification.

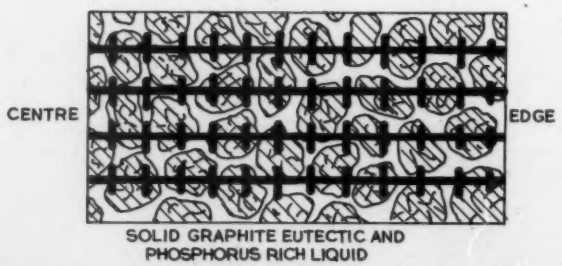


Fig. 3 — Hypoeutectic flake graphite iron (1.1 per cent phosphorus), late stage of solidification.

ite eutectic at a temperature close to its solidification point. Consequently, while liquid remains it can flow and be replaced with ordinary feed metal of near-eutectic composition.

In addition, the eutectic cells in a nodular graphite iron are smaller and far more numerous than in a flake graphite iron, and contain only one graphite particle in comparison with the numerous interconnected flakes within each eutectic cell of a flake graphite iron.

HOW SHRINKAGE DEFECTS OCCUR

Castings liable to contain unsoundness are sounder when poured in rigid molds than similar castings of the same composition and pouring temperature cast in green sand molds.[7] This appears to be true for all types of gray iron, irrespective of the probability that most common compositions solidify with a net expansion. Furthermore, although low phosphorus flake graphite irons are generally sound, nodular graphite irons of the same major composition frequently require feeding or contain appreciable internal porosity.

An explanation of this phenomenon is offered by the work of Gittus,[8] who showed that the expansion tendency of the casting during solidification of the eutectic is much greater in nodular irons than in flake graphite irons. The comparison is shown in Fig. 6.

Clearly an interpretation of the behavior of these two irons suggests that the expansion accompanying eutectic solidification can either cause self-feeding of the casting with exudation of liquid eutectic through a previously formed pipe defect, or alternatively a general expansion of the casting. These two alternatives are represented diagrammatically in Fig. 7. The mechanism (c) is suggested as being the most probable in low phosphorus flake graphite irons, although even in irons of this type some expansion occurs unless rigid molds are employed.

The mechanism (d) represents the behavior of nodular graphite irons in green sand molds, and will also explain the formation of internal porosity in this type of iron. This has indicated a method whereby the soundness and yield of nodular graphite irons can be increased, namely, by the use of a rigid mold to restrict casting expansion. Using rigid molds, casting expansion in nodular irons has been eliminated and purging of liquid eutectic observed (Fig. 18) together with a soundness improvement.

In low phosphorus flake graphite irons the expansion due to the solidification of the last liquid is normally sufficient completely to fill the enlarged casting, that is, the casting is self-feeding. However, increasing the phosphorus content is accompanied by a progressive increase in internal porosity. This may be a consequence of the following factors.

1) Additions of phosphorus progressively reduce the proportion of the total eutectic which appears as the binary austentite-graphite constituent. Quenching experiments have shown that in a particular 0.4 per cent phosphorus iron, 95 per cent of the total eutectic is made up of the austentite-graphite phases, whereas in a 1.1 per cent phosphorus iron of the same carbon equivalent, the proportion of the austentite-graphite phases has been decreased to 70 per cent of the total eutectic. This effect of phosphorus is equivalent to decreasing the amount of eutectic graphite.
2) The presence of a network of phosphorus-rich liquid throughout the casting.



Fig. 4 — Sketch of phosphide eutectic network in a 0.4 per cent phosphorus nodular graphite iron.

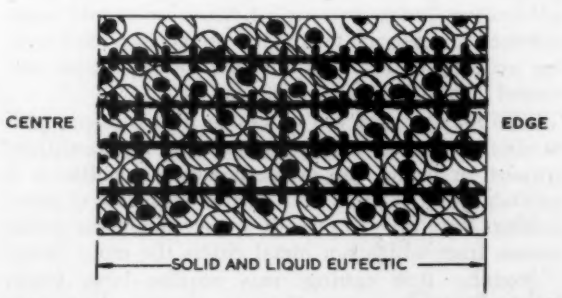


Fig. 5 — Hypoeutectic nodular graphite iron, late stage of solidification.

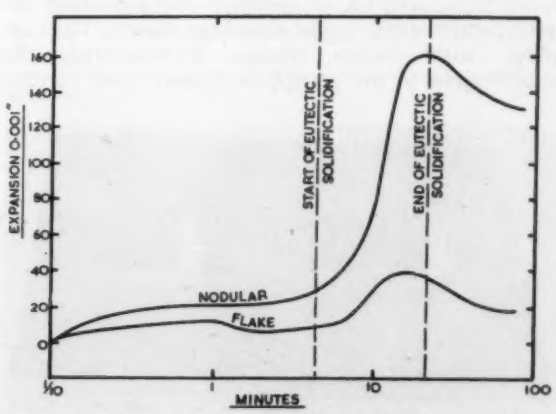


Fig. 6 — Expansion curves for flake and nodular graphite iron castings poured in green sand molds.

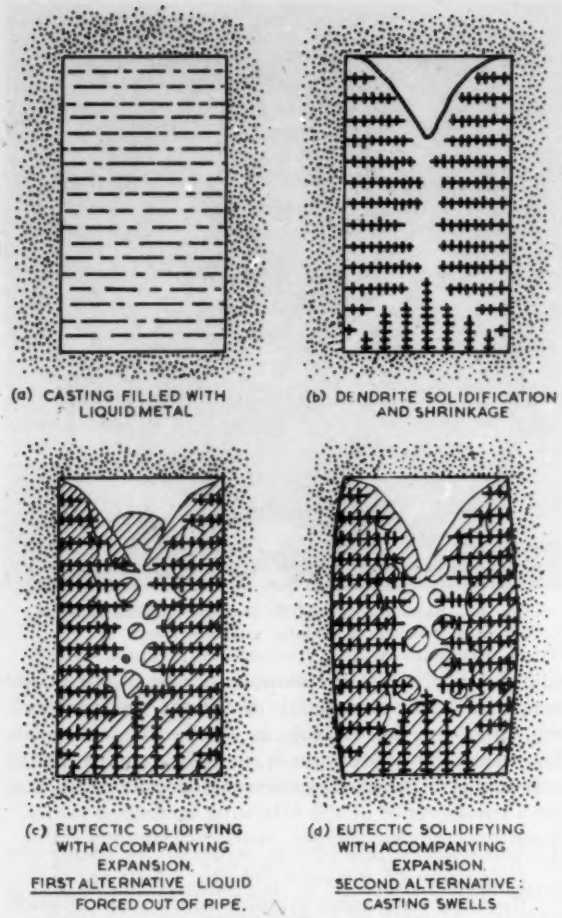


Fig. 7 — Solidification of gray iron castings.

3) Overall expansion of castings during eutectic solidification in softer types of molds may be greater than the expansion due to eutectic graphite. The colder first parts of the casting to solidify therefore draw last eutectic liquid from the hotter parts of the casting. When the last liquid is phosphorus-rich, it will drain from the hot spots by this mechanism to leave a characteristic network of porosity.

During cooling of the casting to the phosphide eutectic temperature, liquid shrinkage occurs. This, together with volume changes accompanying the solidification of the phosphide eutectic, will further promote the internal porosity typical of phosphoric gray irons.



Fig. 8 — Phosphide exudation (sweat) on a light casting.

The fact that the phosphide eutectic liquid remains in a network throughout a casting when solidification is substantially complete is demonstrated by the fact that under certain conditions this liquid may be exuded at numerous positions on the casting skin, forming the familiar phosphide sweat shown in Fig. 8. In spite of the presence of this network, however, normal feeding through these channels is impracticable, since liquid iron from feeders is unlikely to penetrate for any distance to colder regions of the casting which are substantially below the temperature of solidification of the graphite/austenite eutectic.

FEEDING

Detailed studies of a wide range of castings, varying from a few ounces to nearly 150 tons in weight, have indicated that feeding is necessary to compensate for:

1) Liquid shrinkage, the extent of which is governed by the casting temperature and the proportion of which is approximately constant for all common types of gray iron.[9]
2) Mold cavity and casting expansion, which in a wide range of castings has been shown to correspond closely to the quantity of feed metal supplied. This is illustrated by Fig. 19.

The evidence obtained supports the view that the solidification volume changes of normal gray cast irons do not involve an overall contraction and, therefore, make no demands on feed metal. Liquid shrinkage, which occurs prior to solidification, may be calculated, but mold cavity expansion is dependent upon casting design and mold type. Hence, no explicit rule is possible as to the total amount of feed metal necessary.

For any feeder to contain a supply of liquid metal until the remainder of the casting is solid it would have to be unrealistic in size, especially for larger castings. For this reason it is often more expedient and economical to use on large castings small feeders having insulated sleeves which can be frequently refilled with hot metal.

In general, the principles of directional solidification so commonly applied to steels and white irons cannot be properly achieved in gray irons, because of the dispersed nature of eutectic solidification.

In small, thin section gray iron castings, the effect of mold cavity expansion upon soundness is comparatively small and liquid shrinkage is generally compensated for by feeding through the gates; thereafter self-feeding during eutectic solidification should make the need for external feed metal unnecessary. Self-feeding utilizing eutectic graphite expansion is most successful if solidification is uniform.

This may be achieved by design modifications such as eliminating sharp changes in section and re-entrant corners, or by the use of local pads and chills. It is probable that shrink "bobs" fitted to sprays of small castings serve only to provide a reservoir in the gating system from which hot metal enters the mold cavity.

Nodular iron castings may require large feeder heads for two reasons. First, to compensate for the greater mold cavity expansion occurring during eu-

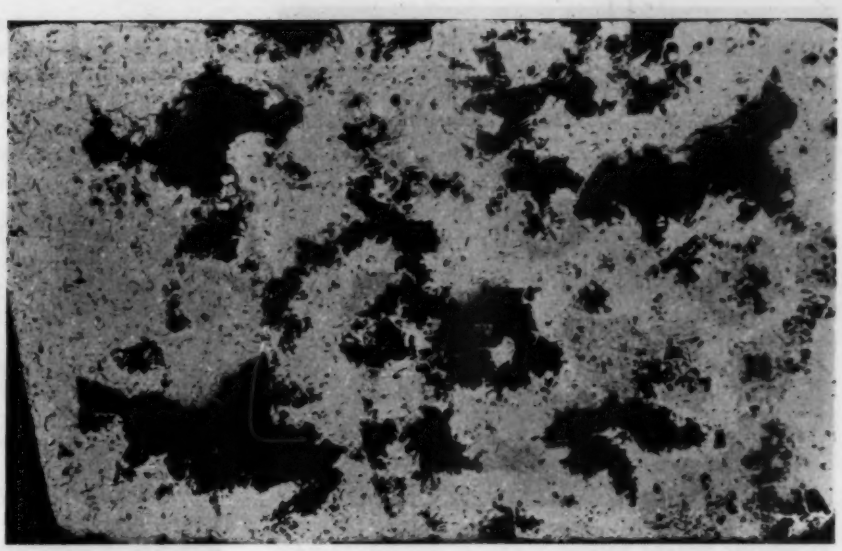


Fig. 9 — Gross porosity in a phosphoric iron casting resulting from mass feeding. 5 ×.

tectic solidification, and second, to promote directional solidification. In an unfed nodular iron casting, the last liquid to solidify, and associated internal porosity arising from casting expansion, is widely dispersed, but unlike phosphoric flake graphite irons normal feeding is practicable.

As has been described previously, shrinkage associated with the solidification of the phosphide eutectic in phosphoric irons is virtually impossible to feed by external means.

If a pronounced hot spot exists in a phosphoric iron casting there may be a flow of phosphide and semi-solid graphite eutectic away from the hot spot. This settling of the semi-solid mass will result in the formation of a gross shrinkage cavity, and is often referred to as mass feeding. This type of defect, illustrated in Fig. 9, can be remedied either by supplying liquid feed metal direct or by removal of the heat center from the casting.

It must always be remembered that a feeder head introduces a hot spot. This is undesirable if the hot spot extends into the casting, so that often it is better to avoid using feeders for flake graphite irons where the necessity for feeding is marginal and to rely upon the self-feeding behavior of the casting.

Green sand, which is the common molding material employed in the light and medium castings industry, suffers from the disadvantage that unless the molds are firmly rammed the castings may exhibit swellings or be oversize. This fact is well known, but it is not so widely recognized that oversized castings tend to be associated with shrinkage defects. The problem is illustrated in Fig. 10, which shows a direct relationship between the incidence of internal porosity and the extent of casting expansion in brake-drum castings.[7]

MOLD CAVITY AND CASTING EXPANSION ARE FUNDAMENTALS

An improvement in casting soundness can be obtained by the use of dry or core sand molds which are more rigid than green sand molds, but such a measure is generally uneconomical except for special or large castings. Common examples of the use of dry sand or loam molds are provided by the manufacture of large engineering castings and ingot molds. Such castings often require substantial post-casting additions of feed metal if they are to be free from shrinkage cavities.

The early feed-metal additions compensate for liquid shrinkage, but it has been clearly demonstrated that large and prolonged additions of feed metal are necessary solely to compensate for expansion of the casting in the apparently rigid mold. Foundry in-

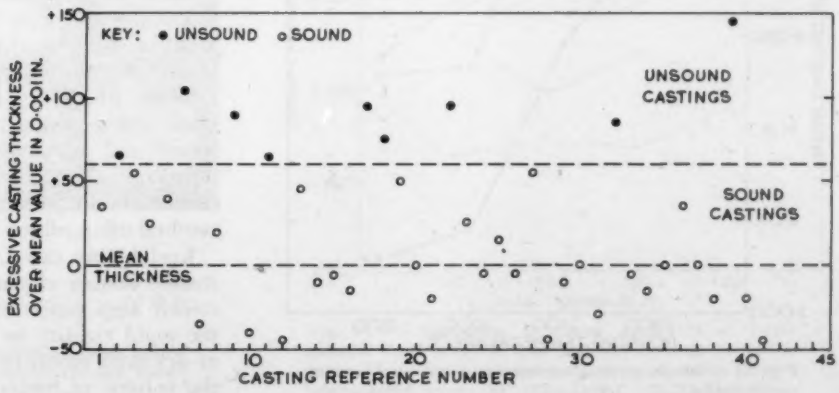


Fig. 10 — Expansion of green sand molded brake drum castings related to internal porosity.

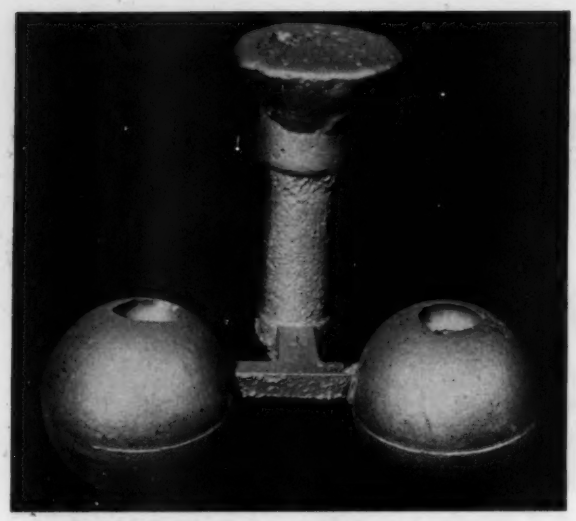


Fig. 11 — Three in. diameter test casting.

vestigations have shown that, for even the largest ingot molds, the increase in casting wall thickness compared with the width of the original mold cavity corresponds closely with the feed metal requirements.

Because of the relatively long freezing range of gray irons, especially those containing phosphorus, the casting is likely to be sufficiently plastic to follow any variation in mold cavity volume during most of the solidification process.[6] Any factor which would alter the volume of the mold cavity during solidification of the casting will be reflected in the ultimate soundness or feed metal requirements of the casting, as is clearly demonstrated by the experimental work described below.

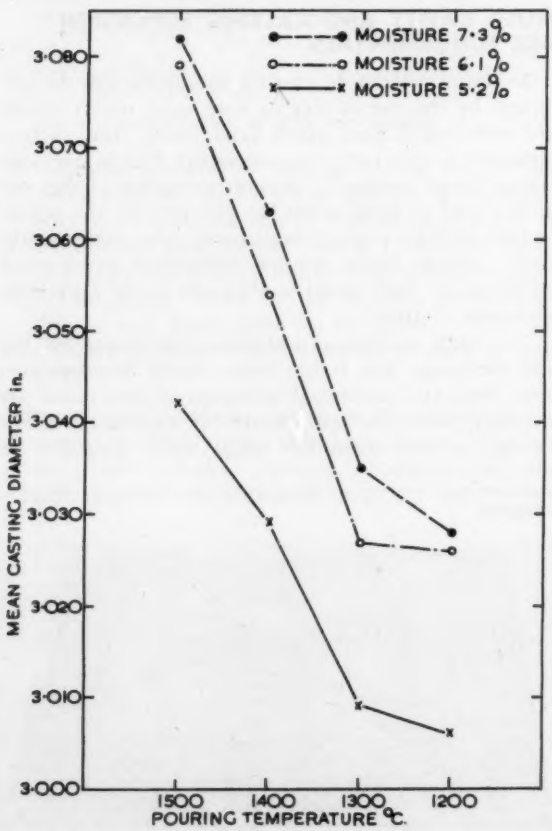


Fig. 12 — Mean casting diameter (in.) vs. pouring temperature.

Investigations to study expansion of unfed gray iron castings have been carried out using 3 in. diameter spheres,[10] as shown in Fig. 11. The variables studied included moisture and coal dust contents of green molding sands. Casting expansion was calculated from direct measurement of the diameter of the sphere in three positions mutually at right angles. The results of these experiments have shown that increasing the moisture content progressively increased the casting expansion (Fig. 12), whereas an addition of 6 per cent superfine coal dust (at constant moisture content) decreased the dilation.

These results are similar to those reported by Gittus,[2] Womochel and Sigerfoos[11] and Wallace and Evans,[12] and clearly indicated that these factors operate by affecting the deformation characteristics of the mold.

Use of the same test piece has shown that the soundness of gray iron castings is significantly improved by the use of materials yielding a more rigid mold. The materials studied included dry sand, CO_2 process and cement sand, and the improvement in soundness arising from the use of these molding materials is shown by sections through the castings, illustrated in Fig. 13. Measurements of casting diameter showed that expansion was greatest in green sand and least in cement molds (Fig. 14).

All these results indicate that the extent of casting expansion should be directly related to the volume of shrinkage in the casting. Such a relationship should be expected on fundamental grounds. Many of the experiments were performed using a gray flake graphite iron of constant composition poured at constand temperatures within the range 1500 C (2732 F) to 1200 C (2192 F). If the relationship between the mean casting diameter and volume of shrinkage defect is examined for a given pouring temperature, there is a significant correlation between these two factors, the volume of the shrinkage defect increasing as the casting expansion increased.

The relationships for pouring temperatures of 1400 C (2552 F) and 1300 C (2372 F) are shown in Fig. 15, and provide evidence that shrinkage is in fact directly related to casting expansion.

Nodular graphite irons present a more difficult problem. Experiments carried out on the soundness of nodular graphite irons have employed keel block castings, of the type shown in Fig. 16. This keel block was so designed that the keel section would be sound and adequately fed. In this casting all the shrinkage is concentrated in the feeder head, and the effects of the variables studied was shown in the residual unsoundness in the head.

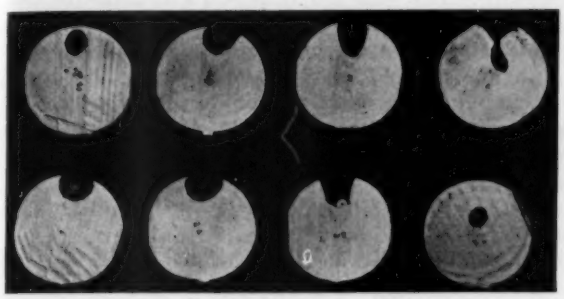
Keel blocks cast in green sand molds showed extensive casting expansion, and the feeder heads contained deep pipes and dispersed porosity. Increasing the mold rigidity by using hard rammed green sand or dry sand molds reduced the casting expansion and the volume of feeder head pipe. This suggested that

a relationship might exist between the amount of feed metal required as measured by the feeder head pipe volume and the casting expansion.

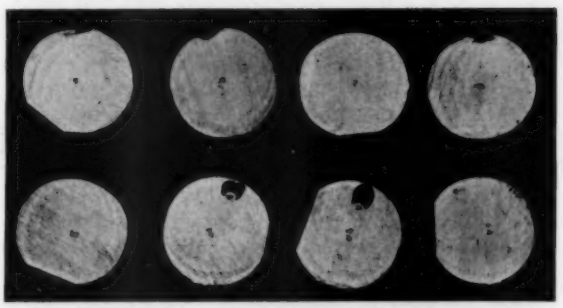
Experiments using a hypoeutectic nodular graphite iron of constant composition were extended to include a rigid cement sand mold, and comparable measurements of the casting expansion with feeder head pipe volume were made in a series of green and cement sand molds. Casting expansion was measured by comparing mold and casting dimensions at corresponding positions on the casting. Green sand molded castings showed a large net expansion relative to the original mold dimensions, while cement molded keel blocks contracted.

The mean changes in casting dimension relative to the mold, feeder head pipe volume, and casting weight are given in the table for the keel blocks cast in either green sand or cement sand molds.

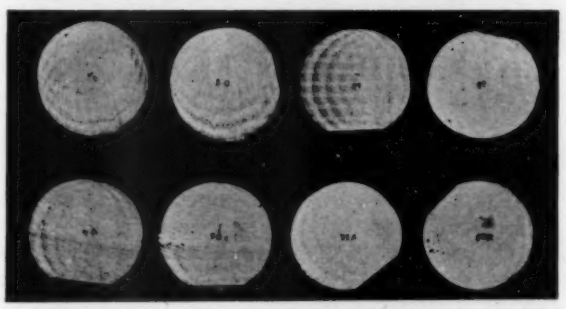
Sections through green sand and cement sand molded keel blocks of the same metal composition and pouring temperature are shown in Figs. 17 and 18, illustrating the expansion of the keel section in a green sand mold associated with gross shrinkage in the feeder head, and the improved dimensional accuracy and soundness obtained from a rigid cement sand mold. If the mean casting expansion or contraction relative to the original mold dimension for each of the keel blocks is plotted against volume of feeder head pipe, a linear relationship is obtained. This result is shown in Fig. 19 and confirms that



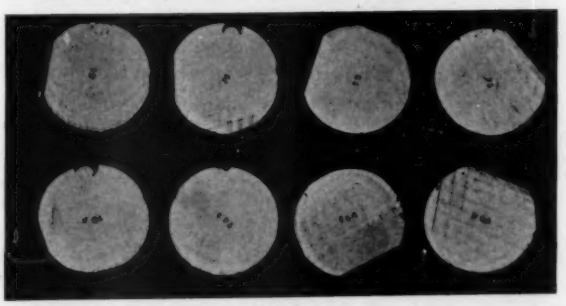
Cast in green sand molds



Cast in dry sand molds



Cast in CO_2 process molds



Cast in cement sand molds

1500 C (2732 F)	1400 C (2552 F)	1300 C (2372 F)	1200 C (2192 F)

Fig. 13 — Sections through 3 in. spherical flake graphite iron test castings. Pouring temperatures are shown under each column.

TABLE — CASTING DIMENSION MEAN CHANGES

Type of Mold	Mean Change in Casting Dimension Relative to Mold, in.	External Feeder Pipe Volume, cc	Casting Weight, lb
Green Sand	+0.095	99.3	45.2
Cement Sand	−0.051	16.9	43.8

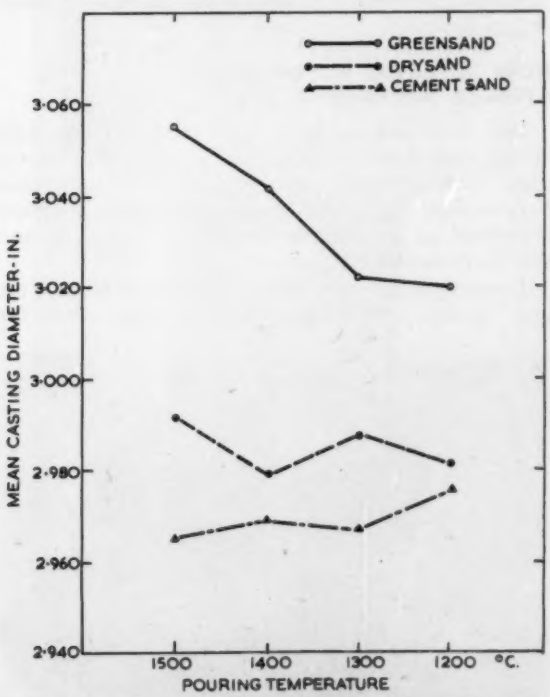


Fig. 14 — Comparison of mean diameters of hypoeutectic flake graphite spherical test castings poured in green sand, dry sand and cement sand molds over a temperature range of 1500-1200 C (2732-2192 F).

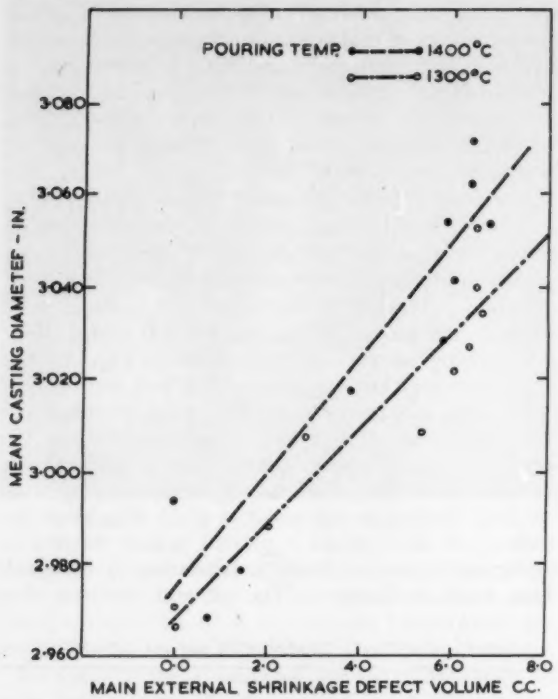


Fig. 15 — Relationship between mean casting diameter and volume of external shrinkage defect volume for 3 in. hypoeutectic flake graphite iron spheres cast in molds of different rigidity at 1400 C (2552 F) and 1300 C (2372 F).

for hypoeutectic nodular irons of fixed composition and pouring temperature, there is a strong positive correlation between casting expansion and feed metal requirements.

HOW EUTECTIC EXPANSION AFFECTS SOUNDNESS

The soundness of gray irons is intimately connected with the precipitation of graphite during eutectic solidification. This provides the expansion force which, under the correct conditions promotes soundness, or alternatively expands the casting and may result in unsoundness.

The proportion of eutectic which is present in the iron reaches a maximum value of unity when the carbon equivalent content, given by the formula C.E. % = Total Carbon % + ⅓ (Silicon % + Phosphorus %), reaches a value of 4.3 per cent.

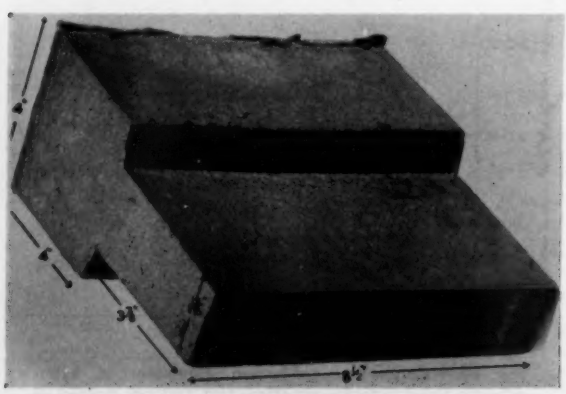


Fig. 16 — Design of keel block casting, showing principal dimensions.

The value given by this equation is only an approximation and the true value of the eutectic, over irons of a wide range of compositions, is in some doubt.[13] However, the proportion of graphite in the eutectic varies with the relative carbon, silicon and phosphorus contents, and should also reach a maximum value to provide maximum expansion.

In low phosphorus hypoeutectic irons, the eutectic graphite may be calculated from a knowledge of the iron-carbon-silicon system.[14] An alternative formula, used at the BCIRA, has been reported by Fuller:[15]

$$\text{Eutectic Graphite \%} = \frac{100(\text{Total Carbon \%} - 1.7 + 0.1\ \text{Silicon \%})}{(100 - 1.7 + 0.1\ \text{Silicon \%})}$$

In low phosphorus hypereutectic irons allowance must be made for the amount of graphite precipitated prior to the solidification of the eutectic when calculating the amount of eutectic graphite. For phosphoric irons other formulas have been proposed,[16] but lack of knowledge of the quantitative role of phosphorus in modifying the constitution of the system makes an accurate formula unobtainable at present, although variations in eutectic graphite with constant phosphorus content have been followed using formulas similar to that stated above.

In unfed castings of medium section thickness made in green sand, increasing the carbon equivalent content or the eutectic graphite content yields a more open grain, and castings become more prone to porosity as has been shown by Timmins[1] and by Jeffery and others.[17] This is believed to be due to the greater casting expansion with increase in eutectic graphite, as illustrated by Figs. 20 and 21, in which the size and the internal porosity, measured by Archimedes' principle, of 4 in. high by 4 in. diameter cylinders both increased with increase in eutectic graphite. These test castings contained 1 per cent phosphorus so that the comparative, but not the absolute values, of the eutectic graphite are valid.

On the other hand, it has been shown that the effect of increasing carbon equivalent or eutectic graphite is to promote soundness in light castings. It is believed that the reason for these contrary effects is the reduced extent of casting expansion in lighter-sectioned castings. The effect of increasing carbon equivalent in improving soundness in hypoeutectic irons has been shown by Gittus[18] employing K-bar castings, and by Hughes[19] for plates with bosses in which surface shrinkage defects occur beneath the bosses. This latter result is reproduced in Fig. 22, which also shows that the effect is reversed in hypereutectic irons.

EUTECTIC CELLS AFFECT SHRINKAGE

Solidification of the gray iron eutectic is initiated at a number of points or nuclei; subsequent solidification proceeds by growth from these centers.[20] The number of nuclei available for solidification can be varied in a given iron. The solidification process and

structure of the iron are considerably influenced by the degree of nucleation of the melt, and it is therefore not surprising that the shrinkage characteristic of an iron should be closely related to its degree of nucleation.

Each region of eutectic, consisting of graphite and austenite solidifying around a nucleus, is termed a eutectic cell, and by suitable etching of a microspecimen these eutectic cells may be revealed. Since each eutectic cell has grown from one nucleus, the number of eutectic cells gives a measure of the degree of nucleation of the melt.

It is a fact of repeated observation that, increasing the number of eutectic cells leads to an increase in unsoundness. The reason for this effect has not yet been fully established, but preliminary experiments suggest that increasing the number of eutectic cells growing leads to a small increase in casting expansion. This may result from changes in the solidification sequence.

The number of eutectic cells which grow may often be quite independent of the composition, and at constant pouring temperature and cooling rate is mainly governed by the treatment which the melt has received. The degree of nucleation of the melt may vary widely, and is probably more readily recognized by foundrymen as variations in the condition of the molten metal from time to time.

Variations in the degree of nucleation lead also to variations in chilling tendency and in mechanical strength. This difference in numbers of eutectic cells in different irons is sometimes recognized as differences in the grain of a casting when fractured. In polished samples, the eutectic cell number may be enumerated by etching and examination at low magnification, and this constitutes a useful form of control test for soundness.[21,22,24]

As an example, Fig. 23 illustrates the eutectic cell structure of a sound and a defective casting, made during the same day in a cylinder foundry. This figure shows that the unsound casting contains the largest number of eutectic cells.

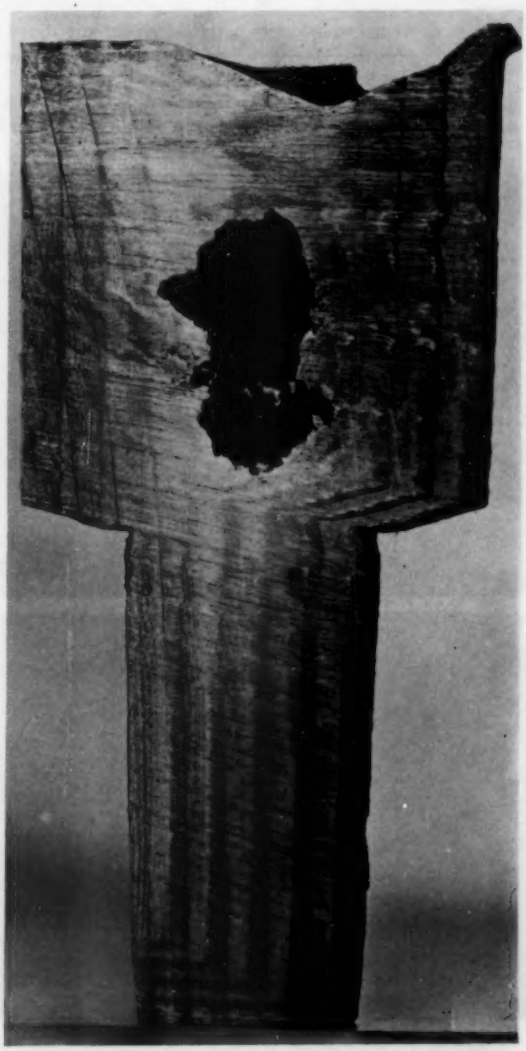


Fig. 17 — Green sand molded nodular graphite iron keel block. Note swelling of keel section and gross cavity in head, in comparison with Fig. 18.

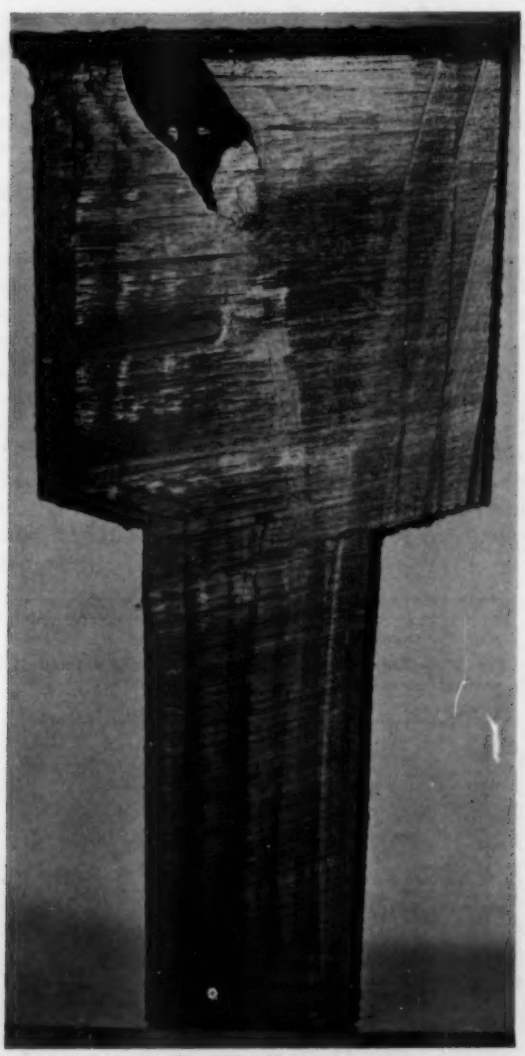


Fig. 18 — Cement molded nodular graphite keel block. Note presence of purged plug in head. Metal composition is similar to casting shown in Fig. 17.

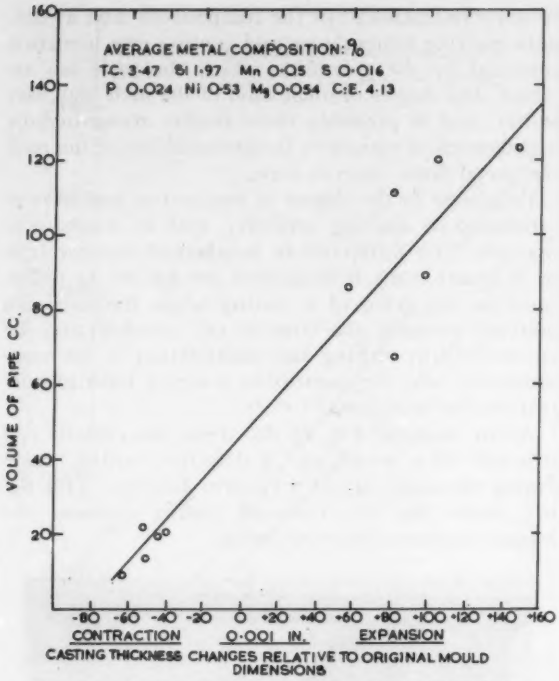


Fig. 19 — Green sand and cement sand molded nodular iron keel block casting expansions and contractions (relative to mold dimensions) vs. volume of pipe in feeder head.

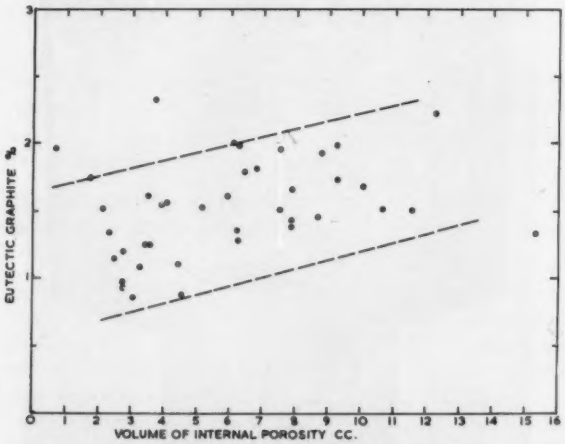


Fig. 21 — Eutectic graphite content and the volume of internal porosity.

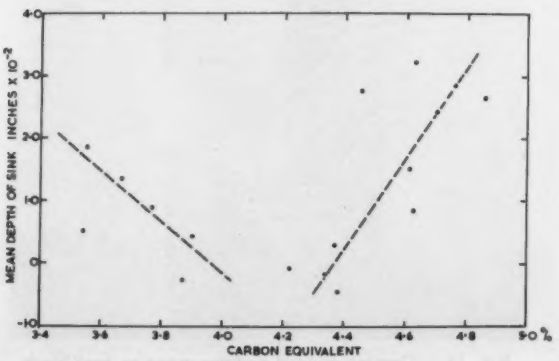


Fig. 22 — Relationship between depth of sink and carbon equivalent content for uninoculated thin plates.

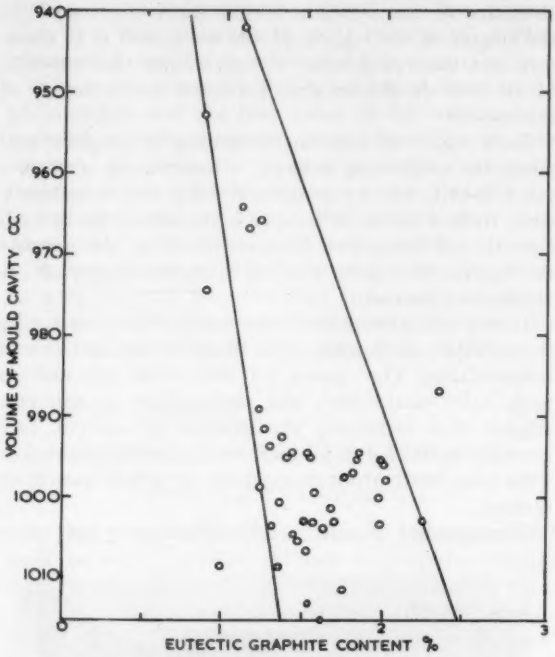


Fig. 20 — Volume of mold cavities and eutectic graphite content.

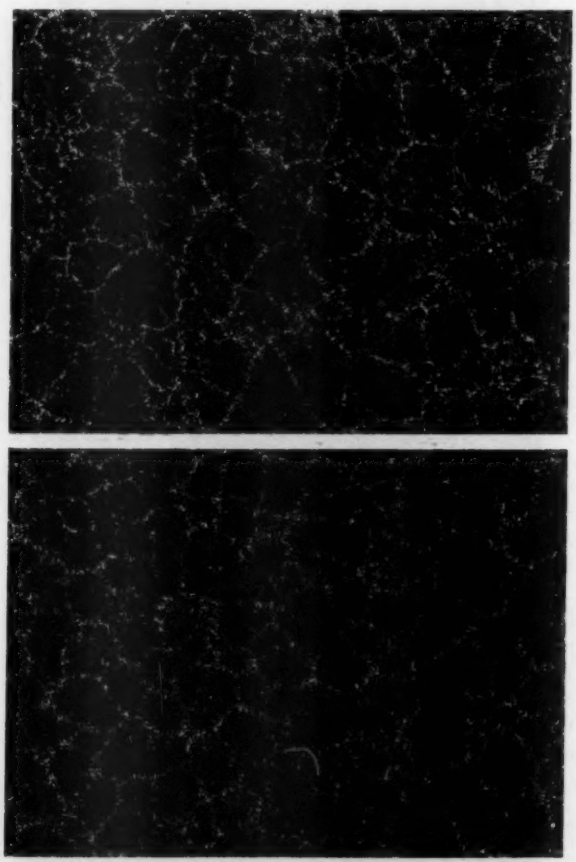


Fig. 23 — Difference in eutectic cell number in sound and defective casting. Sound casting (top) contained 530 cells per sq in.; unsound casting (bottom) contained 890 cells per sq in. Stead's reagent etch. 7 ×.

In an investigation carried out using a test casting which consisted of a plate 9 in. x 6 in. x $\frac{5}{16}$-in. thick on which were mounted two bosses each $1\frac{1}{4}$ in. diameter and 1 in. high (Fig. 24), it has been shown that there is a significant correlation between the number of eutectic cells growing in the iron and surface shrinkage (sinking) which occurs on the face of the casting at the points where the bosses are attached to the reversed side. Sinking increased as the number of eutectic cells increased (Fig. 25).

These results were obtained from experiments using batch furnaces in which the effect of variations in the manganese and sulfur contents of the iron, ladle inoculation with ferrosilicon, and melting history of iron upon sinking were determined using a basic metal composition of:

Total Carbon, %	3.1 – 3.3
Silicon, %	2.7 – 3.0
Manganese, %	0.7 – 0.8
Sulfur, %	0.07 – 0.09
Phosphorus, %	1.05 – 1.15

The composition of this iron is similiar to that employed in many British foundries producing light domestic-type castings.

VARIATIONS EFFECT IN THE IRON MELTING HISTORY

Superheating the melt and holding metal in the furnace are processes which reduce the degree of nucleation of the melt, and hence the number of eutectic cells which grow at a given pouring temperature.[23] A series of experiments[21] was carried out in which the surface shrinkage characteristics of an iron were examined at constant composition and constant pouring temperature. The iron had been superheated to temperatures within the range 1450 C - 1600 C (2642 F-2912 F), and held in the furnace for periods between zero and 30 min. Surface shrinkage decreased with increased furnace temperature, and increased furnace holding time (Fig. 26).

As a result of these observations, it may be predicted that the established preference for melting as hot as possible in cupola operation is consistent with minimizing shrinkage defects, although this effect may in part be offset by increased contact with carbonaceous material when the higher melting temperature is obtained by increasing the coke charge. Holding the metal in a receiver or bull ladle at the spout of a cupola should tend to minimize shrinkage, since during this holding period the effect of inoculation will gradually fade and the degree of nucleation will be decreased.[24] In the case of a batch-type furnace, melting at a slow rate or holding for long periods are likely to promote soundness.



Fig. 24 — Design of plate with bosses test casting.

INOCULATION

Inoculation may be defined as; an addition to the molten iron of a relatively small amount of a substance having a powerful effect in increasing the degree of eutectic nucleation, and hence increasing the number of eutectic cells which grow. Such substances include graphite, coke and ferrosilicon. All iron melted in contact with carbonaceous material such as coke, as for instance in cupola melting, will to some extent be inoculated. The actual degree of inoculation depending upon the melting conditions, and the time of standing out of contact with inoculating materials during which period the inoculation effect wears off.

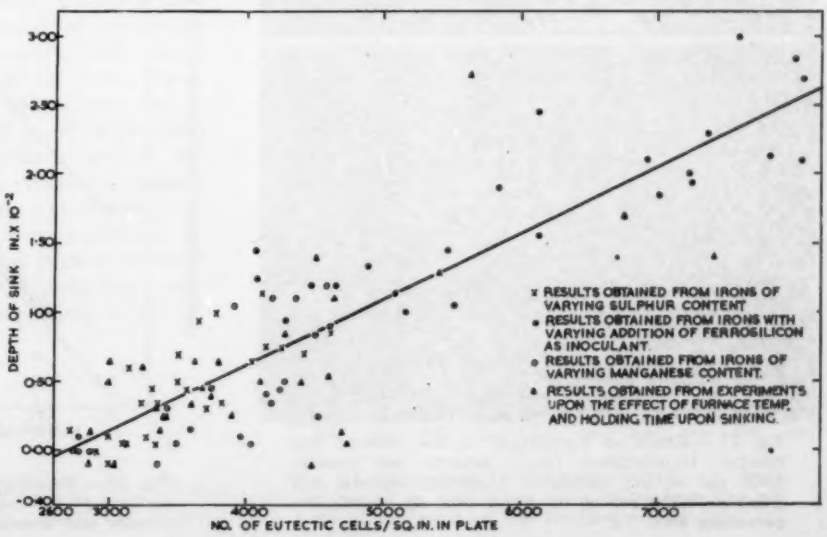


Fig. 25 — Relationship between depth of sink and combined effects of sulfur and manganese contents, ferrosilicon inoculant and furnace temperature and holding time on eutectic cell number.

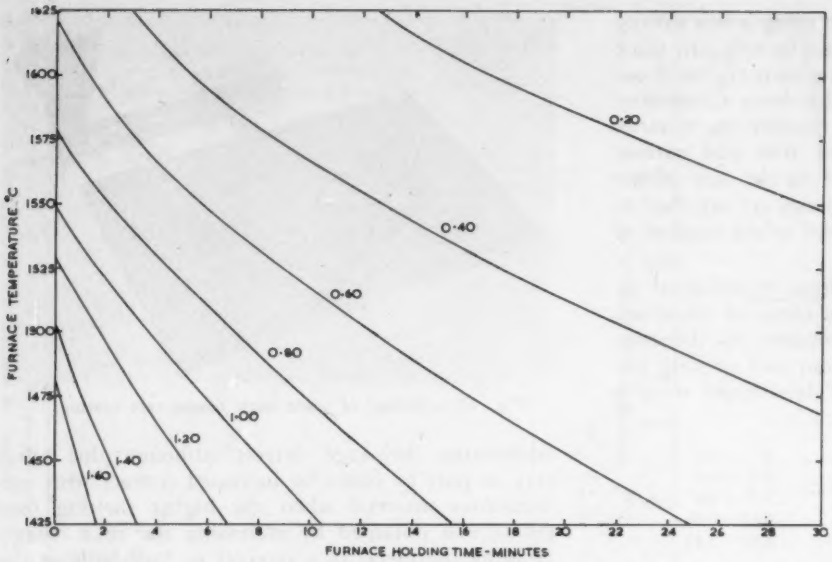


Fig. 26 — Relationship between furnace temperature, furnace holding time and observed depth of sink.

Similarly, metal which has been melted out of contact with materials known to have a graphitizing effect will be relatively uninoculated. This phenomenon has been clearly demonstrated in the laboratory, examples of uninoculated and inoculated irons etched to show their eutectic cell distribution being shown in Fig. 27. A quantitative relationship between the ferrosilicon increment, as an inoculant, the eutectic cell number and soundness has been demonstrated, and is illustrated in Fig. 28.

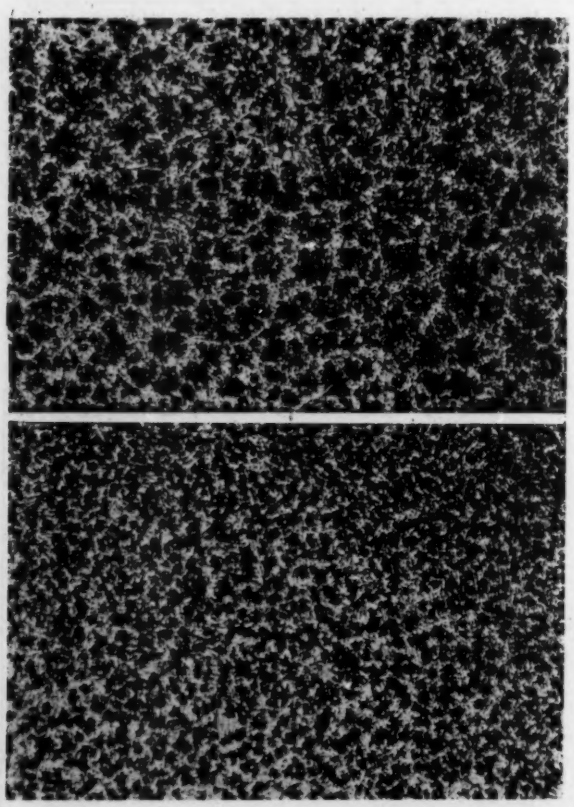


Fig. 27 — Effect of inoculation on the eutectic cell number. Uninoculated (top), eutectic cell number 3,900 per sq in.; inoculated (bottom), eutectic cell number 10,200 per sq in. 5 per cent aq. ammonium persulfate etch. 12 ×.

In British practice, deliberate inoculation is not generally employed except for the production of high duty irons of low carbon equivalent value. Such irons possess relatively high shrinkage, and an increase in shrinkage due to inoculation will probably not be important since feeding would normally be practiced.

Some processes such as recarburization, desulfur-

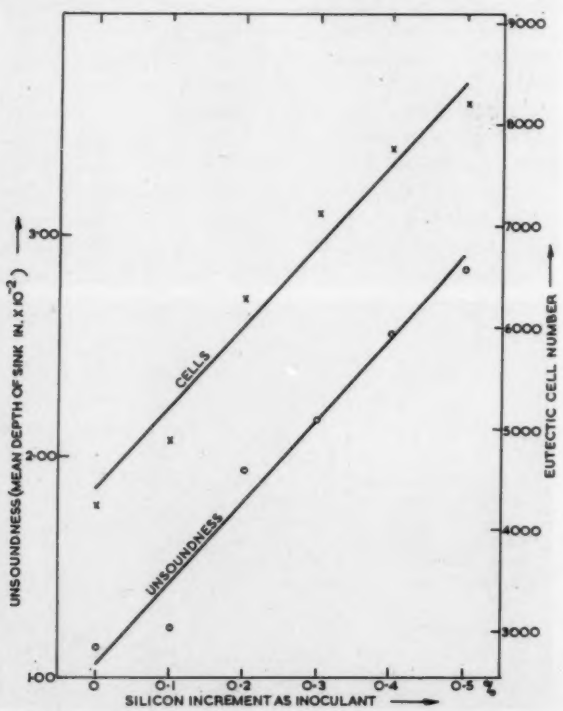


Fig. 28 — Relationship between silicon increment (ferrosilicon 80 per cent silicon) as inoculant, eutectic cell number and unsoundness.

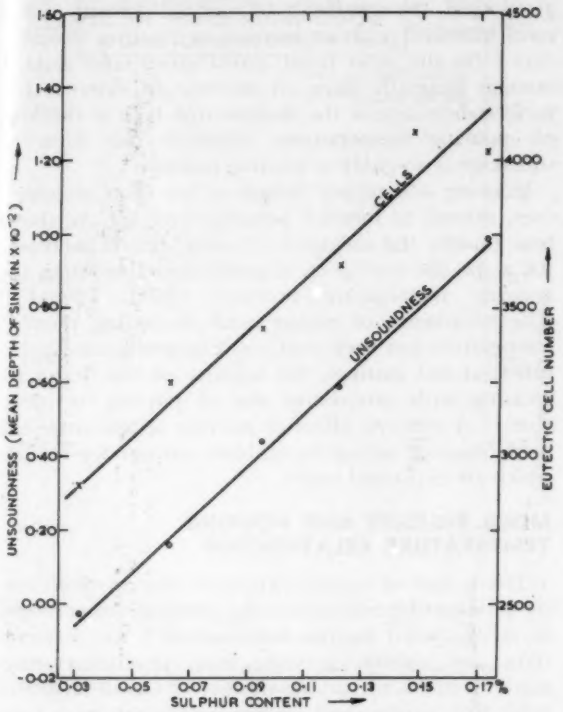


Fig. 29 — Relationship between sulfur content, eutectic cell number and unsoundness.

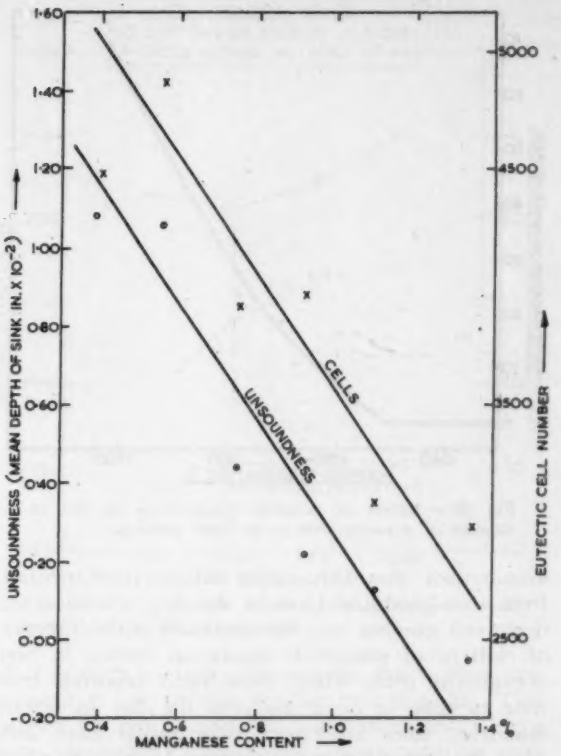


Fig. 30 — Relationship between manganese content, eutectic cell number and unsoundness.

ization and resiliconization, which involve the injection of materials into the iron, will increase considerably the degree of nucleation and will promote unsoundness. On the contrary, an addition of titanium followed by bubbling either carbon dioxide or argon through the metal in the ladle, which has been shown by Timmins[1] to promote soundness, is also explicable in terms of the effect upon the number of eutectic cells growing. This latter process considerably reduces the degree of nucleation of the melt.

SULFUR AND MANGANESE EFFECT

Increasing the sulfur content of liquid iron causes an increase in shrinkage unsoundness. This is accompanied by an increase in the degree of nucleation measured by the number of eutectic cells, as is shown in Fig. 29. Manganese, on the other hand, tends to decrease the number of eutectic cells and to increase soundness, as shown by Fig. 30. This effect has been established in the laboratory and in the foundry, where increasing the manganese content has been shown to reduce sinking and drawing defects in light castings.[8] Although it is difficult to distinguish cause and effect, it is at present believed that manganese has its effect by the progressive neutralization of sulfur in solution according to the reversible reaction:

$$Mn + FeS \rightleftharpoons MnS + Fe$$

EUTECTIC NUCLEATION ROLE IN THE FOUNDRY

Variations in the degree of eutectic nucleation (state of inoculation) are probably one of the major factors affecting shrinkage characteristics within the foundry and, until recently, certainly the least suspected. As a result of laboratory experiments, it was believed that variations in the condition of the metal due to variations in cupola operation, and delays in casting metal during the course of the day, might account for otherwise inexplicable variations in the shrinkage characteristics of an iron of approximately constant composition produced in foundries.

Variations in the degree of nucleation of iron produced from a hot-blast cupola during the course of a day[24] are shown in Fig. 31. The decreased eutectic cell number associated with metal, obtained from a continously tapped cupola shortly after the tap hole was opened, supports the opinion often expressed by

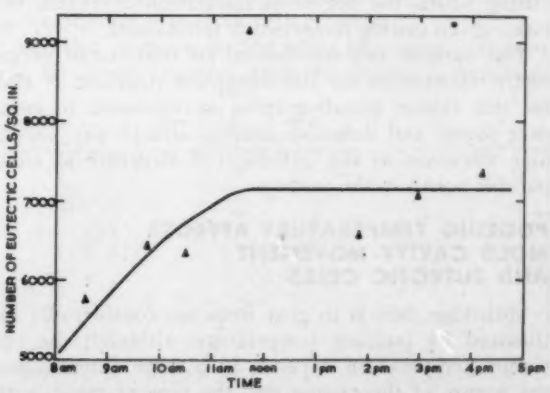


Fig. 31 — Variations in the mean eutectic cell number per sq in. with time, using metal from the bull ladle at the spout of the hot-blast cupola.

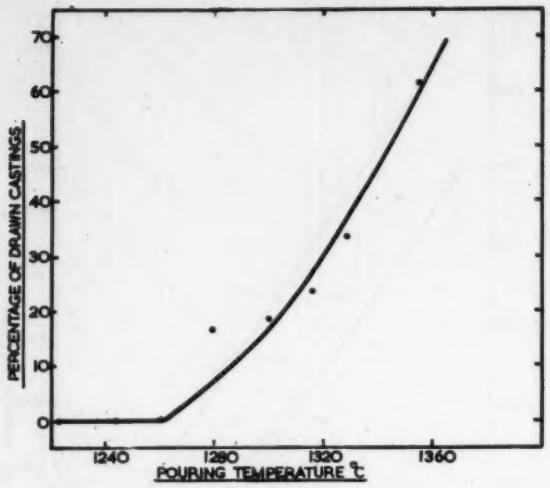


Fig. 32 — Effect of pouring temperature on the incidence of drawing defects in light castings.

foundrymen that this metal differs fundamentally from that produced later in the day. This low eutectic cell number may be associated with outbreaks of chill or of phosphide exudation defects in high phosphorus irons which have been reported from time to time to occur early in the day in certain foundries, since laboratory experiments have indicated that these defects only occur in irons in which the number of eutectic cells growing is low. Throughout the course of the day there was a strong correlation between the degree of nucleation of the iron and the amount of shrinkage which occurred.

Variations in the state of inoculation in the foundry may arise from a wide variety of causes, some of which may involve variations in practice such as cooling some ladles of metal by adding lumps of pig iron. This treatment would produce an iron of a higher state of inoculation, and a greater shrinkage tendency, than one which was allowed to cool in the ladle normally. Similar variations in the temperatures of the iron tapped from two cupolas may affect soundness, not only due to variations in the melting conditions within the cupolas, but also because iron produced in the cupola which was running hotter would require a greater standing time, during which the degree of nucleation decreases, before a given casting temperature is obtained.

The eutectic cell number in an iron is not necessarily reflected in the flake graphite structure,[24] and for this reason metallographic examination to compare sound and defective castings should pay particular attention to the eutectic cell structure at comparable points in the castings.

POURING TEMPERATURE AFFECTS MOLD CAVITY MOVEMENT AND EUTECTIC CELLS

Shrinkage defects in gray irons are considerably influenced by pouring temperature, although the optimum temperature depends to a large extent upon the nature of the casting and the type of mold, with the consequence that early authorities were not always in agreement. In K-bar test castings molded in green sand, the occurrence of internal porosity at hot spots increased with an increase in pouring temperature.[8] On the other hand, unfed green sand molded castings generally show an increase in external top surface shrinkage of the shallow sink type at the lowest pouring temperatures, although this form of shrinkage is no guide to internal porosity.

Drawing and piping defects at hot spots are, however, related to internal porosity, and Fig. 32 shows how rapidly the incidence of these defects increased for a specific casting on a production line when the pouring temperature exceeded 1260 C (2300 F).[7] The occurrence of piping with increasing pouring temperature has been confirmed in green sand unfed spherical test castings, the severity of the defect increasing with progressive rise of pouring temperature.[10] A converse effect of pouring temperature has been observed in rigidly molded castings for reasons which are explained below.

MOLD RIGIDITY AND POURING TEMPERATURE RELATIONSHIP

The extent of casting expansion due to mold cavity dilation depends upon the pouring temperature, as is illustrated by the behavior of 3 in. diameter flake and nodular graphite iron spherical castings made in molds of varying rigidity.[10] Green sand, dry sand, CO_2 process and cement sand molds have been employed in experiments in which the pouring temperature was varied over the range 1375 C - 1200 C (2507 F - 2192 F). The effects of pouring temperature upon casting size, weight and shrinkage volume, were observed for each type of mold.

Pronounced trends were obtained for green sand, CO_2 process and cement sand molds, but no consistent trend was observed for dry sand molds. Green sand molded castings increased in mean diameter and weight with increasing pouring temperature, as shown in Fig. 33, whereas castings made in CO_2 process molds behaved conversely, diameter and weight decreasing with increasing pouring temperature, as shown in Fig. 34. Cement sand molds showed the same trend as CO_2 process molds.

Gittus' work[2] provides an explanation of the influence of pouring temperature upon casting size, at least in green sand molds. If a green sand mold cavity increases in volume when filled with molten metal as a result of the heating of the sand, it is probable that this expansion is time dependent—the hotter the metal entering the mold, the greater will be this expansion of the mold cavity before the casting solidifies and is no longer able to follow changes in the mold cavity volume.

Increasing the pouring temperature also delays the process of solidification. Quenching and thermal analysis of 2 in. diameter bars of a hypoeutectic flake graphite iron at various stages of solidification has clearly demonstrated this effect of pouring temperature, as illustrated in Fig. 35.

Rigid mold materials, such as CO_2 process and cement sand, are believed to expand on heating, causing an inward movement of the mold walls and thus reducing the volume of the mold cavity before solidification of the casting is complete.

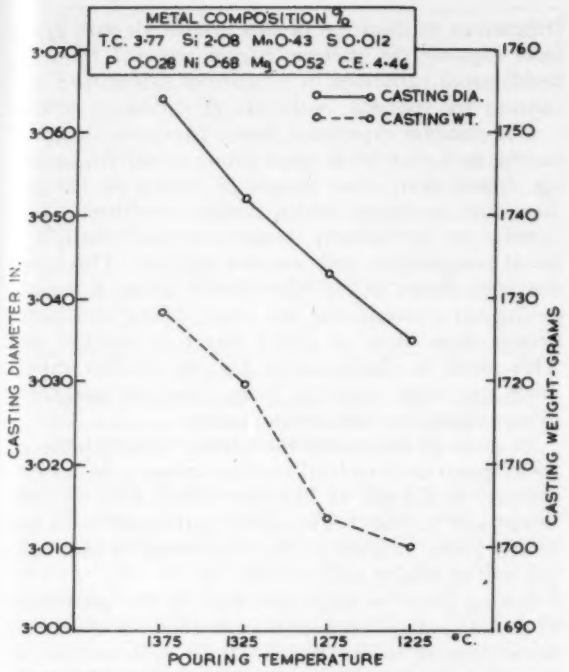


Fig. 33 — Green sand molds: casting diameter and weight vs. pouring temperature.

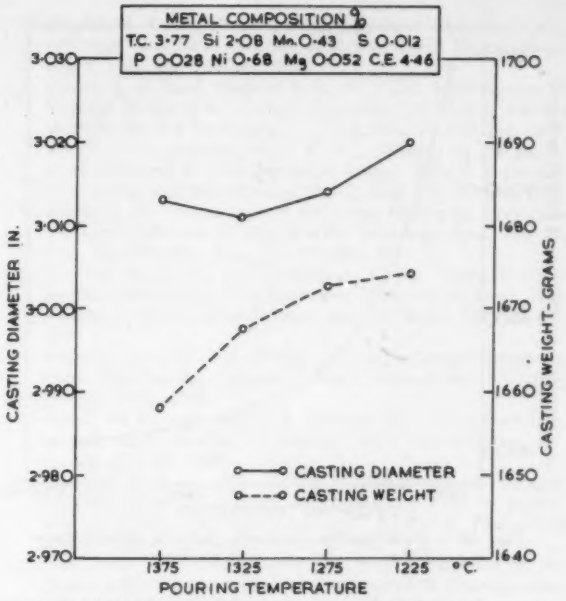


Fig. 34 — CO_2 process molds: casting diameter and weight vs. pouring temperature.

Therefore, the longer complete solidification is delayed by hot pouring, the less will be the casting size and the sounder the casting in these molds. Gittus suggested that in ordinary dry sand the mold cavity decreased when heated by metal.[2] However, further experiments[10] have shown that this effect only becomes pronounced in most rigid molds, and that dry sand molds are usually intermediate in behavior.

Nevertheless, foundry experience has shown[7] that sounder castings are sometimes obtained in dry sand molds when the iron is poured hot. In practice, the correct pouring temperature to minimize shrinkage defects will depend upon the molding procedure employed, and will be lower for green sand molded castings than for castings poured in more rigid molds.

POURING TEMPERATURE AND EUTECTIC CELLS

For an iron of a given degree of nucleation, decreasing the pouring temperature or increasing the cooling rate in the mold will increase the number of eutectic cells growing. This increase is relatively small at high pouring temperatures, but is most pronounced at low pouring temperature (Fig. 36). This increase in cell number may be associated with a sharp increase in unsoundness at low pouring temperatures. This is shown in Fig. 37, in which it is seen that below 1320 C (2408 F) sinking under bosses increased as pouring temperature decreased.

Above 1320 C (2408 F), however, the dominant effect was that of pouring temperature on the mold, and sinking increased slightly with pouring temperature. It would be expected that increasing the pouring temperature would lead to an increase in the inherent tendency of the metal to produce sound castings, due to a reduction in the number of eutectic cells growing. However, in green sand molds, it appears that the contrary influence of increasing pouring temperature in causing mold cavity expansion is generally the overriding feature. In the more rigid molds, however, the effect of pouring temperature upon the mold leads to greater soundness with increased pouring temperature, as described earlier. No doubt this effect is strengthened by the simultaneous decrease in the number of eutectic cells growing.

Whether or not a casting made in a particular iron will be sound depends upon its design as much as

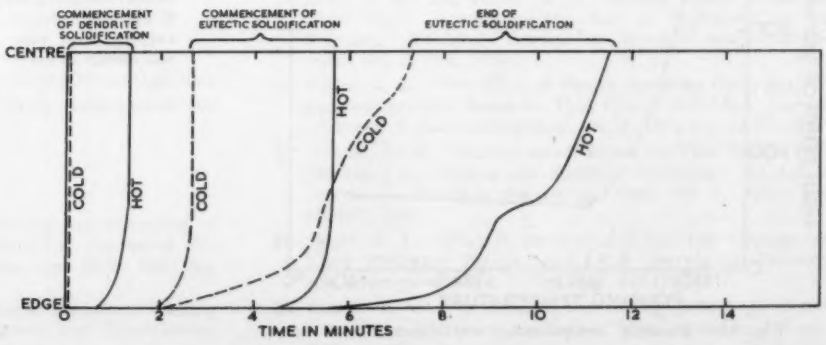


Fig. 35 — Pouring temperature variations effect on the solidification sequence of a 2 in. diameter bar of a low phosphorus hypoeutectic flake graphite iron.

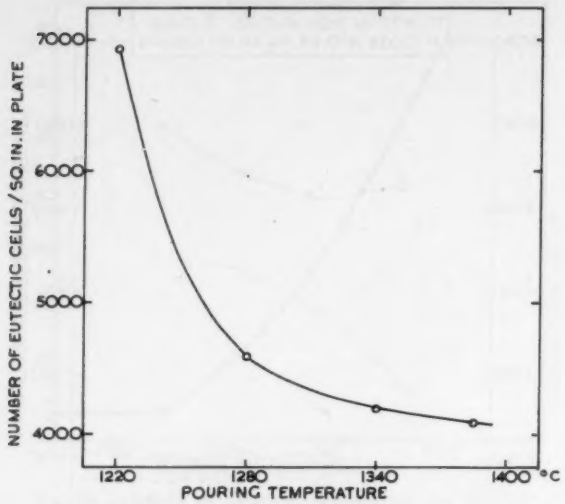


Fig. 36 — Relationship between pouring temperature and the number of eutectic cells growing.

upon the other factors reported. Normally, the design of a given casting is not a variable, although, due to human factors, small variations such as in the thickness may occur with unfortunate results, as is instanced by the brake-drum castings referred to earlier. A further example of the effect of small variations in the dimensions of a casting upon soundness is illustrated by experiments using plate-with-bosses test castings.[25]

GOOD DESIGN AND GATING IMPROVE SOUNDNESS

It was observed that increasing the plate thickness by $\frac{1}{32}$-in. increased the depth of sink beneath the bosses by approximately 0.001 in. This is large compared with the depth of sink of 0.005 in. which was arbitrarily concluded in earlier work upon surface shrinkage defects to render a casting unsatisfactory, particularly if it were to be subsequently enamelled.[26]

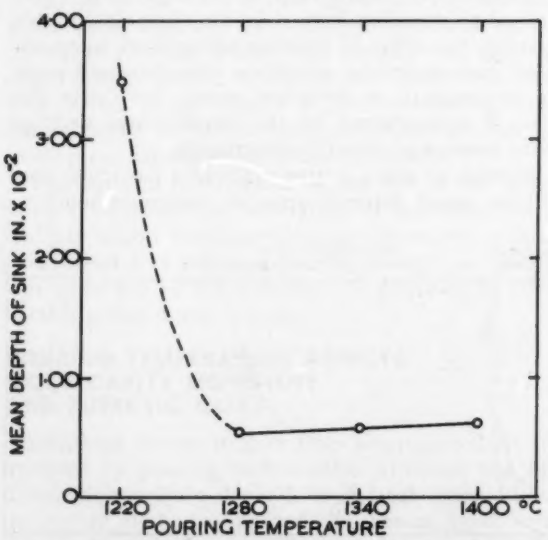


Fig. 37 — Pouring temperature variations effect on sinking.

Differences in thickness of this magnitude may result from rapping the pattern during removal from the mold; small variations in practice of this nature may account for sporadic outbreaks of shrinkage defects.

It is common experience that a particular design of casting in a foundry is more prone to exhibit shrinkage defects than other designs of casting on the produce line produced under similar conditions. Such castings are particularly sensitive to small changes in metal composition and foundry practice. This effect has been shown in the laboratory[27] where K-bar test castings of a certain size were more prone to drawing defects than those of either larger or smaller size. This result is illustrated in Fig. 38. Similar experiences are often reported from foundries making a given casting in a wide range of sizes.

In general, increasing the surface area/volume ratio of green sand molded castings causes a decrease in internal shrinkage, as has been illustrated by Timmins[1] and Szajda.[27] This effect is thought to be due to the lower amount of heat delivered to the mold per unit of casting surface area.

Gating plays an important part in the production of sound castings. Although a considerable amount of research work has been carried out to determine hydraulic factors affecting metal flow through a gating system and the principles of metal flow are well understood,[28] little work has been published upon the influence of the gating system upon the soundness of gray iron castings. Experiments so far[25] have shown that for plate castings bearing bosses it is an advantage to use a stoppered sprue, and a large runner: sprue ratio is desirable, as shown in Fig. 39. Slight pressurization is an advantage.

There is an interaction between the gating system as a whole and soundness, and although the effect of the individual components may be isolated, the most important aspect is the combined effect of all

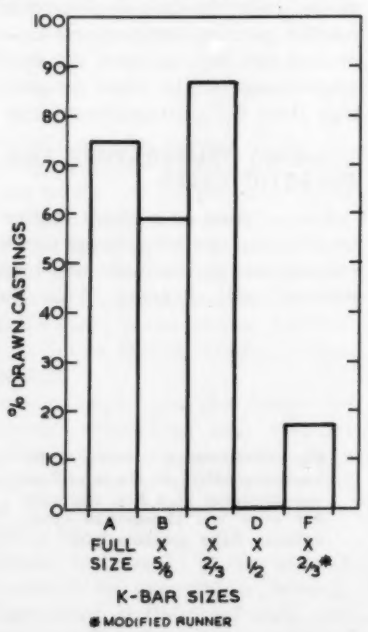


Fig. 38 — Incidence of drawing in K-bar test pieces varying with size and design of casting.

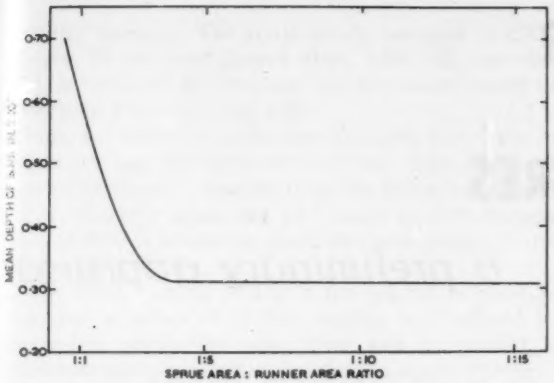


Fig. 39 — Sprue area: runner area ratio associated with minimum sinking at a number of gating ratios. Data abstracted from experimental results.

the variables. This result is at variance with the opinion of some other workers, who suggest that the type of gating employed for iron castings is unimportant provided that the inflow velocities are not sufficiently excessive to cause deterioration of the mold. In general, the commonly recommended gating ratios of the order of 1:1:<1 are not the best for avoiding sinking in this type of casting, and ratios of 1:4:0.5 or 1:5:1 have proved satisfactory. Such systems are slightly pressurized when head losses are considered.

When several castings are made from the same runner they are likely to be different, unless the system is perfectly symmetrical with regard to metal flow. The soundness of castings made as a spray about a long runner will vary considerably according to their relative positions with respect to the sprue.

All variables of design and gating are believed to influence temperature distribution within the casting, and so cause variations in soundness in much the same way as variations in pouring temperature. It is considered that, for flake graphite irons, the gating arrangement and the design of the casting should be such that a uniform temperature distribution is obtained throughout the casting to promote nondirectional solidification.

This leads to the conclusion that it is undesirable for more metal than is necessary to flow through a casting, and gating such that metal flows through one casting to reach another is definitely bad practice. On the other hand, for light-section castings, the use of flowoffs to prevent the first cold metal from entering the mold may not only help to avert cold shuts but may also greatly improve soundness.

ACKNOWLEDGMENT

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REFERENCES

1. Timmins, A. A., "Some Factors Affecting the Formation of Shrinkage Defects in Cast Iron," *B.C.I.R.A. Journal of Research and Development, vol. 2,* Oct., pp. 19-36, Dec., pp. 47-62, 1947.
2. Gittus, J. H., "The Influence of Mould Factors on Casting Defects," *B.C.I.R.A. Journal of Research and Development, vol. 5,* April, pp. 264-277, 1954.
3. Hughes, I. C. H., "Solidification Sequences and Their Significance," *B.C.I.R.A. Journal of Research and Development, vol. 5,* Feb., pp. 518-536, 1955.
4. Gittus, J. H., and Hughes, I. C. H., "The Solidification of Castings in Relation to their Soundness," *B.C.I.R.A. Journal of Research and Development, vol. 5,* Feb., pp. 537-554, 1955.
5. Fuller, A. G., and Hughes, I. C. H., "A Study of the Solidification Process in Grey and White Irons," *B.C.I.R.A. Journal of Research and Development, vol. 7,* Aug., pp. 288-298, 1958.
6. Gittus, J. H., "Formation of Shrinkage Defects in Grey Iron Castings: Influence of the Mould," *Iron and Steel, vol. 28,* July, pp. 331-335, Aug., pp. 387-390, 1955.
7. Nicholas, K. E. L., and Hughes, I. C. H., "Some Physical Factors Influencing Shrinkage and Porosity in Grey Iron Castings," *British Foundryman, vol. 51,* Sept., pp. 428-435, 1958.
8. Hughes, I. C. H., and Gittus, J. H., "Shrinkage Defects in Grey Iron Castings," *Foundry Trade Journal, vol. 101,* July 19, pp. 71-83, 1956.
9. Saeger, C. M., and Ash, E. J., "Properties of Gray Cast Iron as Affected by Casting Conditions," AFA TRANSACTIONS, *vol. 41,* pp. 449-468, 1933.
10. Nicholas, K. E. L., "The Effects of Mould Rigidity, Moulding Sand Properties and Pouring Temperature on the Soundness of Unfed Grey Iron Castings," *B.C.I.R.A. Journal of Research and Development, vol. 7,* Dec. 1958.
11. Womochel, H. L., and Sigerfoos, C. C., "Influence of the Mold on Shrinkage in Ferrous Castings," AFS TRANSACTIONS, *vol. 48,* pp. 591-618, discussion, pp. 619-622, 1940.
12. Wallace, J. F., and Evans, E. B., "Risering of Gray Iron Castings," AFS TRANSACTIONS, *vol. 66,* pp. 49-55, 1958.
13. Piwowarsky, E., *Hochwertiges Gusseisen,* Springer-Verlag, Berlin, 1070 pp., especially Chapter 8, 1958.
14. Schmidt, W. A., Sullivan E., and Taylor, H. F., "Risering of Gray Iron Castings: Progress Report No. 5," AFS TRANSACTIONS, *vol. 62,* pp. 70-77, 1954.
15. Fuller, A. G., "The Soundness of Grey and Nodular Iron Castings," *B.C.I.R.A. Bulletin and Foundry Abstracts, vol. 14,* Nov., pp. 569-574, 1958.
16. Wittmoser, A., Krall, H. A., and Hutter, L., "Testing Cast Iron for Shrinkage Cavities," *Giesserei, vol. 43,* Aug. 2, pp. 409-418, in German, 1956.
17. Jeffery, W. C., Langner, E. E., Mitchell, W. G., and Azizi, G. D., "Relationship of Carbon Equivalent to the Properties of Cast Iron," AFS TRANSACTIONS, *vol. 62,* pp. 568-577, discussion, pp. 577-578, 1954.
18. Gittus, J. H., "Shrinkage Defects in K-bar Test Castings," *B.C.I.R.A. Journal of Research and Development, vol. 6,* Feb., pp. 456-483, 1957.
19. Hughes, I. C. H., "Surface Sinking Defects in Grey Iron Light Castings," *British Foundryman, vol. 50,* April, pp. 202-210, discussion, Oct., pp. 520-524, 1957.
20. Morrogh, H., "The Problem of Grain Size in Cast Iron," *British Foundryman, vol. 51,* Nov., pp. 588-591, 1958.
21. Fuller, A. G., "Further Studies on Some Factors Influencing Sinking under Bosses on Thin Plates: The Role of Eutectic Nucleation," *B.C.I.R.A. Journal of Research and Development, vol. 7,* Feb., pp. 157-170, 1958.
22. Fuller, A. G., paper to be published in *British Foundryman.*
23. Rickard, J., "A Note on the Influence of Furnace Holding Time on Chill and Mottle Formation," *B.C.I.R.A. Journal of Research and Development, vol. 7,* Aug., pp. 23-26, 1957.
24. Dews, J. W., and Fuller, A. G., "Sinking Under Bosses on Thin Plates: A Study of the Role of Nucleation in the Foundry," *B.C.I.R.A. Journal of Research and Development, vol. 7,* Dec., 1958.
25. Fuller, A. G., "The Effect of Design Variables Upon Shrinkage Defects under Bosses on Thin Plates," *B.C.I.R.A. Journal of Research and Development, vol. 7,* Oct., pp. 54-67, 1957.
26. Hughes, I.C.H., "Sinking under Bosses on Thin Plates: Preliminary Experiments on Foundry Variables," *B.C.I.R.A. Journal of Research and Development, vol. 5,* April, pp. 616-627, 1955.
27. Szajda, T. J., "Effect of the Size of K-bar Test Castings on Their Shrinkage Defects," *B.C.I.R.A. Journal of Research and Development, vol. 7,* Dec., pp. 110-128, 1957.
28. Ruddle, R. W., *The Running and Gating of Castings: A Review of the Literature,* Institute of Metals, London, 183 pp, 1956.

BENDING TECHNIQUES FOR EVALUATION OF CAST MATERIALS AND STRUCTURES

a preliminary appraisal

By J. C. Graddy

ABSTRACT

The accepted practices of inspecting the mechanical properties of castings are reviewed and found wanting. Three examples are given of parts that were weaker when made of high tensile strength material than when made of material of lower tensile strength. Therefore, bending tests are investigated to obtain a truer picture of structural behavior.

A cast bending specimen is described, and the test results for 28 different conditions of chemistry and heat treatment in aluminum and magnesium alloy are given. A machined bending bar is described and its results on the 28 different conditions are given. A study of the effect of geometry on bending modulus of rupture specimens of wrought materials is described and the results given.

A "shape of the curve" parameter for acceptability of bending specimens is offered and future work on bend testing is proposed.

TEST METHODS REVIEW

Inspection for specification mechanical properties in the alloys of aluminum and magnesium has been predominantly by tests of "cast-to-size" tensile specimens. These are produced by a gating system that is controlled by a specification standard. The casting specifications acknowledge the difference in the mechanical properties of these cast-to-size bars and of bars machined from the casting. This difference has led to the accepted practice of using a "casting factor" to reduce the specification mechanical properties for design use.

The writer believes that the cast-to-size test bar is a measure of the chemistry balance, melting practice and heat treatment of the test bar itself. It is a useful tool for foundry process control, but it is practically worthless as a measure of the mechanical properties contained in the cast structure.

Some investment foundries have even gated directly into the gage length of the cast-to-size bar. This makes the procedure a complete farce from an inspection viewpoint because of the excessive risered surface.

The ferrous foundries have adopted a method of separately casting a keel block from which tensile specimens are machined. This block is grossly overrisered when compared to the casting it is supposed to represent. Hence it is of even less value as an inspection method than the cast-to-size bars.

In the past, the Air Force required static testing to destruction of three castings from the first production lot. This was done to establish an acceptable quality level for radiographic inspection. Test engineers of the author's company added the practice of cutting a tensile specimen from the failed casting. The tensile strength values obtained were then used as a leveling factor for the static test results. This was always done if the results were higher than specification properties.

About three years ago the author's company began to specify that a tensile specimen be machined from a critical area of production castings on a sampling basis. The composition of the casting lot was carefully limited to insure a representative sample for inspection. The testing of machined tensile specimens, combined with the control of lot composition, is probably the first valid method for inspection of mechanical properties of cast structures.

During this same period they developed a beam-type casting for a critical structural application. A destruction test of a sample beam from each lot was selected as the best inspection method to determine bending behavior. In this particular casting, this test also proved cheaper than machining tensile specimens from the casting.

A destruction-type test can only be used on a sample plan. It can be an expensive method unless both the machining to prepare the rough casting and the loading system are simple.

TENSILE TESTS MAY BE MISLEADING

Most cast structures must be able to withstand bending loads. However, they are usually a complex shape with corners or notches in the load paths of the structure. Mechanical properties of tensile specimens have been used for design because these are the only properties given in the specifications.

Comparative tests show that material having a higher tensile strength is often weaker in the actual part than a lower strength material when the part has a high mechanical notch factor.

Figure 1 shows a comparison of aluminum alloys A356 and C355 in a threaded tubular joint under a

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bending moment. The static tensile strength of C355 alloy is 20 per cent greater than A356, but the cast part is actually 20 per cent weaker when made of C355 than when made of A356.

Figure 2 shows a comparison of A356 and C355 in the beam casting mentioned above. This shape is much less severely notched than the threaded joint of Fig. 1, however again the part made of the stronger alloy (C355) was weaker than the part made of the weaker alloy (A356).

The beam casting of Fig. 2 was placed in production, and a group of 40 test castings was ordered to determine production variability and to predict a minimum margin of safety for the structure. The production variability was accomplished by using four different melts of metal, and then some of each of the four melts were assembled into three heat treating furnace loads.

Table 1 shows the strength of the castings and the mean for each condition. A statistical analysis of the test indicates that heat treatment variations (in the same heat treat schedule) are 1½ times as significant as chemistry in producing a maximum strength casting. These castings, were produced under carefully controlled conditions, and two tensile specimens were cut from an extra casting of each melt. The tensile tests showed a range of 1.6 per cent, while the beam bending showed a range of 13.9 per cent.

During the two years the beam casting has been in large production, records have been kept of the acceptance tests. Figure 3 shows the load distribution of the production bend samples of this casting with the load distribution of the 40 test castings superimposed. Production acceptance bend samples plotted against the tensile samples which were cut by the foundry to monitor the heat treating operation are shown in Fig. 4.

Figure 5 shows the static test results of an investment cast lug of 4340 alloy steel heat treated to two levels of tensile strength. The 125,000 psi level had a mean static strength 14 per cent higher than the 160,000 psi castings. The range of failure load of the 175,000 psi parts was 2½ times the range of the 125,000 psi casting.

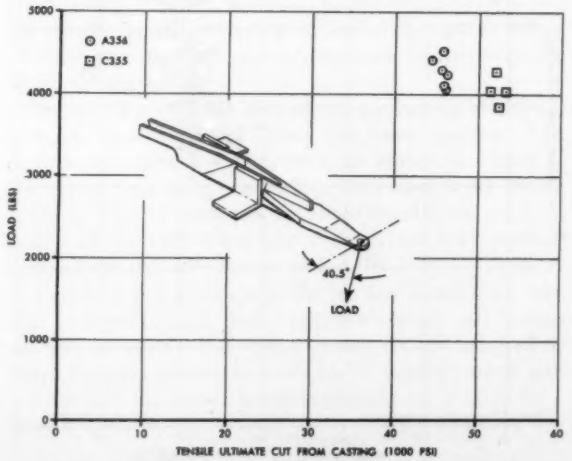


Fig. 2 — Comparison of A356 and C355 tests for a beam casting.

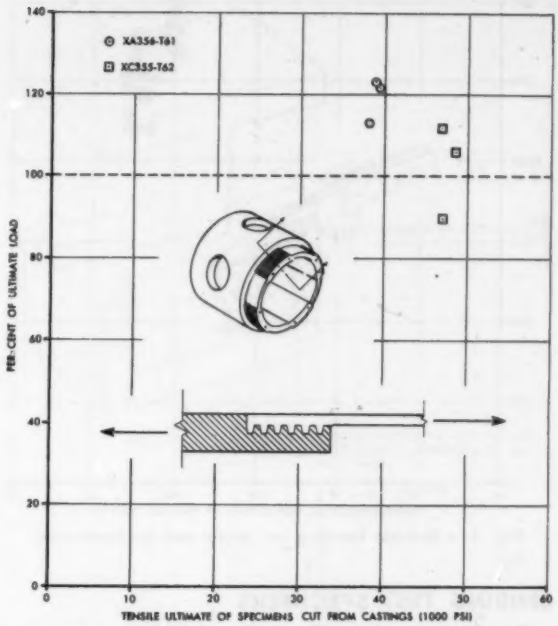


Fig. 1 — Comparison of aluminum alloys A356 and C355 tests for a threaded tubular joint.

TABLE 1 — BEND TESTS OF CONTROLLED TEST LOT OF BEAM CASTING

Heat Treat Lot	Chemistry Lot A	B	C	D	Mean
I	5360	5000	5360	5080	
	5200	5080	5360	5160	5153
	5160	4920	5350	4800	
II	5360	5480	5560	5400	
	5160	4960	5520	5320	5320
	5040	5120	5640	5280	
III	4920	4960	5120	5160	
	5120	5080	5350	5040	5101
	5160	5160	5020	5080	
Mean	5164	5084	5369	5147	

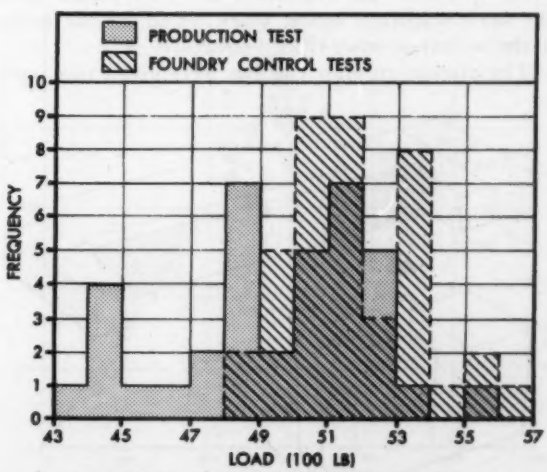


Fig. 3 — Distribution of strength tests of beam casting. The load distributions of the 40 test castings are superimposed.

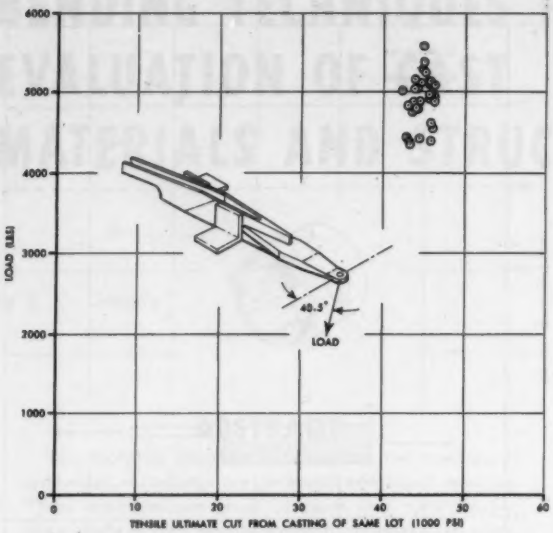


Fig. 4 — Sample bending vs. tensile test lot beam casting.

BENDING TEST SPECIMENS

These examples indicate that bending strength is not necessarily proportional to tensile strength in castings, and that some investigation should be made. To accomplish this, a cast-to-size specimen was designed incorporating notched load path as severe as had been present in many past designs. The casting was designed as an "H" section, and the pattern was made with matching iron chills and boxes for sand cores at the critical bending area. The combination of chills and sand cores made it possible to study the effect of chilling on bending strength with a minimum of other variables.

After casting, the "H" section was cut into two "tees" for tests as a cantilever beam or as tensile and bend specimens cut from the tee. Figure 6 shows the casting and its chill areas. This casting configuration will be referred to hereafter as the tee bar. A second objective of this configuration was to develop a cut specimen for inspection use that would have the same statistical order, when tested, as the order of the tee bar or other type of structure.

The aluminum alloy castings were given two types

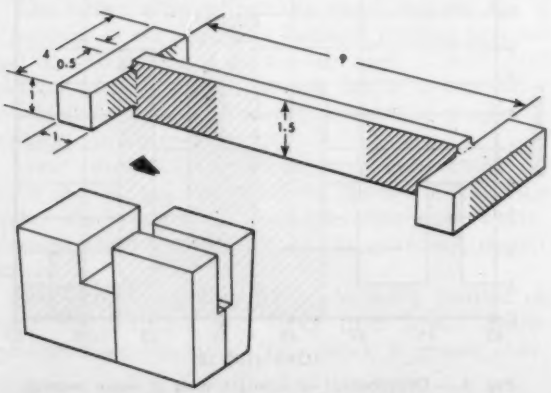


Fig. 6 — Tee bar casting for bend testing, with chill areas shown.

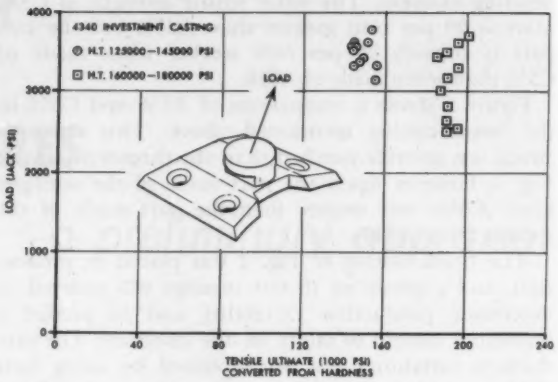


Fig. 5 — 4340 investment cast steel at two heat treating levels.

of aging treatment to determine its effect on bending strength. The longer aging time improved the performance of the chill cast material, and in some cases lowered the performance of the sand cast material. A second purpose of the two types of aging treatments was to provide a greater range in elongation to determine if there is a minimum value of elongation for structural use.

Figure 7 shows the cantilever bending test results for both sand cast and chill cast tee bars of several aluminum and magnesium alloys. These castings were all produced as special test parts under carefully controlled foundry conditions, and were free of radiographic defects at the critical bending area. Therefore, the results represent the best that can be expected from the alloys, and are above the typical values for an average production casting. Figure 8 shows the average value for each alloy and heat treatment.

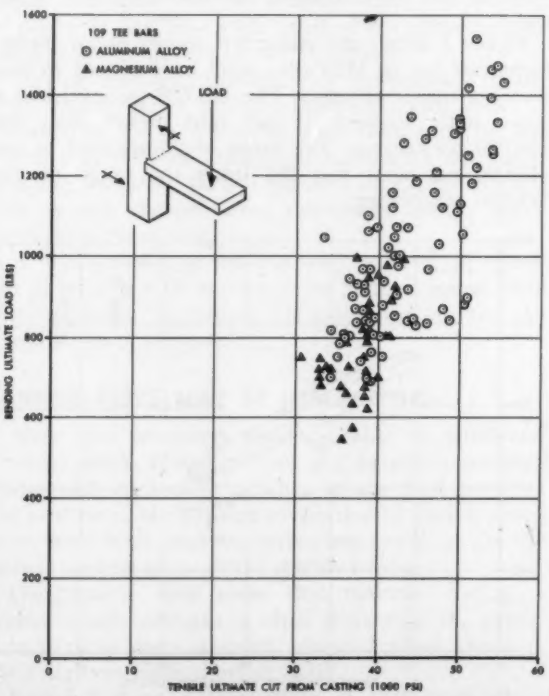


Fig. 7 — Tee bar cantilever beam test vs. cut tensile tests for sand cast and chill cast specimens.

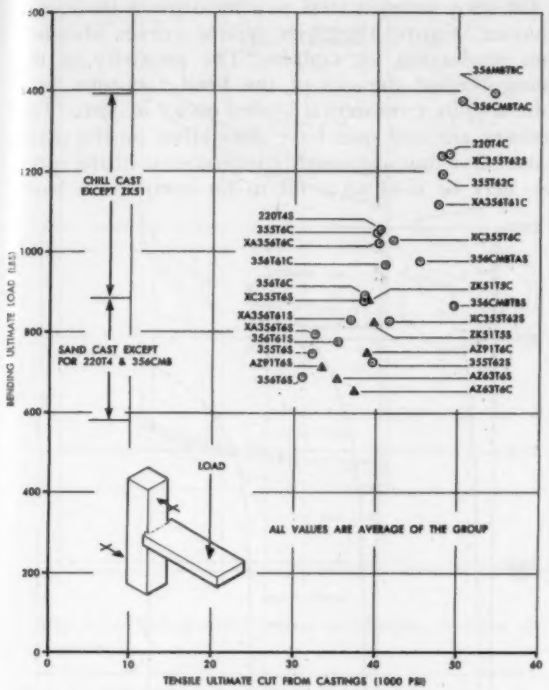


Fig. 8 — Average of tee bar bend vs. tensile tests for each alloy and heat treatment.

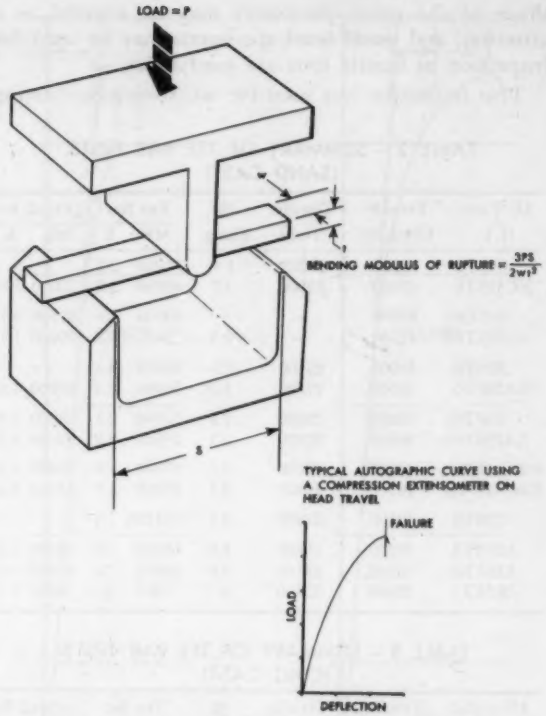


Fig. 9 — Bend test method for cut bars, showing a typical curve and the equation for computing the bending modulus of rupture stress.

The large range (two to one) of load for a given tensile stress, as shown in Fig. 7, indicates that tensile values are not reliable as an inspection method for bending performance of a material.

In order to obtain a better evaluation of bending behavior, the author's company began testing small bend specimens machined from castings. An arbitrary size was selected that could be cut from the same area of a casting from which tensile specimens were already being taken. The bending bar was ¼-in. thick and ⅜-in. wide, and was loaded as a simple beam with a 2-in. span. The ¼-in. thickness was thought to be deep enough to minimize the effect of small casting defects, and the machining was simple because of the constant cross-section without notches. Figure 9 shows the loading method, a typical curve and the equation for computing the bending modulus of rupture stress. (The appendix gives an explanation of the modulus of rupture effect.)

Figure 10 shows the plot of the tee bar vs. the ¼-in. x ⅜-in. bend bar cut from the same casting. The same large range is still present, as shown in Figure 7. The materials tested included some that were intentionally heat treated to a brittle condition, such as sand cast 355T62. This brittle condition may explain some of the low side scatter of Fig. 10.

Tables 2 and 3 compare the tee bar bend tests, tensile tests and small bend specimen tests (all values are the average value for a group of tee bars). The bend tests are shown as modulus of rupture stress and the "K" values $\frac{\text{(modulus of rupture)}}{\text{(tensile ultimate)}}$ are also shown.

Figure 11 shows average value and autographic load vs. beam deflection curves for both the tee bars and the small bend bars. The shapes of the curves are similar, which indicates that the shape characteristic will prove to be the best inspection method for ductile behavior in bending structures. A particular

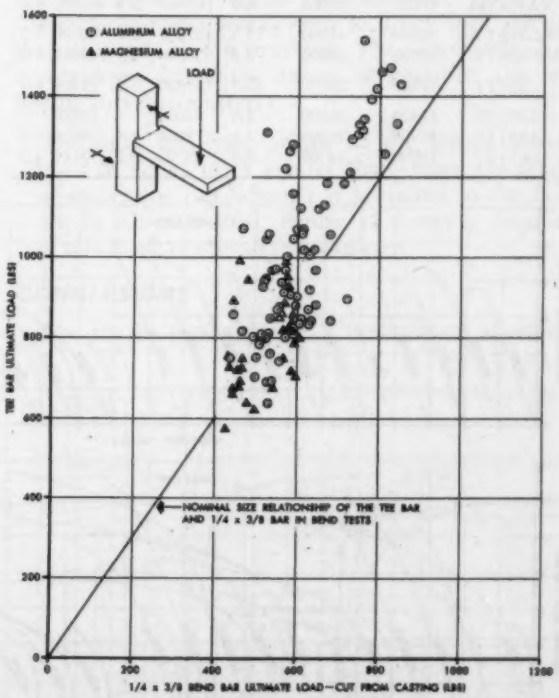


Fig. 10 — Tee bar cantilever test vs. cut bending bars taken from the same casting.

shape of the curve parameter may be selected as a criterion, and small bend specimens may be used for inspection as tensile tests are used today.

This technique was used for an investment casting in 410 alloy stainless steel as a measure of its ductile behavior. Figure 12 shows typical curves obtained from production lot samples. The geometry of the casting limited the size of the bend specimen to a ¼-in. x ¼-in. cross-section loaded on a 1-in. span. The specimen size and span have some effect on the shape of the curve, but any consistant geometry of the specimen may be used on a lot to lot comparison basis.

TABLE 2 — SUMMARY OF TEE BAR TESTS (SAND CAST)

Alloy and H.T.	Tensile, Ultimate	Tensile, Yield	% Elong.	Tee Bar MR	Tee Bar K	¼x⅜x2 Bar MR	¼x⅜x2 Bar K
355T6	32200	27200	1.4	71400	2.2		
XC355T6	38500	33800	1.7	83900	2.2	70000	1.8
355T62	39600	—	—	69300	1.8	67300	1.7
XC355T62	41700	—	0.9	79400	1.9	70400	1.7
356T6	31000	25400	2.5	66100	2.1		
XA356T6	32500	27000	3.0	76000	2.3	61900	1.9
356T61	35200	29900	2.2	74500	2.1	64000	1.8
XA356T61	36800	32800	2.3	79600	2.2	69000	1.9
356CMBTA	45200	37000	5.1	87800	1.9	79400	1.8
356CMBTB	49500	45300	2.4	83000	1.7	84400	1.7
220T4	40300	27900	8.1	101100	2.5		
AZ63T6	35200	19800	3.2	65900	1.9	56500	1.6
AZ91T6	33300	23100	3.0	68600	2.0	57900	1.7
ZK51T5	39400	23400	9.1	79300	2.0	70600	1.8

TABLE 3 — SUMMARY OF TEE BAR TESTS (CHILL CAST)

Alloy and H.T.	Tensile, Ultimate	Tensile, Yield	% Elong.	Tee Bar MR	Tee Bar K	¼x⅜x2 Bar MR	¼x⅜x2 Bar K
355T6	40300	28800	5.3	100800	2.5	69500	1.7
XC355T6	42000	32600	5.7	98600	2.4	82400	2.0
355T62	48000	42700	2.8	114400	2.4	87000	1.8
XC355T62	48200	42200	2.2	118700	2.5		
356T6	38500	29800	4.4	85300	2.2	68500	1.8
XA356T6	40200	28300	10.2	97700	2.4	75800	1.9
356T61	41000	32400	3.8	92700	2.3	75000	1.8
XA356T61	47400	38200	4.0	107400	2.3	85300	1.8
356CMBTA	50300	37400	13.3	137400	2.7	96500	1.9
356CMBTB	54200	45200	8.4	133300	2.5	103300	1.9
220T4	47800	25700	12.3	119300	2.5	76300	1.6
AZ63T6	37400	21000	3.6	62600	1.7	64200	1.7
AZ91T6	38900	22200	4.6	71800	1.8	74600	1.9
ZK51T5	38700	24700	8.4	83700	2.2	73100	1.9

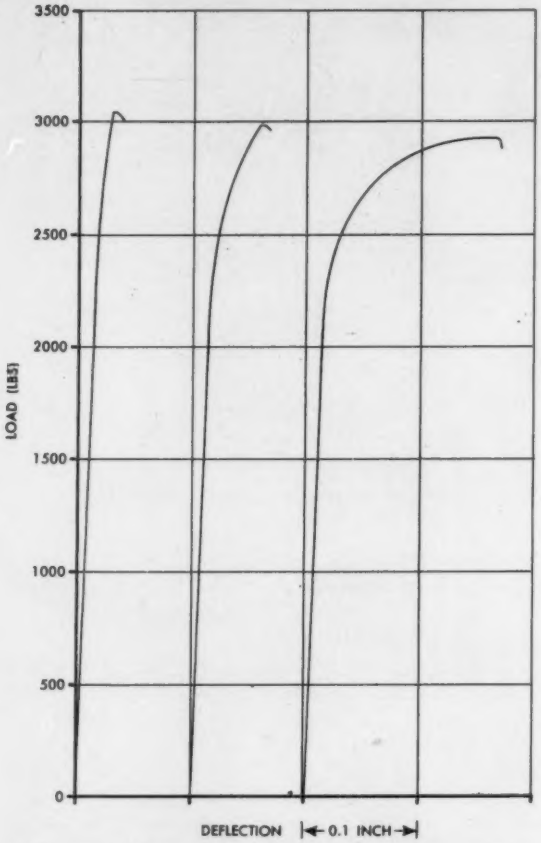


Fig. 12 — Bend tests of 410 stainless steel investment casting obtained from production lot samples.

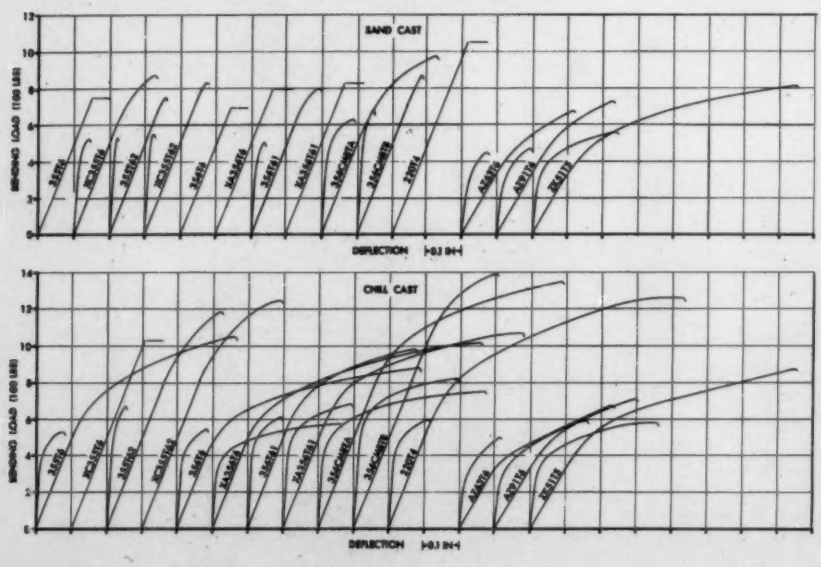


Fig. 11 — Typical curves of tee bars and cut bending bars showing average value and autographic load vs. beam deflection curves.

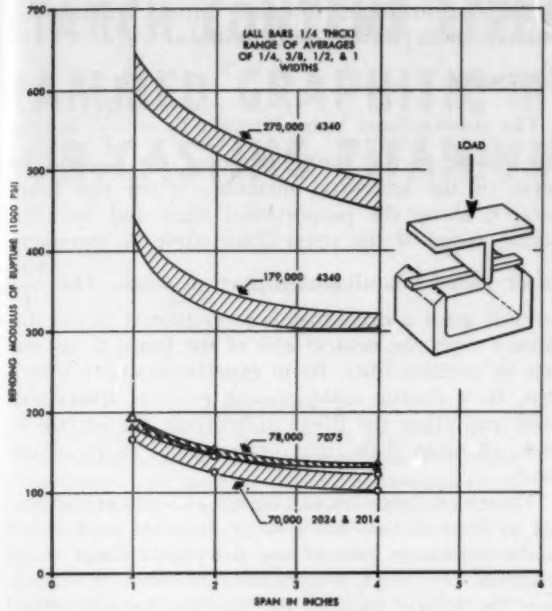


Fig. 13 — Bar geometry effect on bending modulus of rupture.

GEOMETRY EFFECT ON BEND SPECIMENS

A study was made of the geometry effect on bend specimens in wrought material. All specimens of a group were machined from the same piece of plate stock in an effort to minimize the effect of material variations. A constant thickness of ¼-in. was selected and widths of ¼, ⅜, ½ and 1 in. were tested over spans of 1, 2 and 4 in.

The width proved to have only a minor effect on the modulus of rupture over the range that was tested. Figure 13 shows the effect of span on the constant depth beams. Further work is needed to determine the mechanism of this variation before the high modulus values can be used for other than comparative testing. Figure 14 shows the effect of beam width on the shape of load vs. deflection curve, and must be taken into account on curve shape criteria.

A study of the shape of the various load deflection curves indicates that an acceptable level of that vague quality ductility may be defined as a parameter of the curve shape. This parameter may need to be different for various types of structural service.

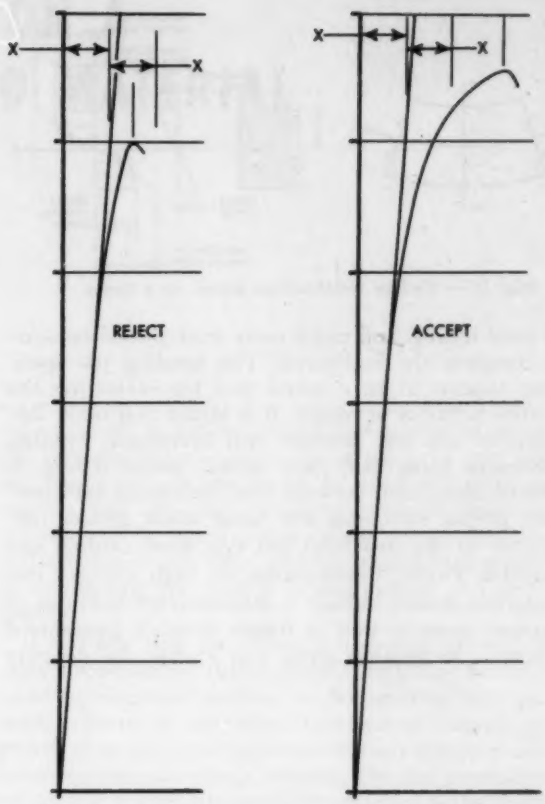


Fig. 15 — Shape parameter for load—deflection curves.

We are really looking for some property "X" that demonstrates the ability of a material to redistribute a load around a notch in the load path. Therefore, the curve parameter may need to differ with the degree of notch in the structure. The writer offers, as a minimum acceptable degree of ductility, the following curve parameter:

> The total deflection offset, at the failure point, must be equal to or greater than, twice the offset of the elastic line when it is projected to the failure or ultimate load. Figure 15 shows a diagram of this shape of the curve parameter.

CONCLUSIONS

The results presented are a preliminary appraisal

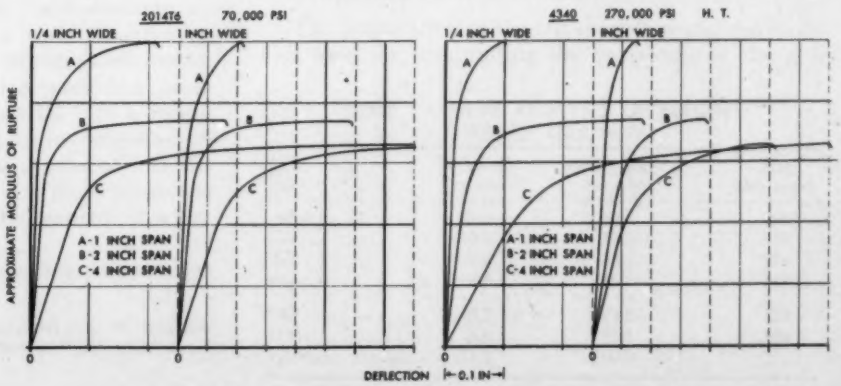


Fig. 14 — Bar width effect on stress—deflection curves.

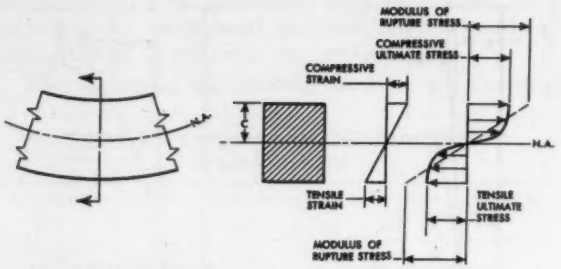


Fig. 16 — Various relationships shown on a beam.

of bend testing, and much more work should be done to complete the evaluation. The bending test specimen appears to be a useful tool for evaluating the ductile behavior of metals. It is hoped that other laboratories and test facilities will investigate bending specimens along with their normal tensile testing. A proved correlation between bend specimens and static tests of cast structures will build much greater confidence in the structural job that good castings can perform. Future specifications for high strength cast materials should include a demonstrated modulus of rupture stress as well as tensile stress. A guaranteed modulus of rupture stress will enable the designer to use cast structures in more critical applications and have more efficient configurations.

APPENDIX

The conventional beam formula $F_b = \frac{MC}{I}$ is true for stresses in the elastic range of the stress-strain curve of the structural material. When the beam stress is above the proportional limit and into the plastic range of the stress-strain curve, a correction factor called modulus of rupture is used. The $\frac{MC}{I}$ formula gives a stress that is proportional to the distance c from the neutral axis of the beam to its surface or extreme fiber. Beam experiments have shown that, in a ductile material, the extreme fibers will yield and allow the fibers underneath the surface to pick up more than their proportional share of the load.

Therefore, modulus of rupture is a correction factor applied to the elastic beam formula to make it usable for beams stressed into the plastic range of the material. It is given as an allowable factor of increase over the tensile stress of the material (usually called K) or, as an equivalent fictitious tensile stress. Figure 16 shows these relationships on a beam.

STARCH CONTENT EFFECT ON A RAMMED GRAPHITIC MOLD MATERIAL FOR CASTING TITANIUM

By H. W. Antes and R. E. Edelman

ABSTRACT

The starch content effect on the properties of the rammable graphitic mold material was determined. Specimens were made by molding at pressures of 30, 60 and 120 psi for graphitic mixtures containing 4, 6, 8 and 10 per cent starch by weight. It was found that the best combinations of properties were obtained with mixtures containing 6 to 8 per cent starch. A mixture containing 6 per cent starch and molded at 120 psi provided molds which were equivalent to machined graphite molds with respect to minimizing surface contamination of resulting titanium castings.

It has been found that sometimes it is necessary to use starch contents higher than 6 per cent to facilitate the removal of complex shaped patterns without damaging the mold. In this case, the starch content may be increased but it should not be in excess of 8 per cent, since maximum strengths are obtained with an 8 per cent starch mix. Maximum density and thermal conductivity also are observed with an 8 per cent starch mix. In addition, it has been found that mold shrinkage is directly proportional to starch content and inversely proportional to molding pressure. The mold surface finish is good and is not sensitive to starch content for the range studied.

INTRODUCTION

The development of the rammable graphitic mold material has facilitated the production of shaped castings of titanium and other high melting point reactive metals.[1,2,3] Common foundry molding techniques have been used with the graphitic material to make complex molds and cores. After the material is molded it is air dried, baked and finally fired to produce an entirely graphitic and carbonaceous mold.

The constituents used in preparing the graphitic material are powdered graphite, carbonaceous cement, starch, powdered pitch, a surface activating agent and water. The proportion of these constituents as originally proposed produced a paste-like product.[1,2] In order to refine the material into a consistency more desirable for molding an investigation on the effect of moisture content was conducted.[3] It was found that by properly mulling the material it was necessary to use only ¼ to ⅓ the amount of water that was proposed originally.

In addition, it was found that molding pressure was an important variable. As a result of the investigation the optimum moisture content and minimum molding pressure were found to be 6 per cent and 100 psi, respectively. It was found, by using this moisture content and molding pressure, rammed graphitic molds could be made which closely approached the performance of machined graphite molds with respect to minimizing the surface contamination of cast titanium.

It was felt that the chemical inertness as well as other properties of the rammable graphitic material could be improved further by determining the optimum content of the other constituents. Starch appeared to be the most important of the remaining constituents, since shrinkage, density, thermal conductivity, as well as strength should be sensitive to the starch content. Therefore, starch content was selected for study in this investigation.

MATERIALS

Electric furnace graphite powder was used as the basic material. The sieve analysis of this material is given in Table 1.

The AFS grain fineness number for this material was approximately 85. The other constituents were instant laundry starch (finely powdered), carbonaceous cement, pulverized pitch, a surface activating agent and tap water.

The standard mix shown in Table 2 was used to prepare the specimens. Various amounts of starch were added to this mixture to give 4, 6, 8 and 10 per cent starch by weight.

PROCEDURE AND METHODS

The preparation of the test specimens and the methods used for determining the properties of the gra-

TABLE 1 — SIEVE ANALYSIS OF ELECTRIC FURNACE GRAPHITE

U.S. No.	Retained, Per Cent	Cumulative Per Cent	Size (Microns)
60	26.5	26.5	246
70	10.8	37.3	210
80	16.0	53.3	175
100	12.3	65.6	147
120	9.7	75.3	125
140	5.8	81.1	104
PAN	18.9	100.0	

H. W. ANTES is Met., and R. E. EDELMAN is Chief, Reactive Metals Unit, Research and Development Group, Frankford Arsenal, Philadelphia.

TABLE 2 — MOLD MATERIAL COMPOSITION USED FOR SPECIMEN PREPARATION

Component	Part by Weight
Graphite Powder AFS No. 85	61.0
Pitch	11.5
Carbonaceous Cement	9.0
Water	6.0
Surface Activating Agent	1.0

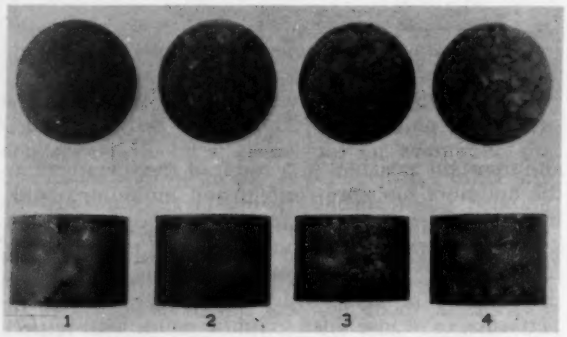


Fig. 1 — Surface standards by rating numbers for graphitic mold material.

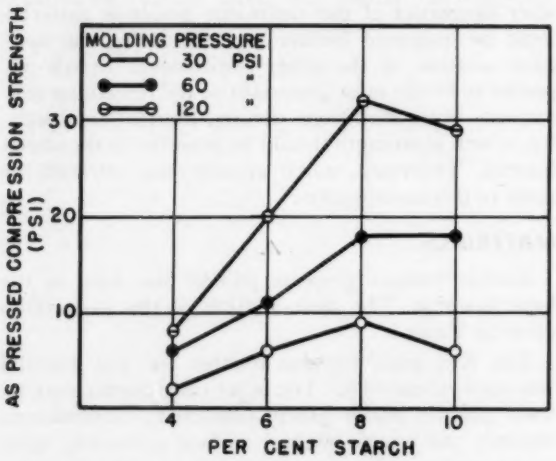


Fig. 2 — As-pressed compression strength vs. per cent starch for graphitic mold material.

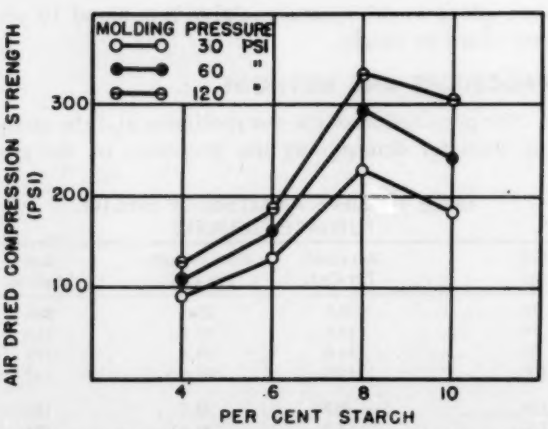


Fig. 3 — Air dried compression strength vs. per cent starch for graphitic mold material.

phitic material have been presented in detail in a previous report.[3]

The starch contents used were 4, 6, 8 and 10 per cent by weight. The molding pressures that were employed for each of the starch contents were 30, 60 and 120 psi.

The properties that were determined for each of the 12 combinations of starch content and molding pressure are listed below:

1) Surface finish.
2) Compression strength.
3) Density.
4) Permeability.
5) Thermal conductivity.
6) Shrinkage.
7) Surface contamination of resulting titanium castings.

TEST RESULTS

Surface Evaluation

An arbitrary set of standards for evaluating the surface finish of the molded specimens after firing had been established in a previous investigation. These standards, reproduced in Fig. 1, represent a gradient from the smoothest to the roughest surfaces that were obtained when the moisture content was varied from 4 to 10 per cent with molding pressures from 30 to 120 psi.

In the current investigation smooth surfaces (ratings of 1 or slightly more than 1) were obtained for all starch contents and molding pressures. This indicates that good surface finish can be obtained consistently as long as the water content is kept at a low level.

Compression Strength

The compression strengths vs. per cent starch, immediately after molding (as pressed), after air drying for 4 days (air dried), after baking at 250 F for 6 hr (baked) and after firing in a reducing atmosphere to 2000 F (fired), are shown in Figs. 2, 3, 4 and 5, respectively. In these figures it can be seen that a maximum is obtained with each molding pressure for the material containing 8 per cent starch.

In addition, by comparing these figures it can be seen that appreciable increases in strength are observed in going from one step of the process to the next, and that the strength is sensitive to both mold-

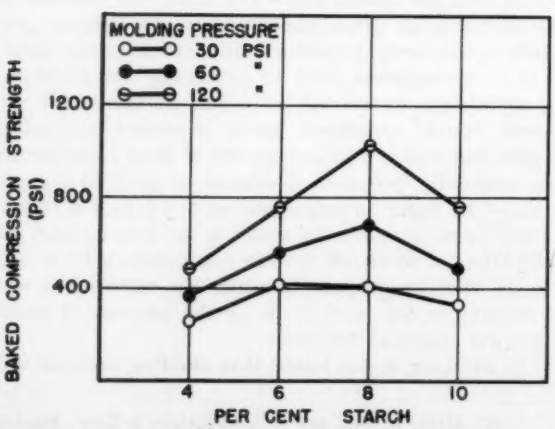


Fig. 4 — Baked compression strength vs. per cent starch for graphitic mold material.

TABLE 3 — COMPRESSION STRENGTH OF GRAPHITIC SPECIMENS

Starch, %	Stress (psi)* at Molding Pressure (psi) 30	60	120
As Pressed			
4	2	6	8
6	6	11	20
8	9	18	32
10	6	18	19
Dried			
4	90	110	129
6	136	162	186
8	227	292	334
10	185	240	305
Baked			
4	254	366	490
6	420	560	760
8	410	680	1030
10	333	492	760
Fired			
4	698	896	1010
6	820	1166	1455
8	1095	1650	1840
10	792	1166	1550

* Average of three tests

ing pressure and starch content. The actual compression strength data are listed in Table 3.

Density

The density of the various specimens was calculated, after firing, from weight and volume measurement. The effects of starch content and molding pressure on the density are shown in Fig. 6. In this figure it can be seen that as the starch content is increased from 4 per cent the density increases slightly. At approximately 8 per cent starch the density reaches a maximum. As the starch content is increased to 10 per cent a decrease from the maximum is observed.

From these curves it can be concluded that molding pressure has a greater effect on the density than has starch content. Higher molding pressures produce higher densities. The variation in density for the various specimens ranged from 0.93 to 1.08 g/cm³, as compared to a density of approximately 1.75 g/cm³

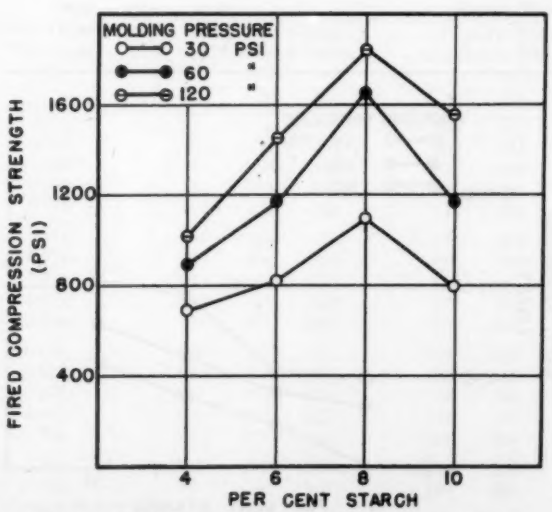


Fig. 5 — Fired compression strength vs. per cent starch for graphitic mold material.

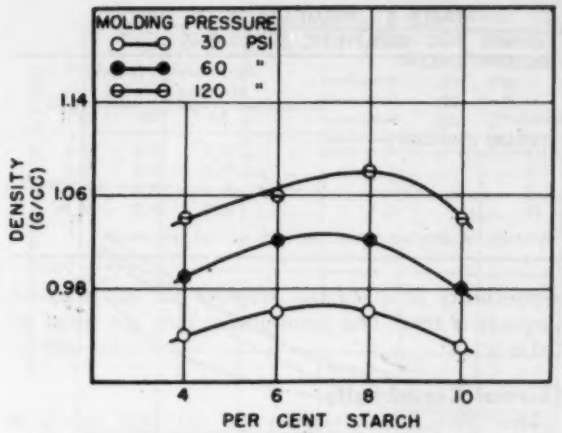


Fig. 6 — Density vs. per cent starch for graphitic mold material.

for machined graphite. These data are listed in Table 4.

Permeability

The permeability vs. per cent starch is plotted in Fig. 7 for the three molding pressures employed. This figure shows that as the starch content is increased from 4 per cent the permeability decreases to a minimum at 8 per cent and then increases. The initial decrease is substantially linear for the two higher molding pressures. Actually, the change in permeability (82 to 89) with starch content is slight for specimens molded at the lowest pressure. The range of permeability that was observed for the molding pressures and starch contents was approximately 30 to 90. The

TABLE 4 — DENSITY OF FIRED GRAPHITIC SPECIMENS

Starch, %	Density (g/cm³) at Molding Pressure (psi) 30	60	120
4	0.94	0.99	1.04
6	0.96	1.02	1.06
8	0.96	1.02	1.08
10	0.93	0.98	1.04

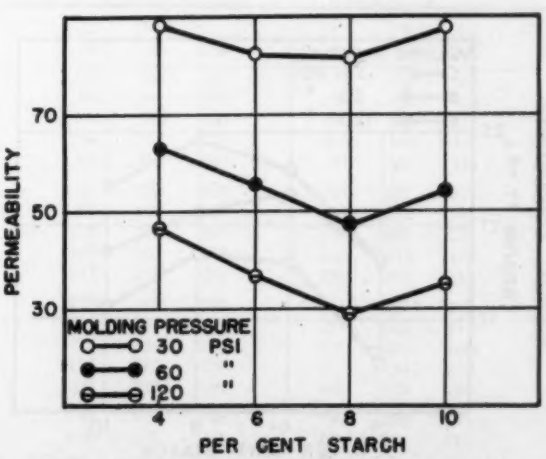


Fig. 7 — Permeability vs. per cent starch for graphitic mold material.

TABLE 5 — PERMEABILITY OF FIRED GRAPHITIC SPECIMENS

Starch, %	Permeability Number* at Molding Pressure (psi) 30	60	120
4	88.6	63.2	46.2
6	82.3	55.4	36.8
8	81.5	47.3	29.0
10	87.7	54.0	35.1

* Average of duplicate tests on each of two specimens

permeability of machined graphite for all practical purposes is zero. The permeability data are listed in Table 5.

Thermal Conductivity

The thermal conductivity vs. per cent starch is plotted in Fig. 8. These curves illustrate the point that the highest conductivities are obtained at the highest molding pressure for each starch content, and that for the three molding pressures a maximum conductivity is observed in the 6 to 8 per cent starch range. These data are relative values and are used to express trends only.[3] The thermal conductivity data, which were calculated from a value of 80 (Btu/hr/ft/degree F) for machined graphite, are given in Table 6.

Shrinkage

The average mold shrinkage as a function of the starch content is plotted for the three molding pressures in Fig. 9. These shrinkage data are the change in linear dimensions from the as-pressed to the fired condition. Figure 9 shows that the shrinkage increases as the starch content increases, and decreases as the molding pressure increases. The actual shrinkage values for the various starch contents and molding pressures are given in Table 7.

Surface Contamination

The most important property of a mold material for casting titanium or other reactive metals is its chemical inertness in the presence of the molten metal. The standard procedure for determining this property is a Knoop microhardness traverse taken on the resulting casting. The microhardness traverses for the castings made with the 4, 6, 8 and 10 per cent starch content molds are shown in Figs. 10, 11, 12 and 13, respectively. In these figures the curve for the machined graphite mold has been corrected to the same base hardness as that of the rammable graphitic molds.

This was necessary because the data for the machined graphite molds were obtained from castings of titanium which had a slightly higher base hardness than that observed in this investigation.[3] It can be seen from these curves that the performance of the rammable mold material compares favorably with the machined graphite molds. The best results were obtained with the 6 and 8 per cent starch molds.

In Fig. 11 it can be seen that the curves for the 6 per cent starch—120 psi molding pressure and the machined graphite mold are almost coincident. The base hardness of the titanium used in this investigation is approximately 120 Bhn. All of the microhardness data are listed in Table 8.

DISCUSSION

The results of this investigation illustrate that expendable molds with smooth surfaces can be obtained with a graphitic material containing 4 to 10 per cent starch, when molding pressures of 30 to 120 psi are

TABLE 6 — THERMAL CONDUCTIVITY OF GRAPHITIC SPECIMENS

Starch, %	*Thermal Conductivity* (Btu/hr ft° F) at Molding Pressure (psi) 30	60	120
4	15.1	19.5	19.5
6	19.4	22.3	23.4
8	19.8	21.6	24.7
10	17.6	19.9	22.7

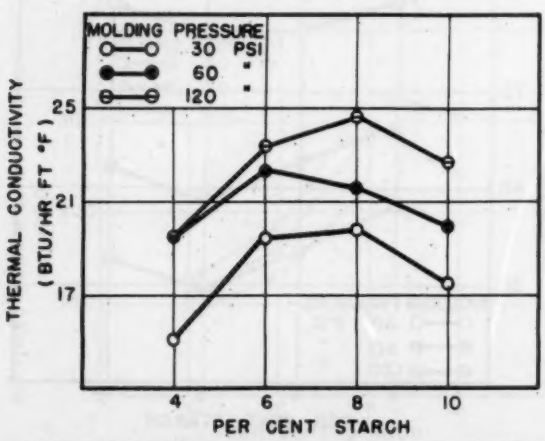


Fig. 8 — Thermal conductivity vs. per cent starch for graphitic mold material.

TABLE 7 — SHRINKAGE OF *GRAPHITIC SPECIMENS*

Starch, %	Shrinkage (in./ft) at Molding Pressure (psi) 30	60	120
4	0.24	0.22	0.16
6	0.25	0.23	0.20
8	0.31	0.26	0.22
10	0.34	0.29	0.25

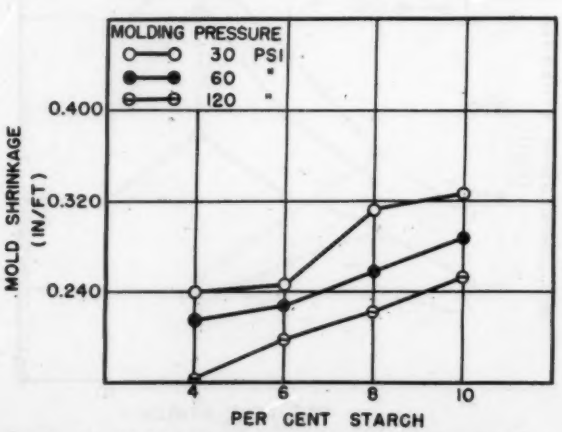


Fig. 9 — Mold shrinkage vs. per cent starch for graphitic mold material.

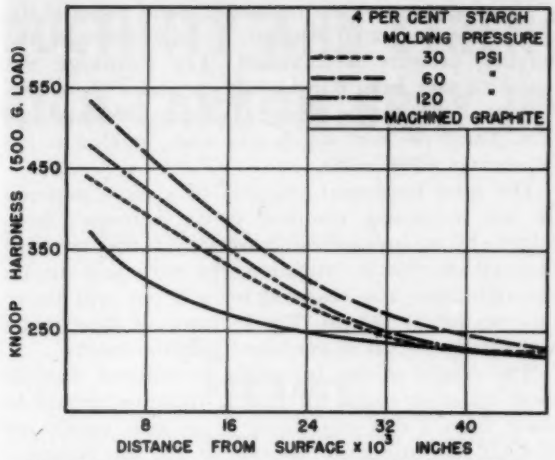


Fig. 10 — Knoop hardness vs. distance into metal from mold-metal surface, 4 per cent starch material.

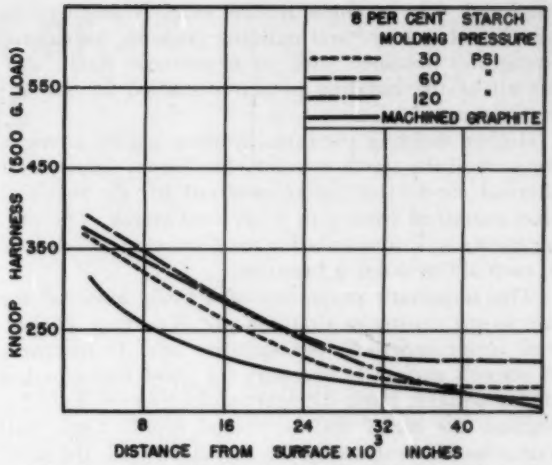


Fig. 12 — Knoop hardness vs. distance into metal from mold-metal surface, 8 per cent starch material.

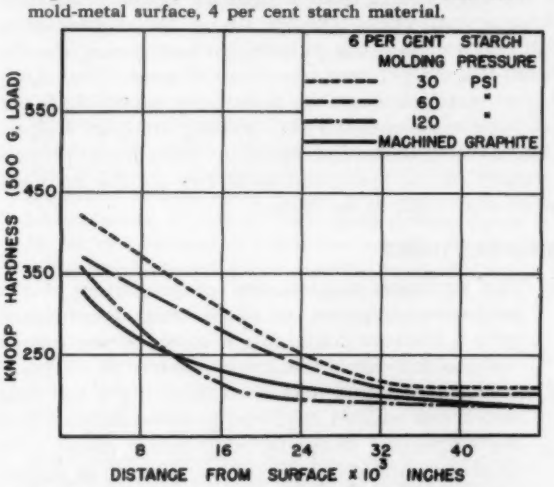


Fig. 11 — Knoop hardness vs. distance into metal from mold-metal surface, 6 per cent starch material.

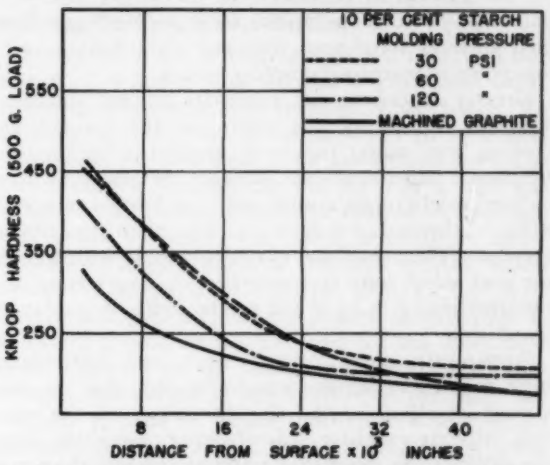


Fig. 13 — Knoop hardness vs. distance into metal from mold-metal surface, 10 per cent starch material.

TABLE 8 — KNOOP HARDNESS (500-g LOAD) VS. DISTANCE FROM MOLD-METAL INTERFACE FOR GRAPHITIC SPECIMENS OF VARIOUS STARCH CONTENTS AND MOLDING PRESSURES

Distance from Interface (in.)	Knoop Hardness* (500g Load)											
	4 Per Cent Starch and Molding Pressure (psi)			6 Per Cent Starch and Molding Pressure (psi)			8 Per Cent Starch and Molding Pressure (psi)			10 Per Cent Starch and Molding Pressure (psi)		
	30	60	120	30	60	120	30	60	120	30	60	120
0.002	—	—	—	358	315	224	322	353	275	472	462	355
0.004	410	411	469	405	344	337	333	335	333	428	425	355
0.006	420	494	430	378	339	305	346	346	337	386	386	358
0.008	380	465	390	381	351	285	351	367	339	360	348	326
0.010	386	438	428	353	329	260	301	358	337	341	381	315
0.012	346	465	476	339	313	236	271	335	339	344	322	271
0.014	350	408	438	320	322	226	297	299	320	326	370	253
0.018	344	396	311	313	277	218	270	271	285	330	353	226
0.022	307	415	303	307	261	222	240	255	238	253	344	233
0.026	317	341	293	289	242	226	244	248	231	256	270	214
0.030	320	287	261	279	207	200	230	219	256	281	247	240
0.034	234	261	263	233	196	197	182	233	199	251	202	211
0.038	227	258	279	208	200	160	173	204	175	218	193	192
0.042	228	283	221	211	184	177	174	215	172	210	184	203
0.046	228	255	216	208	208	170	203	229	160	209	175	205
0.250	228	210	218	214	210	175	160	220	160	196	203	211
0.254	—	—	215	218	201	192	176	217	150	205	218	212

*Average of two tests

Note: Hardness data for machined graphite is the average of three readings taken from reference 3.

employed. The strength of the mold is sensitive to both starch content and molding pressure. Maximum strength is obtained with an 8 per cent starch mix for all of the molding pressures studied in this investigation.

Higher molding pressures produce higher strength for each of the starch contents. Maximum density and thermal conductivity were observed for the material that contained from 6 to 8 per cent starch. The permeability was minimum for an 8 per cent starch mix at each of the molding pressures.

The important properties of a mold material for use in the casting of titanium are 1) surface, 2) thermal conductivity, 3) permeability and 4) inertness. A smooth surface is necessary for good reproduction of the pattern. High thermal conductivity is desirable because the metal will be cooled rapidly, and thus minimize the reaction of the cast metal with the mold material. However, the conductivity can be too high for the amount of super-heat in the metal, and produce rippled surfaces on the resulting castings. Rippled surfaces have been observed on titanium castings made in machined graphite molds.

Greater super-heat can eliminate rippled surfaces; however, because of the nature of the arc-melting process, large power inputs are needed to obtain this additional super-heat. An alternate method that may be used to eliminate ripples is to use a mold material with a thermal conductivity less than machined graphite. The rammable graphitic molds with 6 to 8 per cent starch have conductivities that vary from approximately ¼ to ⅓ of the conductivity of machined graphite.

Apparently, these conductivities are sufficiently high to prevent serious contamination, and yet low enough to eliminate the rippled surface on the castings without excessive super-heating. Since the conductivity of the rammable material is lower than machined graphite or copper, it is entirely feasible to incorporate chills of these materials in the mold to encourage directional solidification.

The main reason for controlling permeability in the mold material is to prevent penetration into the interstices of the mold by the molten metal. In a previous investigation this phenomenon was observed at the edges of molds with higher permeabilities.[3] In the current investigation, however, no edge penetration was observed. This was due first to the finer graphite that was used, and second to the reduced starch content.

Any material that contains constituents which volatilize on firing usually suffers shrinkage. Since the rammable graphitic material does contain an appreciable amount of volatiles, the problem of shrinkage is especially important. The amount of distortion and possible cracking brought about by shrinkage can be excessive in molds where the section size varies. It is therefore necessary that the shrinkage be held to an absolute minimum. In addition to being small, the shrinkage should be consistent to facilitate accurate pattern design.

The shrinkage of the rammable mold material decreases as the starch content is decreased and the molding pressure is increased. The shrinkage was found to vary from 0.16 to 0.34 in./ft for the starch contents and molding pressures used in this investigation. The 6 per cent starch mix when molded at 120 psi shrinks 0.198 in./ft.

The most important property of a mold material for use in casting titanium is its inertness (determined by microhardness traversal of the resulting castings). In this investigation, the minimum surface contamination was observed for a 6 per cent starch mix molded at 120 psi. The inertness of these molds was similar to that of machined graphite molds.

The results of this investigation indicate that the most desirable molds for casting titanium should be made from a mix containing 6 per cent starch and molded at a minimum pressure of 120 psi. However, it has been found from experience at the Frankford Arsenal that when complex shapes are molded it is difficult to draw the pattern without damaging the mold if the 6 per cent starch mix is used. Therefore, higher starch contents up to but not exceeding 8 per cent are recommended for molding complex shapes. The increase in starch content increases the as pressed strength of the mold and facilitates pattern removal without damage to the mold.

CONCLUSIONS

1) The optimum combination of properties of the rammable graphitic mold material is obtained with a mixture containing 6 to 8 per cent starch and molded with a minimum pressure of 120 psi.
2) A mixture of this material containing 6 per cent starch and molded at 120 psi provides molds which are equivalent to machined graphite molds with respect to minimizing depth and degree of surface contamination of resulting titanium castings.
3) The molding of complex shapes requires starch contents higher than 6 per cent, in order to facilitate pattern removal. Starch contents up to but not exceeding 8 per cent may be used to minimize pattern removal difficulties.
4) Maximum compression strength, density and thermal conductivity are obtained with molds made from an 8 per cent starch mix.
5) Shrinkage of rammed graphite molds is directly proportional to starch content and inversely proportional to molding pressure.
6) Mold surface finish is good and insensitive to starch content in the range of 4 to 10 per cent starch.

REFERENCES

1. Feild, A. L., Jr., "Expendable Molds for Titanium Castings," *Metal Progress, vol. 70, No. 4,* pp. 92-96, Oct. 1956.
2. Feild, A. L., Jr., and Edelman, R. E., "The Use of Expendable Graphite Molds in the Production of Sound Ductile Titanium Castings," AFS TRANSACTIONS, *vol. 65,* p. 517, 1957.
3. Antes, H. W., Norton, J. T., and Edelman, R. E., "Foundry Characteristics of a Rammed Graphitic Mold Material for Casting Titanium," AFS TRANSACTIONS, *vol. 66,* p. 135, 1958.

DUCTILITY AND STRENGTH OF HIGH-CARBON GRAY IRONS

By E. M. Stein and H. O. McIntire

ABSTRACT

Composition, microstructure and inoculation effects on ductility and strength of high-carbon (3.5-4.5 per cent) cast irons were evaluated. Irons melted by cupola, induction and blast furnace were included in the study. Some of the cupola and induction-melted irons were inoculated with ferrosilicon or with proprietary inoculating agents.

The criterion of ductility was the total elongation (or total strain) reached before fracture. Strain was measured with SR-4 gages cemented to standard tensile specimens. Ductility decreased with increasing manganese contents above 0.5 per cent. Inoculation of low-manganese (0.3-0.5 per cent) irons produced up to 50 per cent increase in elongation, but the effect of inoculation decreased with increasing manganese content. Ductility was not affected significantly by a wide variation in silicon content and the associated variation in microstructure.

For induction-melted and cupola irons, a relatively close relationship was found between carbon equivalent and strength, but the strength of the blast-furnace irons was considerably higher than this relationship would suggest. Increasing the Mn content above 0.5 per cent markedly increased the tensile strength through the associated increase of pearlite. Tensile strength was not affected by inoculation beyond the normal decrease from the increased silicon. There was a reasonably close correlation between hardness and strength for all of the irons.

INTRODUCTION

Many investigations dealing with properties of gray irons have been reported. Most of this work has involved gray irons with less than 3.5 per cent total carbon. This is understandable, because the bulk of the gray iron tonnage falls below this carbon level. However, higher carbon irons are desirable for many applications, the most notable being for heat resistance.

It was the purpose of this study to determine some of the factors (composition, inoculation and microstructure) that affect ductility and strength of irons having carbon contents from 3.5 to about 4.5 per cent.

Gray irons have so little ductility that this property often is ignored, but in applications where failure by cracking is encountered even a small increase in ductility can be important. For this reason, SR-4 strain gages were used in this study for obtaining accurate measurement of elongation.

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The study included 36 heats made by induction melting, 28 separate casts of cupola iron and 4 casts of blast-furnace iron.

PROCEDURE

Melting, Casting, and Sampling

Thirty-five heats of high-carbon cast iron, and one heat of a low-carbon iron, were melted in a basic-lined, high-frequency induction furnace. The charge for each high-carbon heat consisted of 83 lb of a low-phosphorus pig iron and 15 lb low-carbon steel. For the low-carbon heat, the ratio of pig iron to steel was reduced. Additions to obtain the desired composition were made at melt-down, using ferrosilicon, ferromanganese, ferrophosphorus and iron pyrites. When When the temperature reached 2700 F, the heats were tapped into a ladle and poured at 2550-2600 F. The same procedure was followed for the inoculated heats, except that the inoculants were added and stirred when the metal temperature in the ladle had dropped to 2600 F. Three inoculants were used in this investigation. One of these, designated "A", was the 75 per cent grade of ferrosilicon. Inoculants "B" and "C" are proprietary alloys, and their compositions are given in Table 1.

Y-blocks 3 in. thick were cast from each heat in open-top dry sand molds. Figure 1 shows the Y-block dimensions, which were the same as 3 in. A.S.T.M.[1] nodular iron Y-blocks, except for the over-all height, and the height of the test section proper. These dimensions were 7½ and 4½ in. as compared with 5⅝ and 2⅝ in. respectively, for the A.S.T.M. blocks. Immediately after filling the molds, they were covered with dry sand, and the castings were allowed to cool in the molds overnight. Y-blocks of the same dimensions were cast with cupola and blast-furnace iron, and these also were cooled in the molds overnight.

Duplicate tensile specimens were cut from symmetrical locations at the bottom of the Y-blocks, according to the drawing in Fig. 2, which also shows the dimensions of the tensile specimens. These were standard A.S.T.M. Type B specimens, but with a somewhat longer reduced section than is customarily used. This longer section provided adequate space for installation of strain gages and connection of the wires.

Brinell hardnesses were obtained for each set of duplicate bars prior to machining. The hardness impressions were located approximately 1 in. from an as-cast surface.

Mechanical Testing

The method for measuring strain was similar to that used by Flinn and Chapin.[2] Two SR-4 strain gages (Type A-1) were cemented 180 degrees apart on each bar. These gages were connected in series, and were balanced by two temperature-compensating gages, also connected in series. An SR-4 strain indicator was used to measure strains during loading, which was performed at a constant crosshead speed of 0.005 in. per min. The strain was followed to the point of fracture, with readings taken at load increments of 1000 lb. The strain immediately preceding fracture is referred to in this paper as elongation* and is the total elongation, which consists of plastic plus elastic strain. After fracture, one bar from each duplicate set was sectioned at the gage length and examined metallographically.

*Elongation and ductility are used interchangeably.

RESULTS

The data obtained in this study are listed in Tables 1 and 2, the former dealing with induction-melted irons, and the latter with cupola and blast-furnace irons. The induction-melted irons are grouped according to the chemical or inoculation variables. For example, Heats 2-6 were nominally of the same composition except for silicon contents. In several instances in Table 1 the same heat is listed more than once, if it fits into a progressive change in one of the

TABLE 1 — PROPERTIES OF INDUCTION-MELTED IRONS

Heat No.	Major Variable	% T.C.	% Si	% Mn	% P	% S	% Si Added as inoculant*	Tensile Strength, 1000 psi First Bar	Tensile Strength, 1000 psi Second Bar	Tensile Strength, 1000 psi Ave.	Elongation, % First Bar	Elongation, % Second Bar	Elongation, % Ave.	Bhn, 3000 kg	% Ferrite in Matrix	Carbon Equivalent, T.C.+⅓ (Si+P)
1	Low carbon	3.00	1.96	0.54	0.14	0.04	—	29.0	28.6	28.8	0.77	0.86	0.82	165	10	3.70
2	Silicon	3.53	1.07	0.52	0.11	0.04	—	26.2	25.8	26.0	1.15	0.10	1.12	159	Trace	3.93
3		3.74	1.56	0.53	0.13	0.03	—	17.7	17.0	17.3	0.94	1.21	1.08	126	25	4.30
4		3.67	1.96	0.51	0.14	0.03	—	10.6	10.6	10.6	0.93	0.90	0.92	101	45	4.37
5		3.74	2.02	0.55	0.13	0.03	—	8.7	8.4	8.5	1.01	0.83	0.92	105	65	4.45
6		3.57	2.44	0.55	0.13	0.03	—	8.9	9.0	9.0	0.66	1.23	0.94	86	85	4.42
7	Silicon	3.74	1.11	0.31	0.13	0.03	—	19.6	19.9	19.8	0.87	1.04	0.95	154	10	4.15
8		3.67	1.43	0.31	0.13	0.03	—	20.3	21.1	20.7	0.88	0.90	0.89	155	20	4.19
9	Silicon	3.68	1.29	0.62	0.14	0.08	—	20.1	20.0	20.1	0.91	0.88	0.90	137	10	4.16
10		3.72	1.84	0.64	0.14	0.09	—	13.0	12.7	12.8	0.88	0.91	0.90	109	30	4.38
3	Manganese	3.74	1.56	0.53	0.13	0.03	—	17.7	17.0	17.3	0.94	1.21	1.08	126	25	4.30
11		3.64	1.51	0.77	0.13	0.03	—	22.2	22.1	22.1	0.95	0.92	0.93	146	Trace	4.18
12		3.68	1.58	1.08	0.13	0.03	—	20.9	21.0	21.0	0.77	0.70	0.74	146	0	4.25
13		3.92	1.51	1.34	0.13	0.03	—	18.0	17.2	17.6	0.78	0.60	0.69	140	0	4.46
14	Phosphorus	3.69	1.61	0.79	0.04	0.03	—	17.4	18.3	17.9	0.78	0.82	0.80	128	10	4.24
11		3.64	1.51	0.77	0.13	0.03	—	22.2	22.1	22.1	0.95	0.92	0.93	146	Trace	4.18
15		3.70	1.58	0.80	0.19	0.03	—	20.8	20.8	20.8	0.89	0.82	0.85	143	5	4.29
11	Sulfur	3.64	1.51	0.77	0.13	0.03	—	22.2	22.1	22.1	0.95	0.92	0.93	146	Trace	4.18
16		3.74	1.55	0.80	0.13	0.08	—	19.7	19.6	19.7	0.91	0.90	0.91	143	Trace	4.30
17		3.72	1.59	0.78	0.13	0.13	—	16.8	16.8	16.8	0.85	0.85	0.85	126	10	4.29
18	Kind of	3.69	1.47	0.51	0.13	0.03	0.20 A	18.1	18.0	18.1	1.30	1.29	1.30	126	15	4.22
19	inoculant	3.55	1.45	0.50	0.13	0.03	0.20 B	21.0	21.2	21.1	1.33	1.32	1.32	126	15	4.07
20		3.73	1.52	0.52	0.13	0.03	0.20 C	15.8	15.6	15.7	1.44	1.33	1.38	111	35	4.28
7	Inoculation	3.74	1.11	0.31	0.13	0.03	—	19.6	19.9	19.8	0.87	1.04	0.95	154	10	4.15
21		3.73	1.10	0.32	0.13	0.03	0.20 C	20.8	20.8	20.8	1.37	1.42	1.39	137	20	4.14
8		3.67	1.43	0.31	0.13	0.03	—	20.3	21.1	20.7	0.88	0.90	0.89	155	20	4.19
22		3.73	1.38	0.32	0.13	0.03	0.20 C	17.6	17.5	17.6	1.39	1.29	1.34	126	30	4.23
2	Amount of	3.53	1.07	0.52	0.11	0.04	—	26.2	25.8	26.0	1.15	1.10	1.12	159	Trace	3.93
23	inoculant	3.72	1.06	0.51	0.13	0.03	0.05 C	20.8	21.0	20.9	1.30	1.28	1.29	143	10	4.11
24		3.72	0.98	0.55	0.13	0.03	0.10 C	21.8	21.8	21.8	1.03	0.96	0.99	143	5	4.09
25		3.65	0.94	0.53	0.13	0.03	0.20 C	24.5	24.6	24.6	1.09	1.11	1.10	163	5	4.00
26		3.71	1.03	0.53	0.13	0.03	0.40 C	23.4	22.8	23.1	1.47	1.33	1.40	156	5	4.09
27	Amount of	3.78	1.29	0.51	0.13	0.03	0.05 C	18.0	17.9	18.0	1.15	1.13	1.14	131	20	4.25
28	inoculant	3.68	1.41	0.52	0.13	0.03	0.10 C	18.6	18.4	18.5	1.14	1.20	1.17	136	20	4.19
29		3.71	1.27	0.52	0.13	0.03	0.20 C	20.6	20.4	20.5	1.24	1.22	1.23	142	20	4.17
30		3.72	1.35	0.50	0.13	0.03	0.40 C	17.8	17.8	17.8	1.39	1.30	1.35	136	30	4.21
11	Amount of	3.64	1.51	0.77	0.13	0.03	—	22.2	22.1	22.1	0.95	0.92	0.93	146	Trace	4.18
31	inoculant	3.61	1.50	0.77	0.13	0.03	0.05 C	22.2	22.1	22.2	0.98	1.00	0.99	156	5	4.15
32		3.65	1.50	0.77	0.13	0.03	0.10 C	21.2	21.4	21.3	1.06	1.05	1.05	143	5	4.19
33		3.71	1.46	0.76	0.13	0.03	0.20 C	18.6	18.6	18.6	1.07	1.07	1.07	143	10	4.24
34		3.67	1.47	0.79	0.13	0.03	0.40 C	20.8	20.8	20.8	1.14	1.08	1.11	143	10	4.20
35	Sulfur	3.73	1.46	0.67	0.12	0.02	0.20 C	16.3	16.7	16.5	0.88	1.13	1.01	131	20	4.26
36		3.76	1.43	0.68	0.14	0.08	0.20 C	17.4	17.4	17.4	1.10	1.08	1.09	140	25	4.29

*A—Ferrosilicon containing 75 per cent silicon.
B—Proprietary alloy containing 45-50 per cent Si, 9-11 per cent Ti, and 5-8 per cent Cr.
C—Proprietary alloy containing 60-65 per cent Si, 5-7 per cent Mn, and 5-7 per cent Zr.

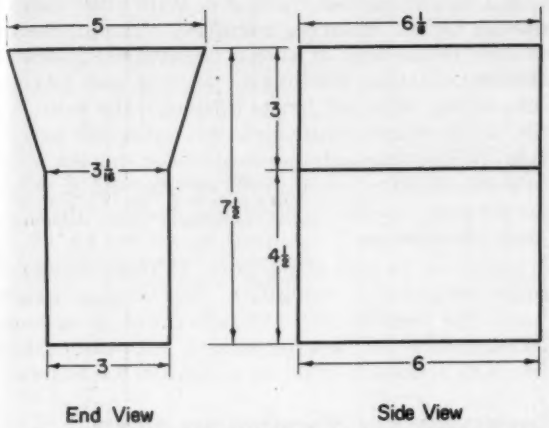


Fig. 1 — Dimensions of Y-block in in., which are the same as 3 in. A.S.T.M. nodular iron Y-blocks, except for the over-all height and height of test section proper.

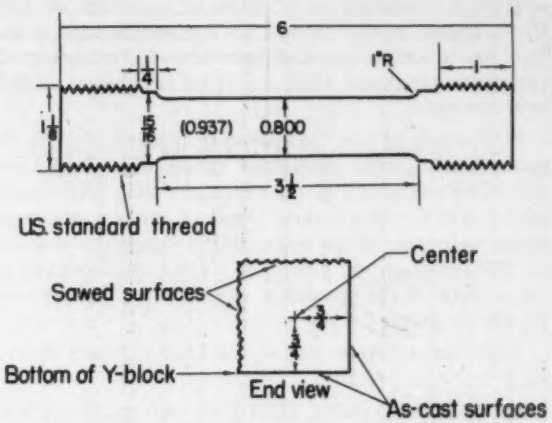


Fig. 2 — Location and dimension in in. of tensile bars cut from symmetrical locations at the bottom of the Y-blocks.

elements or a change in type or amount of inoculant. Table 2 lists the uninoculated and inoculated cupola irons and blast-furnace irons, each group being listed according to increasing carbon equivalent.

Composition and Inoculation Effects on Ductility

Of the variables evaluated, only manganese content and inoculation appeared to have important effects on ductility. The effects of manganese level and of inoculation are illustrated for induction-melted irons in Fig. 3. As the manganese content was increased above 0.5 per cent, there was a tendency for the elongation values to decrease. This was true for both inoculated and uninoculated irons. Particularly at manganese levels of 0.5 per cent or less, inoculation markedly increased the elongation. In the irons containing approximately 0.8 per cent manganese and over minute particles of massive cementite were detected, and it is believed that these particles may account for the decrease in elongation. Figure 4 shows one of these cementite particles, and it will be noted that two of its branches extend along pearlite grain boundaries. A hard, brittle constituent such as this would be expected to reduce ductility.

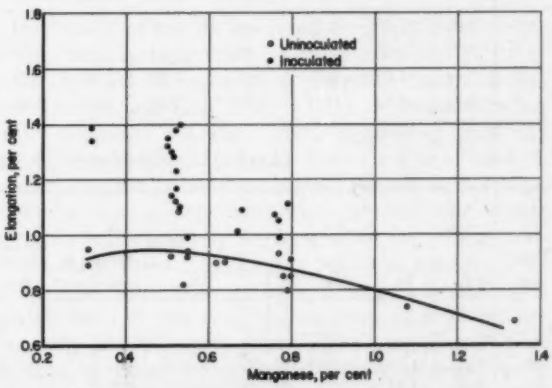


Fig. 3 — Manganese and inoculation effects on elongation for induction-melted irons.

All of the cupola irons had relatively high manganese contents, ranging from 0.64 per cent to 1.00 per cent. Also, there are only seven inoculated irons available for comparison. Nevertheless, by grouping the irons into two ranges of manganese contents with and without inoculation in Table 3, the effects of these variables become apparent. The results indicate that decreasing the manganese content from an average of about 0.88 per cent to 0.75 per cent increased the elongation from about 0.93 to 1.04 per cent. Also, the inoculated irons had slightly higher elongations than the uninoculated irons. Considering the results obtained for induction-melted iron, it seems likely that ductility of the cupola irons would have been higher at lower manganese levels and that the effect of inoculation would have been more pronounced.

The effect of the amount of inoculant on ductility is illustrated for induction-melted irons in Fig. 5. For manganese levels of about 0.5 and 0.8 per cent, increasing the amount of inoculant (in this instance in-



Fig. 4 — Particle of massive cementite in pearlitic matrix from heat 11 (0.77 per cent Mn). Picral etch. 500 ×.

oculant C) resulted in progressive increases in ductility. Figure 5 also shows that the effect of inoculation was more pronounced for the iron containing 0.3 per cent manganese than for irons of higher manganese contents.

Elongation of the blast-furnace irons tended to be quite erratic, with individual values of 0.47-1.22 per cent. The coarseness of the graphite flakes probably is responsible for this erratic behavior, since a few large flakes oriented in an unfavorable direction (normal to the direction of principal stress), particularly at the surface of the specimen, could lower the capacity for deformation.

The silicon content was varied from 1.07 per cent in heat 2, to 2.44 per cent in heat 6. With other compositional factors remaining essentially constant, the increased percentages of silicon resulted in increased amounts of ferrite, reaching 85 per cent with 2.44 per cent silicon. Although ferrite by itself is the most ductile of the micro-constituents, increasing the percentage of ferrite through increased silicon did not result in increased ductility. Instead, the high-silicon, high-ferrite irons tended to have less ductility than the lower silicon irons.

Little can be said of the effect of phosphorus and sulfur contents on elongation. The ranges covered were rather narrow (0.04-0.19 per cent of phosphorus, and 0.02-0.09 per cent of sulfur), but within these limits no noticeable effect on elongation was observed.

TABLE 3 — MANGANESE CONTENT AND INOCULATION EFFECTS ON CUPOLA IRONS ELONGATION

No. of	Manganese Range, %				Elongation, %		
Tests	Min.	Max.	Ave.	Inoculation	Min.	Max.	Ave.
11	0.64	0.80	0.75	no	0.83	1.22	1.04
3	0.77	0.80	0.78	yes	1.04	1.14	1.09
9	0.81	1.00	0.88	no	0.82	1.00	0.93
4	0.81	0.89	0.84	yes	0.76	1.09	1.01

Composition and Microstructure Effects on Tensile Strength

The relationship between carbon equivalent, having the customary formula of Total Carbon + 1/3 (Si + P), and tensile strength for the three groups of irons, is shown in Fig. 6. A reasonably narrow band encompasses all but four of the induction-melted and cupola irons, and includes even the low-carbon iron

TABLE 2 — PROPERTIES OF CUPOLA AND BLAST-FURNACE IRONS

Heat	%					% Si Added as	Tensile Strength, 1000 psi			Elongation, %			Bhn, 3000	% Ferrite in	Carbon Equivalent,
No.	T.C.	Si	Mn	P	S	Inoculant*	First Bar	Second Bar	Ave.	First Bar	Second Bar	Ave.	kg	Matrix	T.C.+1/3(Si+P)
						Cupola Irons									
37	3.63	1.62	0.75	0.12	0.04	—	18.2	18.8	18.5	1.00	1.16	1.08	128	10	4.21
38	3.64	1.63	0.70	0.13	0.03	—	19.6	19.5	19.5	1.04	1.20	1.12	128	10	4.22
39	3.74	1.33	0.73	0.13	0.04	—	18.8	18.8	18.8	1.06	1.02	1.04	130	Trace	4.22
40	3.63	1.75	0.77	0.14	0.03	—	18.2	18.8	18.5	0.83	0.83	0.83	127	Trace	4.26
41	3.66	1.67	0.78	0.12	0.03	—	16.9	17.2	17.1	1.12	1.21	1.17	130	10	4.26
42	3.71	1.62	0.64	0.13	0.03	—	16.1	15.8	16.0	1.11	1.00	1.06	120	20	4.29
43	3.71	1.79	0.79	0.14	0.04	—	16.3	16.8	16.6	0.99	0.97	0.98	132	10	4.36
44	3.77	1.65	0.93	0.14	0.02	—	16.6	16.2	16.4	0.94	0.81	0.87	131	5	4.37
45	3.82	1.53	0.83	0.14	0.02	—	15.3	15.0	15.1	0.90	1.06	0.98	130	10	4.38
46	3.79	1.62	0.76	0.17	0.03	—	16.8	16.6	16.7	0.91	0.94	0.92	131	5	4.39
47	3.90	1.41	0.82	0.15	0.03	—	11.6	11.6	11.6	0.98	1.00	0.99	126	5	4.42
48	3.87	1.53	0.92	0.14	0.02	—	11.8	11.7	11.8	0.95	0.96	0.95	120	10	4.43
49	3.86	1.60	0.90	0.17	0.03	—	12.3	12.4	12.3	1.00	1.00	1.00	107	10	4.45
50	3.84	1.77	1.00	0.14	0.02	—	12.4	12.4	12.4	0.95	0.90	0.93	121	10	4.48
51	3.68	1.61	0.81	0.13	0.04	—	17.0	17.0	17.0	1.08	0.89	0.99	136	10	4.26
52	3.68	1.82	0.81	0.13	0.04	0.20 A	15.6	14.9	15.2	1.19	1.01	1.10	121	15	4.33
53	3.70	1.78	0.77	0.14	0.04	—	16.4	16.4	16.4	1.00	0.99	0.99	121	15	4.34
54	3.70	1.91	0.77	0.14	0.04	0.20 C	13.7	13.5	13.6	1.16	1.12	1.14	108	25	4.39
55	3.72	1.76	0.77	0.13	0.04	—	15.4	14.9	15.1	1.12	1.05	1.08	116	20	4.35
56	3.72	1.96	0.77	0.13	0.04	0.20 A	12.0	12.0	12.0	1.09	1.07	1.08	109	30	4.41
57	3.78	1.64	0.80	0.14	0.04	—	15.4	14.8	15.1	1.20	1.24	1.22	120	15	4.38
58	3.70	1.86	0.80	0.13	0.04	0.20 C	12.5	12.1	12.3	1.07	1.01	1.04	106	25	4.36
59	3.75	1.79	0.89	0.13	0.04	—	12.8	12.6	12.7	0.92	0.86	0.89	111	15	4.39
60	3.75	1.89	0.89	0.13	0.04	0.20 C	11.7	11.6	11.7	1.05	1.14	1.09	110	20	4.42
61	3.79	1.78	0.80	0.13	0.02	—	12.2	12.0	12.1	1.02	1.10	1.06	111	20	4.42
62	3.77	1.93	0.82	0.13	0.02	0.20 C	10.2	9.8	10.0	1.04	1.13	1.08	108	30	4.45
63	3.80	1.94	0.84	0.14	0.04	—	13.1	12.6	12.8	0.83	0.81	0.82	116	30	4.50
64	3.80	2.03	0.84	0.14	0.04	0.20 C	10.1	10.5	10.3	0.54	0.97	0.76	109	35	4.53
						Blast-Furnace Irons									
65	4.39	1.82	1.22	0.13	0.03	—	9.6	10.5	10.0	0.68	0.88	0.78	104	30	5.04
66	4.47	1.65	1.07	0.16	0.02	—	9.2	10.4	9.8	0.94	0.72	0.83	117	15	5.07
67	4.32	2.28	1.20	0.14	0.02	—	11.4	11.6	11.5	1.22	0.82	1.02	101	50	5.13
68	4.44	2.19	1.25	0.14	0.02	—	9.2	9.6	9.4	0.47	0.62	0.54	96	40	5.22

*A—Ferrosilicon, containing 75 per cent silicon.
C—Proprietary alloy, containing 60-65 per cent Si, 5-7 per cent Mn, and 5-7 per cent Zr.

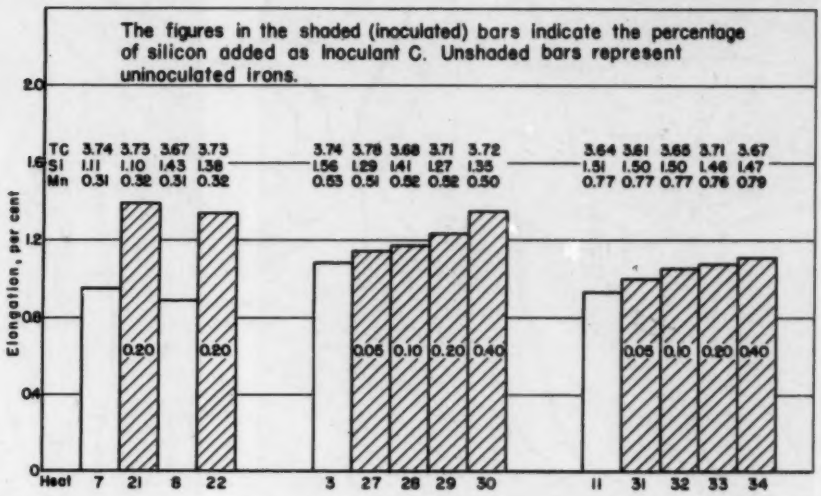


Fig. 5 — Inoculation effect on elongation of induction-melted irons at three manganese levels (0.3, 0.5 and 0.8 per cent). Increasing the amount of inoculant resulted in progressive increases in ductility.

with a carbon equivalent of 3.70. This relationship coincides closely with that shown by Schneidewind and McElwee[3] for 4-in. thick sections.

In Fig. 6, the points for the blast-furnace irons are displaced considerably from those representing induction-melted and cupola irons. The carbon equivalents for the four blast-furnace irons were extremely high (5.04-5.22), yet the tensile strengths obtained for them were at a level that was obtained with cupola and induction-melted irons having carbon equivalents of about 4.5. It is suggested, as a possible explanation, that despite the greater amount of graphitic carbon in the blast-furnace irons, they had larger graphite-free areas than the cupola or induction-melted irons of similar strength.

Figures 7-10 show typical microstructures of low-strength irons melted by the three different methods. The largest circles that can be inscribed in graphite-free areas in the induction-melted and cupola irons (Figs. 7 and 8, respectively), are much smaller than those that can be inscribed in graphite-free areas in the blast-furnace irons (Figs. 9 and 10). Furthermore, because of some thick graphite flakes in the blast-furnace irons, it is likely that the remainder actually possessed a lower carbon equivalent than the induction-melted and cupola irons of Figs. 7 and 8.

With essentially constant carbon content and a manganese level of approximately 0.5 per cent, the tensile strength was drastically reduced when the silicon content was increased from 1.07 in heat 2 to 2.02 per cent in heat 5. The tensile strength decreased from 26,000 psi for the 1.07 per cent silicon iron, to 8500 psi for the iron of 2.02 per cent silicon content. These changes in silicon content and tensile strength were accompanied by an increase in free ferrite, from a trace in the lower silicon iron to 65 per cent in the higher silicon iron. A further increase in the silicon content to 2.44 per cent (heat 6), increased the amount of ferrite to 85 per cent, but the tensile strength increased slightly to 9000 psi. This slight in-

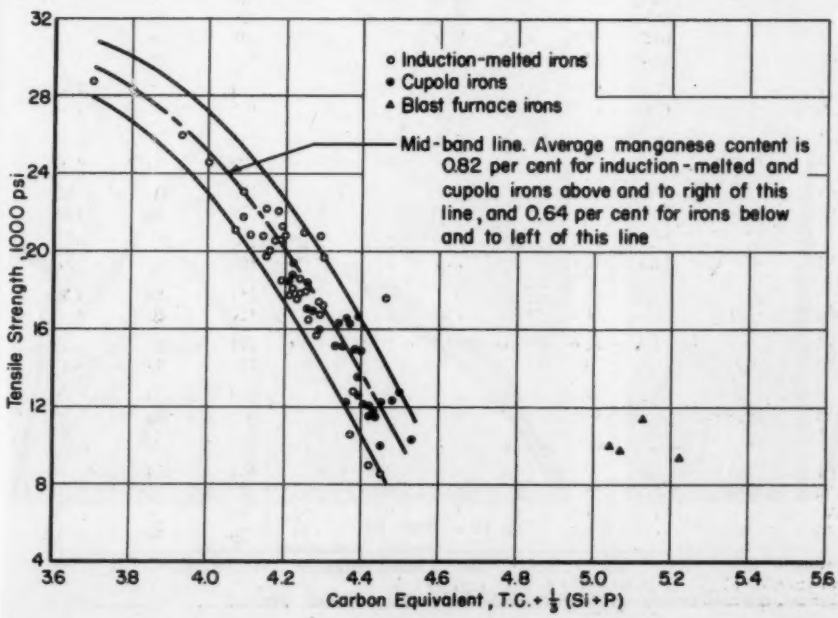


Fig. 6 — The relationship between carbon equivalent and tensile strength for the three groups of irons.



Fig. 7 — Typical microstructure of low-strength induction-melted iron, from heat 5. Tensile strength—8400 psi, carbon equivalent—4.45. Picral etch. 100 ×.

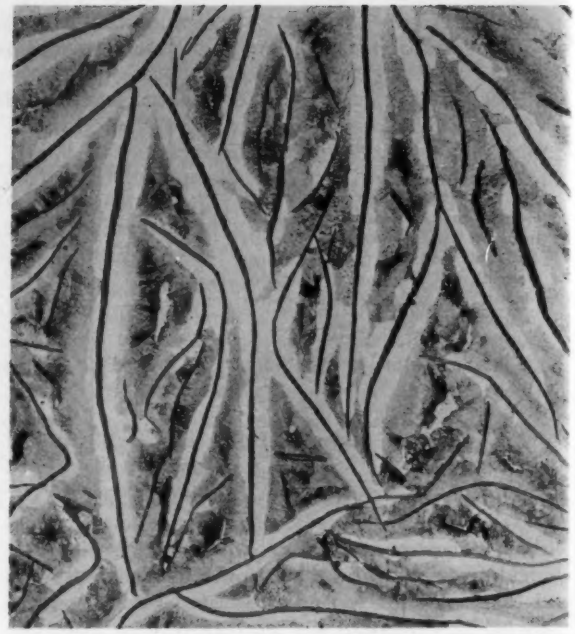


Fig. 8 — Typical microstructure for low-strength cupola iron, from heat 62. Tensile strength—10,200 psi; carbon equivalent—4.45. Picral etch. 100 ×.

crease in strength above that for the 2.02 per cent silicon iron is explained by the decrease in carbon equivalent as the result of a lower carbon content.

An indication of the effect of manganese content on tensile strength can be obtained by comparing heats 3, 11 and 12, all of which had similar total carbon (3.7 per cent) and silicon (1.55 per cent) contents. The tensile strengths were 17,300 psi for the iron having 0.53 per cent manganese, and 22,100 psi for the iron of 0.77 per cent manganese content. This substantial increase in strength resulted mainly from the reduction in the amount of free ferrite, the former having 25 per cent ferrite as compared with only traces of ferrite in the latter.

Heat 12 with 1.08 per cent manganese had a tensile strength of 21,100 psi. This indicates that addition of

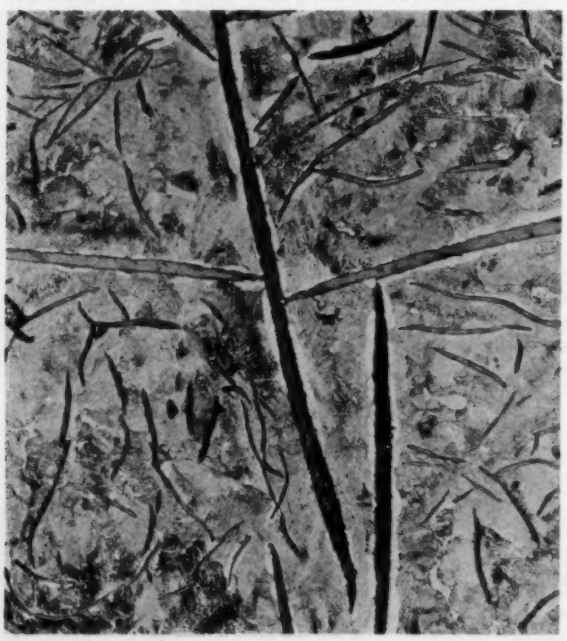


Fig. 9 — Heat 66
Carbon equivalent — 5.07
Tensile strength — 10,400 psi

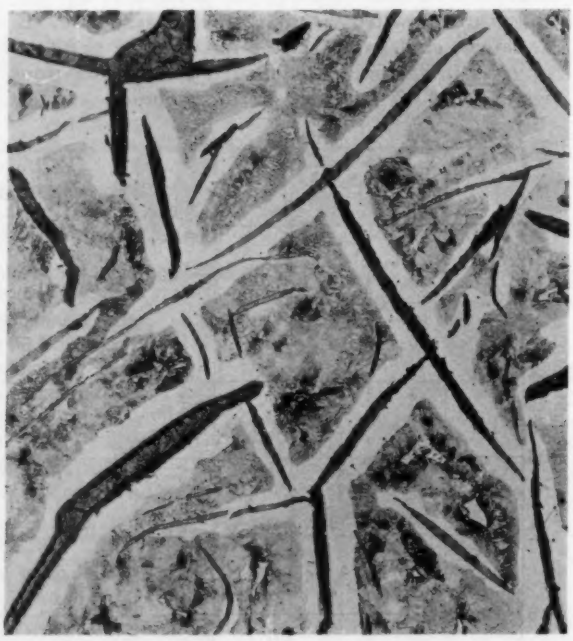


Fig. 10 — Heat 67
Carbon equivalent — 5.13
Tensile strength — 11,400 psi

Figs. 9-10 — Typical microstructure of blast-furnace irons. Picral etch. 100 ×.

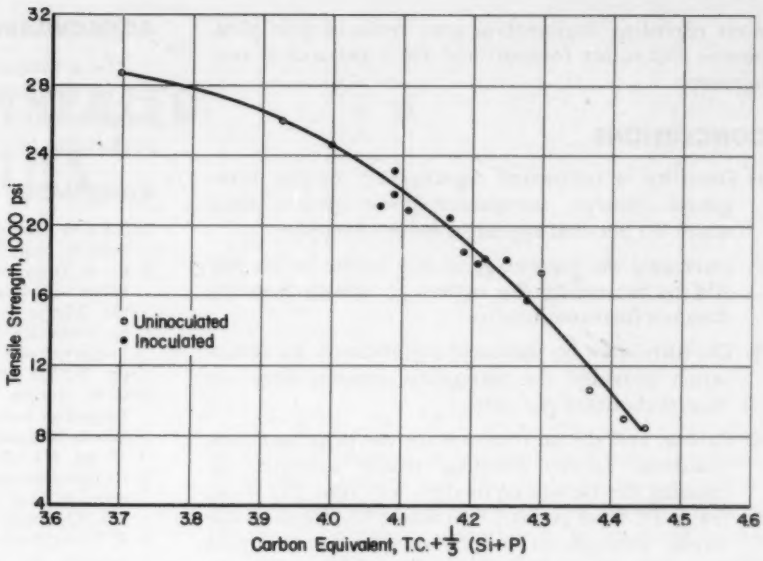


Fig. 11 — Graph showing the carbon equivalent-tensile strength relationship for all induction-melted irons containing 0.50-0.55 per cent manganese.

manganese beyond that required to give a fully pearlitic matrix does not result in increased strength. Heat 13 contained 1.34 per cent manganese, but it cannot be included conveniently in the comparison because it also had a relatively high carbon equivalent which would tend to offset the strengthening effect of the higher manganese. The effect of manganese in strengthening the iron also is apparent in Fig. 6. The irons above and to the right of the middle of the band averaged 0.82 per cent manganese. Those below and to the left of the middle of the band averaged 0.64 per cent manganese.

Figure 11 reveals that inoculation had no effect on tensile strength except through the normal effect of increased silicon. The data plotted in Fig. 11 represent induction-melted irons only, but the same was true for the cupola irons. This observation is in agreement with the conclusion reached by Burgess and Bishop[4] that, as the carbon equivalent approaches 4.0, ladle inoculation does not materially affect the tensile strength. In lower-carbon irons that tend to have dendritic arrangement of the graphite flakes, inoculation generally serves to reduce this dendritic tendency. However, the graphite in these higher carbon irons was essentially of random arrangement (Type A) without inoculation.

A straight-line relationship was found between hardness and tensile strength for the induction-melted, cupola, and blast-furnace irons, regardless of composition or treatment. A plot of hardness and tensile strength for the irons investigated is shown in Fig. 12. Approximately 95 per cent of the points were within a range of about ± 3000 psi from the mean. Only 3 points fell outside of this range, one point for each of the induction-melted, cupola, and blast-furnace irons. It is especially interesting to note that three of the four blast-furnace irons correlated well with the other irons in this respect.

Results of this study permit the following conclu-

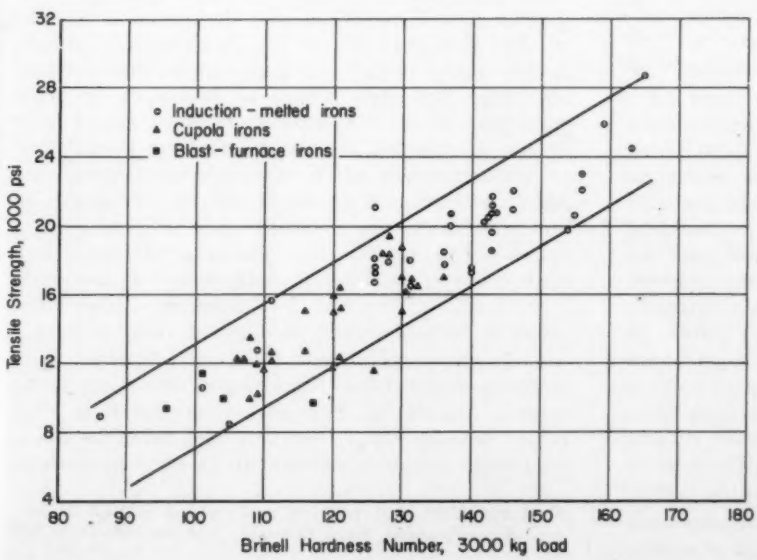


Fig. 12 — Graph showing Brinell hardness and tensile strength relationship for the irons investigated.

sions regarding high-carbon gray irons of low phosphorus and sulfur contents and for a particular cooling rate.

CONCLUSIONS

1) Ductility is influenced significantly by the manganese content; manganese levels greater than about 0.5 per cent appear to reduce ductility.

2) Increasing the percentage of free ferrite in the matrix by increasing the carbon or silicon contents does not increase ductility.

3) Ductility may be increased significantly by inoculation provided the manganese content does not exceed about 0.6 per cent.

4) Carbon and silicon contents are the principal compositional factors affecting tensile strength. Increasing the carbon equivalent, per cent TC + ⅓ (Si + P), from about 4.0 to about 4.5 decreases the tensile strength of induction-melted and cupola irons from about 25,000 psi to 10,000 psi.

5) Manganese additions result in higher strengths so long as they produce increased proportions of pearlite in the matrix.

6) Inoculation has no significant influence on the tensile strength except through the normal effect of the increased silicon content.

ACKNOWLEDGMENT

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REFERENCES

1. A.S.T.M. Designation: A339-55. Standard Specifications for Nodular Iron Castings.
2. R. A. Flinn, and H. J. Chapin, "Ductility and Elasticity of White and Gray Irons", AFS TRANSACTIONS, *vol. 54*, pp. 141-153 (1946).
3. R. Schneidewind, and R. G. McElwee, "Composition and Properties of Gray Iron", Part I, AFS TRANSACTIONS, *vol. 58*, pp. 312-323 (1950).
4. C. O. Burgess and R. W. Bishop, "Alterations in Cast Iron Properties Accompanying the Use of a Strong Inoculant of the Silicon-Manganese-Zirconium Type", AFS TRANSACTIONS, *vol. 52*, pp. 671-710 (1944).
5. R. Schneidewind and R. G. McElwee, "Correlation of Properties of Gray Irons", Part II, AFS TRANSACTIONS, *vol. 58*, pp. 323-332 (1950).
6. R. Schneidewind and R. G. McElwee, "The Influence of Composition on the Physical Properties of Gray Irons", AFS TRANSACTIONS, *vol. 47*, pp. 491-512 (1939).
7. T. E. Barlow, and C. H. Lorig, "Gray Cast Iron—Tensile Strength, Brinell Hardness, and Composition Relationships", AFS TRANSACTIONS, *vol. 54*, pp. 545-552 (1946).
8. V. A. Crosby, "Gray Cast Iron Section Size vs. Physical Properties", AFS TRANSACTIONS, *vol. 54*, pp. 791-794 (1946).
9. H. T. Angus, F. Dunn, and D. Marles, "Properties of Cast Iron in Relation to the Carbon Equivalent Value", AFS TRANSACTIONS, *vol. 57*, pp. 21-44 (1949).

DUCTILE IRON AS-CAST AND ANNEALED TENSILE PROPERTIES

By A. H. Rauch, J. B. Peck and E. M. McCullough

ABSTRACT

Data from production heats of ductile iron were studied, principally by means of regression analyses. Some of the conclusions resulting from these studies are presented in this paper.

Significant relationships exist between tensile properties and Brinell hardness in as-cast pearlitic-ferritic ductile iron. However, these relationships are strongly affected by changes in composition of the iron. Accurate prediction of the tensile properties of this material requires knowledge of both hardness and composition.

INTRODUCTION

In planning for the production of pearlitic-ferritic ductile iron to be used in the as-cast condition, an investigation was undertaken to determine the ranges of tensile properties that could be expected, and to establish whether or not control of the Brinell hardness of the castings would insure control of the tensile properties of the castings. Encouragement for this approach came from a report that a reasonable indication of the tensile strength of pearlitic-ferritic ductile iron could be obtained from a Brinell hardness reading through the use of an approximate tensile strength-Brinell hardness ratio of 420 to one.[1]

As a preliminary step, simple regression analyses were performed on Y block data from production heats of as-cast ductile iron purchased from a jobbing foundry. Although the 420 to one ratio could not be exactly confirmed, significant relationships, which could be expressed as simple regression equations, were found to exist between all tensile properties and Brinell hardness. However, analyses of covariance showed the elevation of the regression lines to be dependent on the chemical composition of the iron. Therefore, the objectives of the investigation were expanded to include determination of the degree of control of composition necessary to insure control of the tensile properties.

Further investigation was then conducted to determine more precisely, and over wider ranges of variables, the tensile properties-Brinell hardness relationships in as-cast ductile iron, and the effects of composition on these relationships. Data from the initial production heats at the authors' company were used in multiple regression analyses of tensile properties on Brinell hardness and chemical composition. In order to relate the findings from this investigation to the ranges of tensile properties to be expected in the castings, a method was developed to calculate the variation of tensile properties resulting from variation of Brinell hardness and composition within specified ranges.

Complete annealing of ductile iron castings of the same chemical composition as those used in the as-cast state has been contemplated. Accordingly, an investigation was undertaken to determine the tensile properties that could be expected, and to determine the factors that would have to be controlled to insure control of the tensile properties in castings of this material.

The tensile strength-Brinell hardness ratio of 420 to one has been reported to be also applicable to type 60-45-10 ductile iron,[1] which is similar to this annealed ductile iron. However, Schneidewind and Wilder[2] and Reynolds, Adams and Taylor[3] have reported experiments indicating that the tensile properties of annealed ductile iron are primarily dependent on chemical composition. Reynolds, Adams and Taylor present equations which are claimed to accurately predict the mechanical properties of annealed ductile iron from composition alone, and further state that the ratio of hardness to strength can be changed by changes in composition.[3]

In order to determine whether Brinell hardness measurements or chemical analysis represented the better method of control of tensile properties, simple regression analyses of tensile properties on Brinell hardness and multiple regression analyses of tensile properties on chemical composition were performed on data from production heats. It was found that chemical composition is the more accurate indication of tensile properties. The multiple regression analysis of tensile properties on chemical composition produced relationships which, when combined with the method for calculation of variation mentioned above, make possible the prediction of the tensile properties of fully annealed ductile iron castings having a composition within specified ranges.

The statistical techniques used in these investigations may not be familiar to some readers. The Appendix to this paper is devoted to the explanation of

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assumptions underlying these techniques and to proper interpretation of the results.

PROCEDURE

The data for the as-cast ductile iron investigation were obtained from 194 one-in. and three-in. Y-block castings poured from production heats over a three month period. These castings were dumped from the molds at varying intervals of time following solidification in order to obtain a wide distribution of cooling rates. The annealed ductile iron data were taken from 102 one-in. Y-block castings poured from production heats over a four day period. These castings were annealed with the equipment and cycle available for malleable iron, i.e., 6 hr to 1750 F, 10 hr at 1750 F, 5 hr from 1750 F to 1375 F, 11 hr from 1375 F to 1300 F, air cooled to room temperature.

Standard, threaded 0.505 in. test bars were machined from the lower 1 in. of the Y blocks for determination of tensile properties. Yield strengths were determined from stress-strain recordings using the 0.2 per cent strain offset method, and elongations were determined over 2 in. gage lengths. Test bars that fractured at or outside of the gage marks were not included in the investigations. Brinell hardness, microstructure and chemical composition, with the exception of the carbon content of the as-cast material, were determined on the portions of the Y blocks from which the test bars were machined. In the as-cast ductile iron investigation, carbon was determined from chilled carbon pins poured immediately after the Y blocks.

For the annealed ductile iron investigation, the carbon determination procedure was changed to analysis of wafers cut from the Y blocks, after it was discovered that the carbon content of a chilled pin does not always represent the carbon content of a Y block. This subject is covered in a paper referred to in the Appendix. The carbon content reported for as-cast ductile iron is the carbon content of the molten iron, and the carbon content reported for annealed ductile iron is the carbon content of the Y blocks.

The graphite structure in all of the castings included in these investigations was essentially nodular. Castings containing vermicular graphite were rejected. Up to 10 per cent crab graphite was tolerated, although the general level was nil. Figure 1 defines the laboratory's graphite classification. Up to 15 per cent pearlite was considered acceptable in the annealed ductile iron castings. The general level was much lower, nil to 2 per cent. No castings containing primary cementite were included, with the exception of several Y blocks containing traces of cell boundary carbide.

TABLE 1 — RANGES OF VARIABLES STUDIED

Variables	As-Cast Ductile Iron	Annealed Ductile Iron
Hardness, Bhn	179-293	146-174
Tensile strength, psi	74,500-132,500	56,300-69,500
Yield strength, psi	48,500-82,000	36,800-51,000
Elongation, %	3-15	20-28
Pearlite, %	15-97	Nil-15
C, %	3.34-4.08*	3.50-3.80**
Si, %	2.15-3.17	2.15-3.19
Ni, %	0.34-0.90	0.38-0.66
Mg, %	0.030-0.109	0.030-0.119
Mn, %	0.17-0.59	0.28-0.54
Cr, %	0.038-0.088	0.045-0.092
P, %	0.010-0.029	0.012-0.025
Cu, %	0.08-0.12	0.08-0.12

*Analysis of chilled carbon pins, represents carbon content of molten iron.

**Analysis of Y block wafers, represents carbon content of test specimen.

The data were analyzed by means of multiple linear regression. In the as-cast ductile iron investigation, the tensile properties were considered dependent variables, while the Brinell hardness and chemical determinations were considered independent variables. In the annealed ductile iron investigation, the tensile properties and Brinell hardness were considered dependent variables, while the chemical determinations were considered independent variables. The ranges of variables studied in these investigations are shown in Table 1.

A criterion of 95 per cent confidence was used in determining statistical significance of a relationship which, in turn, determined whether or not an independent variable was used in a regression equation. The accuracies of predictions from the regression equations were calculated in the form of 95 per cent confidence limits. Linear approximations to the true parabolic form of confidence limits were used.

AS-CAST DUCTILE IRON

Subject to the qualifications stated in the Appendix, the tensile properties of as-cast pearlitic-ferritic ductile iron can be predicted from the hardness and composition by means of the following regression equations:

Tensile Strength, psi = 488 (Bhn) − 2,730 (% Si) − 6,260.
95% Confidence Limits: ± 5,650 psi. *(1)*

Yield Strength, psi = 259 (Bhn) + 9,740 (% Si) + 4,750 (% Ni) − 36,800 (% Cr) − 45,200 (% Mg) − 20,680.
95% Confidence Limits: ± 3,570. *(2)*

Elongation, % = 8.8 − 0.0920 (Bhn) + 3.84 (% Si) − 22.2 (% Mg) + 3.47 (% C).
95% Confidence Limits: ± 2.9%. *(3)*

In an iron containing 2.60% Si, 0.50% Ni, 0.070% Cr, 0.060% Mg, and 3.80% C, these equations would be:

Tensile Strength, psi = 488 (Bhn) − 13,350. *(4)*

Yield Strength, psi = 259 (Bhn) + 1,730. *(5)*

Elongation, % = 30.6 − 0.0920 (Bhn). *(6)*

Equations (4), (5) and (6) are shown in graphic form in Fig. 2. Figure 3 shows similar equations for four other compositions.

A comparison of equation (1) with the tensile strength-Brinell hardness ratio of 420 to one shows that use of the simple ratio would not lead to serious inaccuracies for irons with the most common composition and hardness ranges. However, examination of equations (1) and (2) shows that the use of a yield strength to tensile strength ratio, for prediction or evaluation of the yield strength of this material, could result in significant errors. The ratio of yield strength to tensile strength depends not only on composition but also on hardness level. Unless these other factors are known, this ratio is meaningless.

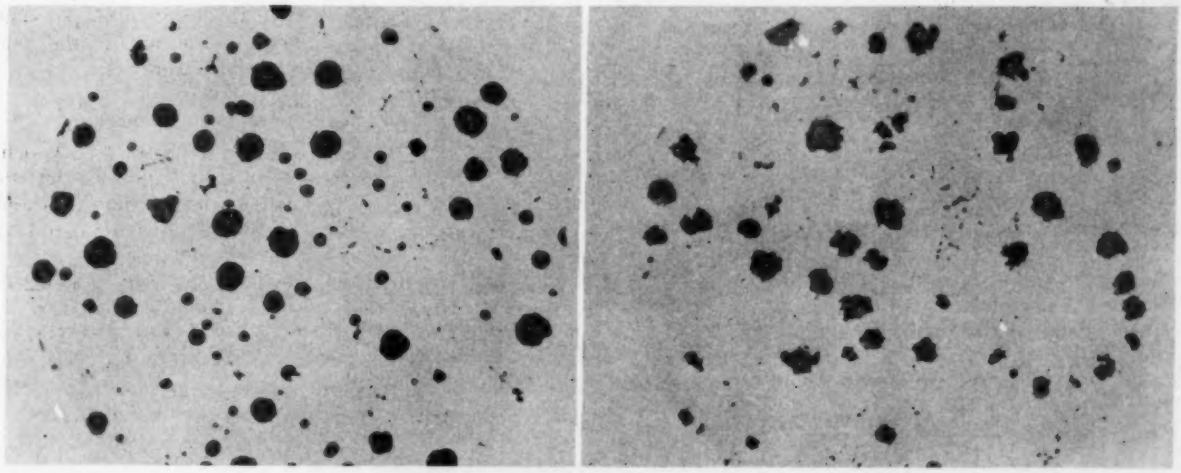
As demonstrated graphically in Fig. 3, the coefficients for chemical composition in equations (1), (2) and (3) may be regarded as the effects of composition on the tensile properties-Brinell hardness relationships. These effects of composition can be rationalized. Increases in the ferrite soluble elements, silicon and nickel, increase the hardness of the ferrite and, therefore, result in a material with less pearlite at a given composite hardness. Increases in these soluble elements also increase the strength and decrease the ductility of the ferrite, although not necessarily at the same rate as the increase in ferrite hardness.

It appears that the tensile strength and elongation depend more upon the amount of pearlite present at a given composite hardness than upon the strength and ductility of the ferrite, whereas the yield strength depends more upon the strength of the ferrite. The absence of nickel from the tensile strength and elongation equations can be attributed to a relatively weak effect in comparison with experimental error. The apparently deleterious effect of increases in chromium or magnesium on yield strength at a given composite hardness can be explained by an increase in hardness of the pearlite due to chromium and magnesium carbides, resulting in a material with the same ferrite strength but less pearlite at a given composite hardness.

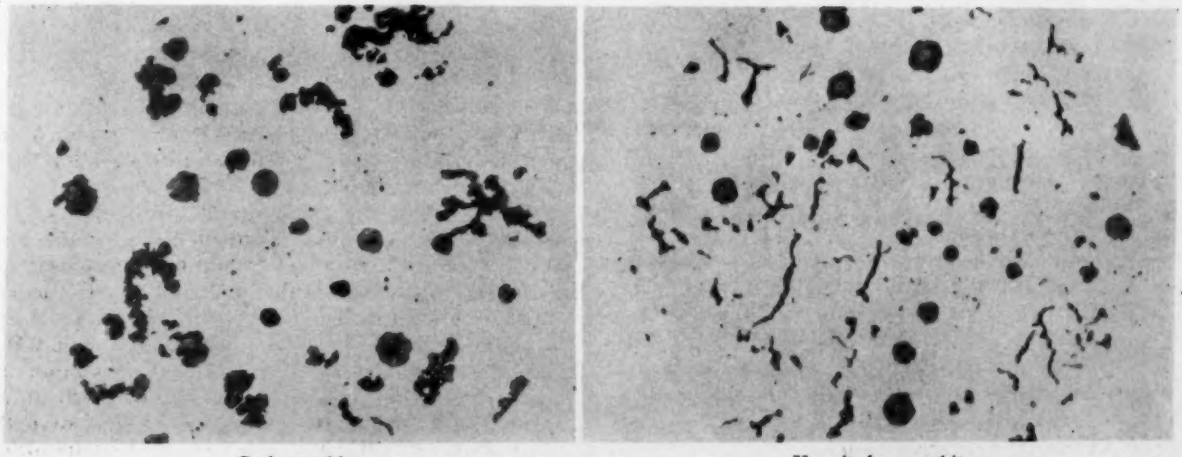
It would be more probable that this effect would be detected in the yield strength regression analysis than in the other regression analyses, because of the relatively greater accuracy of the yield strength regression equation. This hypothesis can also help explain the deleterious effect of increases in magnesium on elongation at a given composite hardness. However, other evidence indicates that this effect is mostly due to an increase in minute amounts of primary cementite with increases in magnesium. The beneficial effect of increases in carbon on elongation at a given composite hardness can then be explained by the recognized ability of increases in carbon to decrease primary cementite.

ANNEALED DUCTILE IRON

Subject to the qualifications stated in the Appendix, the tensile properties and hardness of fully annealed ductile iron can be predicted from the composition by means of the following regression equations:



Nodular graphite, typical variations



Crab graphite — Vermicular graphite

Fig. 1 — Classification of graphite forms found in ductile cast iron. 100 ×.

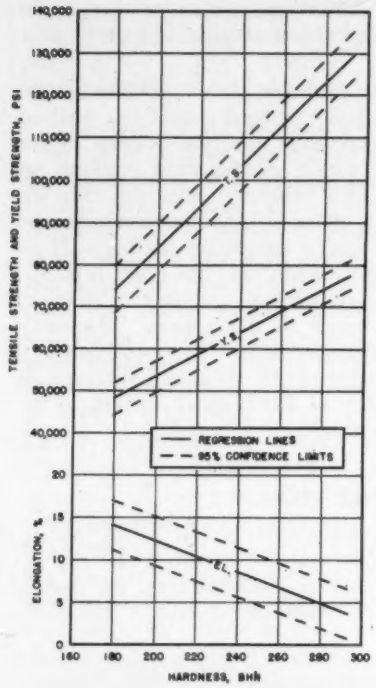


Fig. 2 — (Left) Regression of tensile properties on Brinell hardness for as-cast ductile iron.

C — 3.80%	Cr — 0.070%
Si — 2.60%	Mn — 0.17-0.59%
Ni — 0.50%	P — 0.010-0.029%
Mg — 0.060%	Cu — 0.08-0.12%

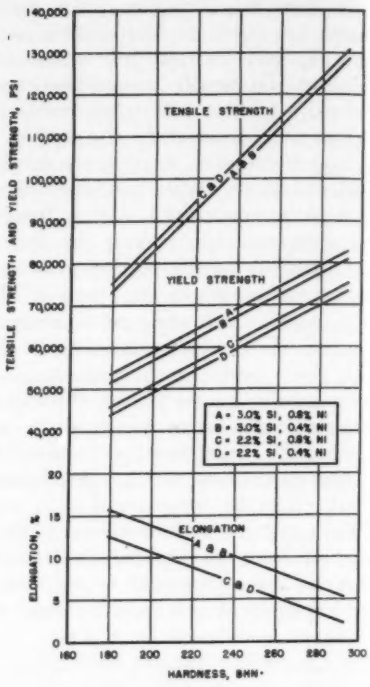


Fig. 3 — (Right) The effect of changes in silicon and nickel content on the elevation of the regression lines in the regression of tensile properties on Brinell hardness for as-cast ductile iron.

C — 3.80%	Mn — 0.17-0.59%
Mg — 0.060%	P — 0.010-0.029%
Cr — 0.070%	Cu — 0.08-0.12%

Tensile Strength, psi = 76,480 + 8,880 (% Si) + 3,340 (% Ni) − 10,520 (% C).
95% Confidence Limits: ± 1,800 psi. *(7)*

Yield Strength, psi = 59,400 + 11,550 (% Si) + 4,970 (% Ni) + 5,880 (% Mn) − 14,010 (% C).
95% Confidence Limits: ± 1,340 psi. *(8)*

Elongation, % = 85.4 − 6.6 (% Ni) + 11.7 (% Mg) − 16.3 (% C).
95% Confidence Limits: ± 2.6%. *(9)*

Hardness, Bhn = 91 + 25.5 (% Si).
95% Confidence Limits: ± 7 Bhn. *(10)*

The relatively narrow confidence limits associated with these equations indicates that the tensile properties of fully annealed ductile iron are determined principally by composition. In as-cast ductile iron, cooling rate following solidification is an additional variable. The use of Brinell hardness as an independent variable in the regression equations for the as-cast material measures the effects of cooling rate as well as partially measuring the effects of composition. With annealed ductile iron, the use of Brinell hardness as an independent variable is not necessary as this material is essentially annealed, alloyed ferrite, modified by a dispersion of graphite nodules.

The cooling rate following solidification should not be a factor in such a material and the cooling rate following annealing, which can affect the hardness of alloyed ferrite according to Austin,[4] did not vary sufficiently to be a factor. As mentioned earlier in this paper, Brinell hardness does not measure the effects of the alloying elements and the graphite on the tensile properties as well as the composition measures these effects.

The effects of the ferrite soluble elements, silicon, nickel and manganese, correspond reasonably well with the effects expected for solid solution alloys in ferrite. The order of effectiveness of these elements in increasing the strength and hardness of annealed ductile iron corresponds with the order of effectiveness in strengthening ferrite reported by Lacy and Gensamer,[5] and the order of effectiveness in hardening ferrite reported by Bain[6] and Austin.[4] The absence of manganese from equations (7), (9) and (10), and of nickel from equation (10), can be attributed to insufficient variability of these elements in comparison with experimental error.

The absence of silicon from equation (9) may be due to partial cancellation of its effect, in lessening the ductility of ferrite, by its effect in promoting more complete annealing. Generally, quantitative comparisons of the effects of these elements on annealed ductile iron with the effects reported on ferrite show few inconsistencies that cannot be explained.

The effects of silicon and nickel on the tensile properties and hardness of annealed ductile iron reported by Reynolds, Adams and Taylor[3] compare well with those found in this investigation. However, the effects of carbon reported by Reynolds, Adams and Taylor[3] are not consistent with the effects found in this investigation. Most of the apparent disagreement can probably be attributed to different methods of carbon analysis. The data shown in an earlier report[7] indicate that Reynolds, Adams and Taylor used chilled pins for carbon determinations. As mentioned earlier in this paper, the carbon content of a chilled pin does not always represent the carbon content of the tensile test specimen.

The general effect of an increase in carbon is to reduce all tensile properties. This effect is probably due to a combination of the reduction in effective cross-section by the graphite nodules with stress concentration dependent on nodule size and shape. The beneficial effect of magnesium on elongation may be

related to the dependence of stress concentration on nodule shape.

APPLICATION OF EQUATIONS

The regression equations shown in this paper permit direct specification of the factors that control the tensile properties of as-cast and fully annealed ductile iron castings. If the composition of the iron and, in the case of as-cast ductile iron, the Brinell hardness of the castings are allowed to vary only within specified ranges, the mean tensile properties of the castings can be calculated from the mean composition and hardness values by means of the regression equations. The variation in tensile properties of the castings can be calculated by means of a method described in the Appendix.

As an example, consider two groups of ductile iron castings, one group in the as-cast condition and the other annealed. The chemical composition of both groups is specified to be to the mean values and within the limits given in Table 2.

It is also specified that no more than 15 per cent pearlite is present in the annealed castings, and that the hardness of the castings to be used in the as-cast condition is controlled to a mean of 226 Bhn and within 95 per cent confidence limits of 197-255 Bhn. It is further specified that, in all cases, the graphite form is essentially nodular, and that no more than a trace of primary cementite exists. The means and ranges of the inherent tensile properties of the castings would then be like those given in Table 3.

Minimum properties designations for these irons would probably be 80-50-06 for the as-cast iron, and 60-40-20 for the annealed iron. It is readily apparent from the equations, and from Fig. 3, that ranges of tensile properties substantially different from those shown can be obtained. These ranges of tensile properties refer to the inherent properties of the castings as measured by standard test specimens machined from adequately fed sections. They do not refer to the surfaces of castings, to areas containing voids, slag or carbon flotation, or to the geometric centers of most sections. With these exceptions, test bars machined from production castings have been found to conform reasonably well to predictions.

CONCLUSIONS

The tensile properties of as-cast pearlitic-ferritic ductile iron can be predicted with reasonable accuracy from the Brinell hardness and composition. The most direct and effective way of insuring control of the inherent tensile properties of castings of this material is to specify composition ranges and Brinell hardness ranges for the castings. Obviously, additional specifications concerning such factors as graphite shape and permissible amount of primary cementite are also necessary.

The tensile properties of annealed ductile iron can be predicted with reasonable accuracy from the composition. The most direct and effective way of insuring control of the inherent tensile properties of castings of this material is to specify composition ranges. Again, it is obvious that additional specifications concerning such factors as graphite shape and permissible amounts of primary cementite are necessary. For this material, it is also necessary to specify the amount of residual pearlite allowable in the castings.

TABLE 2 — CHEMICAL COMPOSITION MEAN AND RANGE VALUES

Composition, %	Mean	95% Confidence Limits
C*	3.80	3.60-4.00
Si	2.60	2.40-2.80
Ni	0.50	0.40-0.60
Mg	0.060	0.040-0.080
Mn	0.50	0.35-0.65
Cr	0.070	0.052-0.088
P	0.016	0.008-0.024
Cu	0.10	0.09-0.11

*Carbon content of the molten iron.

TABLE 3 — INHERENT TENSILE PROPERTIES, MEAN AND RANGE VALUES

	As-Cast Ductile Iron		Annealed Ductile Iron	
Property	Mean	95% Confidence Limits	Mean	95% Confidence Limits
Tensile Strength, psi	96,930	81,550-112,310	62,490	59,380-65,600
Yield Strength, psi	60,260	51,550-68,970	43,250	39,550-46,950
Elongation, %	9.9	5.8-14.0	23.3	20.4-26.2

ACKNOWLEDGMENTS

The authors gratefully acknowledge the cooperation and assistance given by John Deere Malleable Works, who supplied the castings used in these investigations. Appreciation is extended to the I.B.M. Processing Dept., Deere & Company, for performing on an electronic computer the lengthy calculations involved in multiple regression analysis. The authors also thank the personnel of the Materials Engrg. Dept., who conducted the mechanical tests, chemical determinations and metallographic examinations, and those who provided the guidance and encouragement for this program.

REFERENCES

1. *Engineering Properties and Application of Ductile Iron*, International Nickel Co., p. 17, 1957.
2. R. Schneidewind and H. H. Wilder, "Commercial Experience with Higher Silicon Nodular Irons," AFS TRANSACTIONS, *vol. 60*, pp. 322-329, 1952.
3. C. C. Reynolds, C. M. Adams and H. F. Taylor, "Prediction of Mechanical Properties from Chemical Composition for Fully-Annealed Ductile Cast Iron," AFS TRANSACTIONS, *vol. 61*, pp. 510-515, 1953.
4. C. R. Austin, "Effect of Elements in Solid Solution on Hardness and Response to Heat Treatment of Iron Binary Alloys," *A.S.M. Transactions, vol. 31*, pp. 321-339, 1943.
5. C. E. Lacy and M. Gensamer, "The Tensile Properties of Alloyed Ferrites," *A.S.M. Transactions, vol. 32*, pp. 88-110, 1944.
6. E. C. Bain, *Functions of The Alloying Elements in Steel*, A.S.M., pp. 65-67, 1939.
7. C. C. Reynolds and H. F. Taylor, "Mechanical Properties of Spherulitic Graphite Cast Iron," AFS TRANSACTIONS, *vol. 60*, pp. 687-713, 1952.

APPENDIX

Multiple linear regression analysis is a standard statistical technique which enables the analyst to determine whether or not statistically significant linear relationships exist between dependent and independent

variables, calculate the best linear equations for predicting values of the dependent variable from knowledge of the independent variables and measure the accuracy of these predictions. The derived regression equations are not functional equations, and should not be used to predict values of an independent variable from knowledge of the other variables.

As stated in the description of procedure, a criterion of 95 per cent confidence was used in these investigations for determining statistical significance of a relationship. Only those independent variables significantly related to the dependent variables were included in the regression equations. Therefore, there are less than five chances in 100 that no relationship exists when one is shown, or that the direction of relationship shown is wrong. Lack of statistical significance, and the absence of an independent variable from a regression equation, does not mean that this variable has no effect on the dependent variable in question.

The proper interpretation is that the existence of a relationship could not be demonstrated statistically with a reasonable degree of certainty from the available data for one or a combination of the following reasons:

1) The variation in the independent variable was too small in comparison with experimental error.
2) The relationship is too weak in comparison with experimental error.
3) The curvature of the relationship is too great for the linear assumption.

On the other hand, a relationship shown by a regression analysis to be statistically significant does not constitute proof of cause and effect. A significant relationship only demonstrates that a high degree of association existed during the period and under the circumstances in which the data were collected. However, for reasons of convenience, these relationships are occasionally referred to as causes and effects in this paper.

It is also stated in the description of procedure that the accuracies of predictions from the derived regression equations are presented in the form of 95 per cent confidence limits. These limits should contain 95 of 100 measured values of the dependent variables. Predictions from a regression equation are valid only over the ranges of independent variables studied. Extrapolation beyond these ranges, which, for these investigations, were presented in Table 1, could be misleading.

Care should also be taken, in predicting from a regression equation, that the ranges of variables not shown in the regression equation, yet suspected of affecting the dependent variable, do not differ substantially from the ranges studied. For instance, copper and phosphorus are both known to affect the tensile properties of ductile iron, yet neither is shown in any of the regression equations. If these regression equations are used for irons with copper and phosphorus contents substantially different from the ranges encountered in these investigations, adjustments in elevation of some of the regression lines would be necessary.

As another example, manganese is not shown in the regression equation of tensile strength of annealed ductile iron on composition, yet it would not be advisable to predict from this regression equation for irons with manganese contents much beyond the range investigated.

The variation in a dependent variable resulting from variation of the independent variables within stated limits can be computed by means of equations of the following type:

$$\sigma_{TS}^2 = (b_{Bhn}\sigma_{Bhn})^2 + (b_{Si}\sigma_{Si})^2 + \sigma_E^2$$

where

σ_{TS}^2 = The total variance in tensile strength.
b_{Bhn} = The coefficient for Brinell hardness in equation (1).
σ_{Bhn} = The standard deviation of Brinell hardness, a value dependent on control limits.
b_{Si} = The coefficient for silicon in equation (1).
σ_{Si} = The standard deviation of silicon content, a value dependent on control limits.
σ_E^2 = The variance of a point predicted from equation (1), contains the experimental error.

The above equation applies to the tensile strength of as-cast ductile iron. Similar equations apply to the other dependent variables investigated. Such calculations assume an approach to normality in the distribution of each variable. It is also assumed that the so-called independent variables are truly independent.

An objection may arise that, in the case of as-cast ductile iron, hardness is dependent on composition, yet is considered an independent variable along with the composition. This objection is valid to a limited extent. However, hardness is also varied independently of composition by means of cooling rate variation, which, in turn, is dependent on section size and mold dumping time.

In the case of annealed ductile iron, carbon is sometimes not independent of silicon. As stated in an earlier section of this paper, the carbon determinations in the annealed ductile iron investigation differ from those in the as-cast ductile iron investigation, in that the former represent the carbon content of the castings whereas the latter represent the carbon content of the molten iron. According to a paper soon to be published by A. H. Rauch, J. B. Peck and G. F. Thomas, the carbon equivalents (% C + ⅓% Si) of medium section size castings, based on carbon determinations from wafers cut from the castings, have a top limit of approx. 4.55 per cent.

No difficulty arises in calculation of variation, if the carbon and silicon ranges in the molten iron are such that a carbon equivalent of 4.55 per cent is not reached. If the carbon and silicon ranges are such that the carbon equivalent of the molten iron seldom falls below 4.55 per cent, the carbon content may be considered wholly dependent on the silicon content, carbon may be eliminated from the regression equation and a new coefficient for silicon may be computed. No solution is presented for intermediate cases.

GRAY CAST IRON MACHINABILITY

quantitative measurements of pearlite and graphite effects

By W. W. Moore and J. O. Lord

ABSTRACT

Common types of unalloyed cupola-melted gray cast iron were studied to determine how changes in some of the major constituents of their microstructures affected machinability. The objective of the study was to make a contribution to the understanding of fundamental and basic factors affecting the machining ease of gray cast iron.

INTRODUCTION

The conditions for the study were set up so as to minimize or eliminate effects from three major factors known to have important adverse effects on the machinability of gray iron. These factors are:

1) The presence of the as-cast surface.
2) The presence of massive carbides.
3) The presence of iron phosphide eutectic (steadite).

The effects of these factors on machinability already have been studied in detail and are well known. Emphasis in this study was directed elsewhere—at the effects of other variables which have received less attention in prior investigations. The factors considered in this study included:

1) The relative volume of pearlite in the matrix.
2) The relative volume of graphite in the iron.
3) The relative size of the graphite flakes in the iron.

Standard A.S.T.M. bars 1.2 in. in diameter were cast in green sand at several commercial foundries from unalloyed gray irons meeting the specifications of A.S.T.M. Designation A-48, Classes 20 and 30. Some of the irons were inoculated and some were not. The matrix of the irons contained about 75 to 100 per cent pearlite, the balance being ferrite. Bars contained graphite flakes of A.S.T.M. Types A, B, D and E. The bars were free of massive cementite, and most of them were free of all but the small amounts of steadite typical of such irons.

The influence of the as-cast surface was eliminated by machining the bars to 1 in. in diameter prior to subjecting them to a constant-pressure lathe test. Bars fitting the foregoing description are, of course, generally conceded to be easily machinable.

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S.A.E. Classification 1113 free-cutting steel (AISI B1113), was used as a basis for comparison, and was arbitrarily assigned a machinability index of 100. All gray irons in the study machined more easily than the reference steel. The irons had machinability indices of 139 to 232. The research was aimed at a determination of the causes for the machinability variation shown by these common types of gray iron.

The common practice of using reference photomicrographs to classify microstructures was not followed in this investigation. Classifications based on reference photographs are only qualitative. This study made use of lineal analysis, a quantitative method of classifying microstructure. Using this method, numerical values were obtained for:

1) The relative volume of pearlite.
2) The relative volume of graphite.
3) The relative size of the graphite flakes.

These measurements were made by metallographic examinations of specimens from each bar. Because photomicrographs were not used in this investigation, none are included in this paper.

Multiple Regression Analysis

The results of machining tests, and the results of lineal analyses of microstructures, were subjected to a multiple regression analysis of a digital computer to derive an equation relating machinability to microstructure.

The resulting equation showed that an increase in machinability rating was obtained by:

1) A decrease in the relative volume of pearlite.
2) An increase in the relative volume of graphite.
3) An increase in the relative size of the graphite flakes.

The effect of a change in size of the graphite flakes was small compared to the effects of the other two variables. Examples of the quantitative effect of these variables on machinability indices are summarized in Table 1.

Machinability ratings calculated from the new equation were compared with ratings based on tensile strength, hardness and chemical composition of the irons. Each of these three parameters is commonly used as an indication of probable machin-

ability. The method derived in this study yielded better agreement between calculated machinability index and actual machinability index than was obtained by prediction from tensile strength, hardness or chemical composition.

The findings of this study were consistent with the common understanding that machinability generally is improved in any metal by 1) reducing its strength, 2) reducing its ductility, or 3) reducing its coefficient of friction in contact with the tool. The results of this study express these effects quantitatively in terms of the major microstructural constituents in ordinary unalloyed gray iron.

SOURCE OF IRONS

Bars of 18 gray irons were produced by member foundries of the Gray Iron Research Institute. They were standard A.S.T.M. bars 1.2 in. in diameter. The mechanical properties and chemical analyses of these bars are given in Table 2.

The conditions represented in the study involved the following limitations. The bars were unalloyed, and ranged in tensile strength from about 20,000 to 40,000 psi. The bars were all of the same original diameter, and were cast in green sand molds from metal melted in commercial cupolas. Machinability tests were made after about 0.1 in. had been machined from the surface of the bars so as to eliminate effects of surface skin, surface finish and possible surface defects.

MACHINABILITY PROCEDURES

The machinability indices for this study were obtained from constant-pressure lathe tests.[1] This test was based on the fact that materials of different machining qualities cut at different rates when surface speed, depth of cut and tool geometry are held constant, and a fixed lateral force is applied to the tool. For example, the present tests on gray cast iron were conducted with:

1) Surface speed of 27 ft/min.
2) 1/8-in. depth of cut.
3) High-speed steel tool with a 12-degree side rake angle and a 0-degree back rake angle. The lateral force applied to the lathe carriage by a weight and pulley was 60 lb.

In each series of tests, several bars of cast iron, and a reference bar of B1113 steel, were treated as a group. Each individual test consisted of a number of 2-in. cuts along each bar. During each test, a record was made of the number of spindle revolutions which occurred during each of ten consecutive 0.2-in. increments of tool travel. The average of these numbers was termed the R value. Higher R values indicated smaller feeds and a bar harder to machine. To obtain a reproducible index of machinability, the R value of the bar under study was compared with the R value of the bar of reference steel used in the same series. The machinability index of the standard was taken arbitrarily as 100.

MICROSTRUCTURE EXAMINATION PROCEDURE

The microstructure of each bar was evaluated by lineal analysis.[2] Measurement was made of the amount of each phase present in the sample. A value for the relative size of graphite particle was also obtained.

The apparatus used in lineal analysis was a Hurlbut counter, a device for measuring the travel of a mechanical stage under a microscope. The mechanical stage was driven by a motor through a gear train. The stage could be moved by engaging any one of seven similar gear trains. Each gear train was provided with a numerical counter to record the

TABLE 1 — MICROSTRUCTURE CHANGES EFFECTS ON GRAY CAST IRON MACHINABILITY

Condition[1]	Machinability Index in the Constant-Pressure Lathe Test (1113 Steel = 100, Higher Indices Indicate Better Machinability) Index	Change in Index Due to Indicated Change in Microstructure
Average for irons studied (93.8% pearlite, 8.29% graphite, relative size of graphite — 3.32 microns)	179	0
Decrease of 10 per cent in volume of pearlite only. For example, from 93.8% to 84.4%	191	+12
Increase of 10 per cent in volume of graphite only. For example, from 8.29% to 9.12%	189	+10
Increase of 10 per cent in relative size of graphite only. For example, from 3.32 to 3.65 microns	179.4	+0.4
Cumulative effect of a decrease of 10 per cent in volume of pearlite, increase of 10 per cent in volume of graphite, and increase of 10 per cent in relative size of graphite	201.4	+22.4

1. Percentages refer to relative volumes.

TABLE 2 — TEST BARS PROPERTIES AND COMPOSITIONS[1]

Bar No.	Tensile Str., psi	Bhn	Chemical Composition, % TC	Si	C.E.[2]	Mn	S	P
W - 1	38,000	207	3.26	2.06	3.95	0.92	0.09	0.11
W - 2	38,100	207	3.24	2.27	4.00	0.92	0.09	0.11
W - 3	27,900	183	3.35	2.42	4.16	0.66	0.15	0.09
W - 4	27,600	179	3.37	2.65	4.25	0.66	0.15	0.09
W - 5	32,200	197	3.24	2.35	4.02	0.51	0.13	0.18
W - 6	34,200	192	3.20	2.53	4.04	0.51	0.13	0.18
W - 7	31,900	187	3.37	2.01	4.04	0.71	0.10	0.10
W - 8	27,700	174	3.39	2.41	4.22	0.71	0.10	0.10
W - 9	30,900	208	3.48	1.70	4.05	0.38	0.06	0.49
W - 10	29,600	199	3.48	1.90	4.11	0.38	0.06	0.49
W - 11	20,900	170	3.63	2.58	4.49	0.57	0.11	0.17
W - 12	19,500	174	3.63	2.67	4.52	0.57	0.11	0.17
W - 13	22,100	172	3.63	2.58	4.49	0.57	0.11	0.17
W - 14	22,200	174	3.63	2.67	4.52	0.57	0.11	0.17
W - 15	31,900	192	3.38	2.29	4.14	0.90	0.11	0.08
W - 16	32,400	197	3.38	2.42	4.19	0.90	0.11	0.08
W - 17	38,900	207	3.26	1.78	3.85	—	—	0.30
W - 18	39,200	207	3.22	1.91	3.86	—	—	0.30

1. Information in this table furnished by Gray Iron Research Institute.

2. C.E. = T.C. + $\frac{1}{3}$ Si.

travel of the stage while that gear was engaged. The microstructure of the specimen being examined was viewed through a metallurgical microscope at a magnification of 1000 ×.

A cross-hair in the eye piece provided a reference point. A series of traverses was made of each specimen. Each microconstituent under study was assigned to one gear train. The stage was moved by engaging only the gear train assigned to the particular microconstituent under the reference hair at any given instant.

Tabulated results of the measurements and calculations for:

1) Relative volume of pearlite;
2) relative volume of graphite;
3) relative size of graphite flakes;

are given in Table 3 for each of the 18 irons included in the study. Table 3 also includes the machinability indices determined experimentally on each iron.

The total distance of travel of the reference point across the specimen during any one traverse was taken as L. The sum of individual distances traveled across one particular phase during the same traverse was taken as L'. The ratio of L'/L was taken as V', the relative volume of the particular phase as a fraction of the total volume of all phases.

Graphite Size Determination

The average relative size of graphite flakes was determined with the use of the Hurlbut counter in the following manner. In a given traverse, measurements were made of:

1) The sum of the individual distances in microns* that the reference point traveled across graphite flakes.
2) The number of graphite flakes.

The division of the sum of the distances in microns by the number of flakes yielded a value which was taken as the average relative size of the graphite flakes in microns. The resulting numerical value for relative size of flakes was not a measure of any specific dimension of the flakes.

In the Hurlbut-counter method, accuracy of measurement is inversely proportional to the size and the number of the areas of the particular phase undergoing measurement. When a traverse involves measurements on a large number of particles of small size (as in the measurement of the volume of graphite flakes in gray iron), the accuracy of the Hurlbut-counter method is lower than when a fewer number of larger particles or phases is measured. To double check on the counter method, separate and independent calculations were made to estimate the relative volume of free graphite in each bar.

The relative volume of pearlite in the microstructure was measured. The total carbon content of the irons was determined by analysis. The pearlite in each bar was assumed to contain 0.8 per cent of carbon. Because the irons contained no massive carbides, the balance of the total weight of carbon in the iron was assumed to exist as free graphite.

*1 micron = 0.001 mm = 0.00004 in.

TABLE 3 — MACHINABILITY TESTS AND MICROSTRUCTURE ANALYSES DATA

Bar No.	Machinability Index[1]	Relative Volume of Pearlite (V_p), %	Relative Volume of Graphite (V_g), %[2]	Relative Size of Graphite (S_g), microns[3]
W - 1	151	99.8	7.78	2.67
W - 2	153	100.0	7.59	3.46
W - 3	184	99.5	7.96	3.85
W - 4	185	97.2	8.08	3.65
W - 5	198	76.6	8.18	3.12
W - 6	179	92.7	7.73	3.16
W - 7	177	98.1	8.06	3.24
W - 8	177	99.6	8.08	3.77
W - 9	177	99.9	8.54	5.01
W - 10	169	99.6	8.52	4.00
W - 11	203	89.7	9.09	2.61
W - 12	203	85.2	9.21	2.12
W - 13	193	83.0	9.26	2.18
W - 14	232	72.8	9.65	2.15
W - 15	164	97.8	8.09	1.39
W - 16	165	99.7	8.09	3.55
W - 17	164	97.3	7.76	4.93
W - 18	145	99.2	7.63	4.90

1. Basis is machinability index of B1113 steel = 100. Higher indices indicate better machinability. Tabulated values are averages of six 2-in. cuts on each bar.
2. Based on calculation from relative volume of pearlite and total carbon content as described in the text.
3. 1 micron = 0.001 mm = 0.00004 in.

By use of the relative densities of graphite and iron, the calculated weight of free graphite in the iron was converted to relative volume of graphite in the iron. These calculated values were uniformly about 20 per cent lower than the values obtained by the counter method. Because of this uniform relationship between the two methods, it was possible to use the results of either method for defining the relative volume of graphite in different samples.

VARIABLES EFFECTS ON MACHINABILITY INDEX

The machinability indices of the irons were found to be related to each of the three variables studied. Figure 1 shows how the machinability index decreased when the amount of pearlite in the matrix

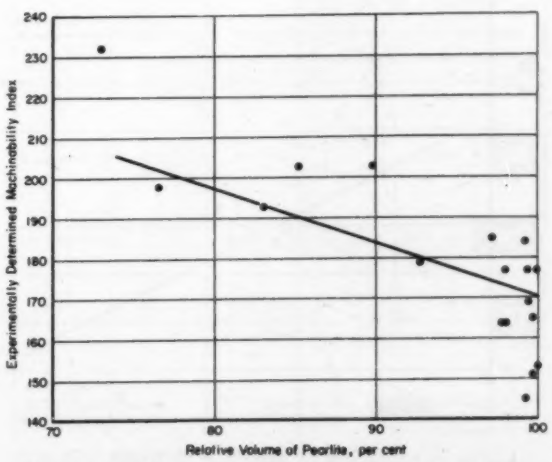


Fig. 1 — Variation of machinability of gray iron with relative volume of pearlite in the iron.

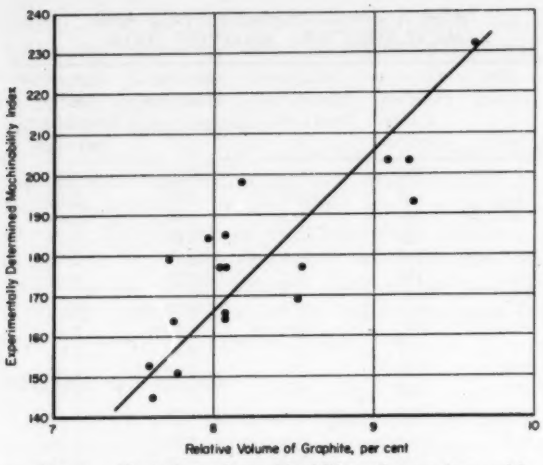


Fig. 2 — Variation of machinability of gray iron with relative volume of graphite in the iron.

increased. The spread of data points above about 97 per cent of pearlite is easily explainable. The iron cannot contain more than 100 per cent of pearlite. Once an iron contains about 100 per cent of pearlite, factors which affect machinability adversely obviously cannot be accompanied by a proportional increase in the percentage of pearlite. This explains the results on the three irons which had low machinability indices of about 140 to 160, and which also had virtually a completely pearlitic matrix.

Figure 2 shows how machinability increased with an increase in the relative volume of graphite in the iron. Figure 3 shows that an increase in the average relative size of the graphite flakes tended to cause a decrease in machinability, but that the relationship between machinability index and the size of the graphite flakes was more variable than the relationship shown in Figs. 1 and 2.

Figures 1 through 3 show that each of the three variables under study have approximately a linear effect on machinability index over the range of

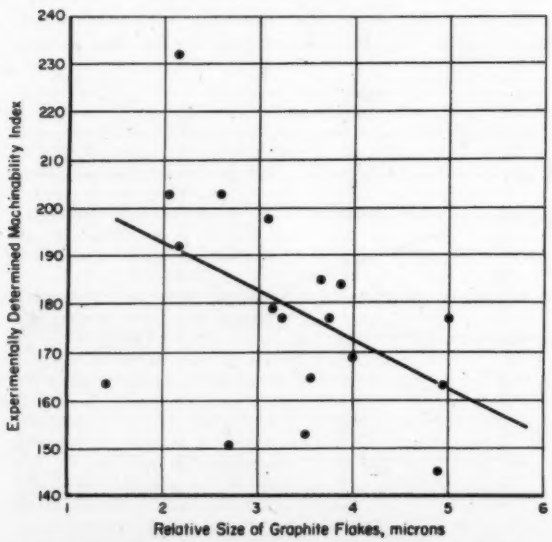


Fig. 3 — Variation of machinability of gray iron with relative size of graphite flakes in the iron.

variables studied. Because each variable had an effect, it is to be expected that graphs of machinability index against only one of the three variables would show considerable scatter. It was decided, therefore, to combine the effects of the three variables into one parameter.

Assuming that a linear relationship existed for each variable, a multiple regression analysis was performed on an I.B.M. digital computer to derive an equation which combines the effects of all three variables on the machinability index. The resulting equation which best fits the experimental data was as follows:

$$M = 195.5 - 1.26\,V_p + 11.7\,V_g + 1.2\,S_g$$

where

M = Machinability index.
V_p = Relative volume of pearlite in per cent.
V_g = Relative volume of graphite in per cent.
S_g = Relative size of graphite in microns.

Figure 4 shows how the machinability indices calculated from this equation compared with the indices which were determined experimentally.

NEW METHOD EVALUATION

Three bases sometimes used to predict the machinability of gray iron are:

1) Chemical composition as defined by the carbon equivalent (C.E. = T.C. + ⅓ Si) of the iron.
2) Brinell hardness of the iron.
3) Tensile strength of the iron.

Figures 5, 6 and 7 show the relationship between each of these three parameters and the machinability indices determined experimentally.

A comparison was made of the accuracy with which each of these three parameters and the new method would predict machinability. The results of

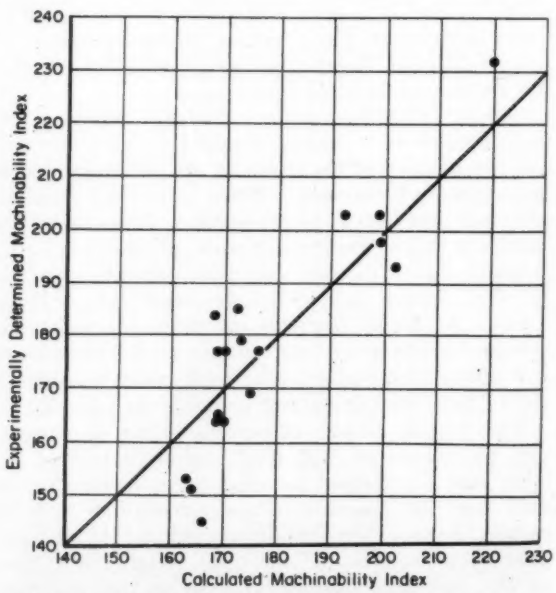


Fig. 4 — Comparison of machinability indices calculated from the multiple regression equation with machinability indices determined experimentally.

this comparison are shown in Fig. 8. The new method based on microstructure gave the best correlation between predicted and actual values for machinability. Prediction based on carbon equivalent was second.

A particular value of the new method based on microstructure is that its use would call attention to the presence of factors other than those included in this study. Specifically, examination of the microstructure will reveal the presence of massive carbides and steadite which are extremely harmful to machinability. None of the other methods can be depended upon to do this.

INOCULATION EFFECT

Because some of irons included in the study had been inoculated and some had not, a determination was made of how inoculation affected ability to predict machinability from examination of the microstructure. Figure 9 is the same as Fig. 4, except that Fig. 9 indicates which irons had been inoculated and which had not.

Figure 9 shows that the relationship between calculated and actual values for the machinability indices was not affected by inoculation. This does not imply that inoculation had no effect on machinability. Figure 9 shows that any effect which inoculation did have on machinability was detected and evaluated by the method used. The method, therefore, is equally applicable to inoculated and to uninoculated irons machined under the same conditions as those used in this study. In this connection, it is necessary to recall that the surface layer of the bars is most affected by inoculation, and that the surface layers of all bars were removed prior to testing.

PHOSPHORUS CONTENT EFFECT

Phosphorus in gray iron is generally considered to have a deleterious effect on machinability because

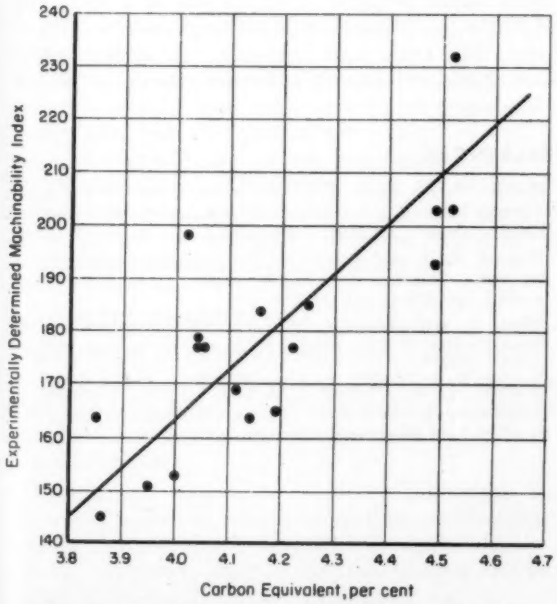


Fig. 5 — Relationship of machinability to carbon equivalent.

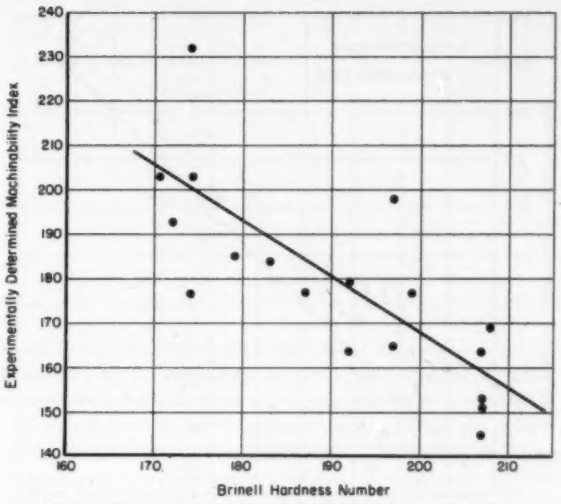


Fig. 6 — Relationship of machinability to Brinell hardness number.

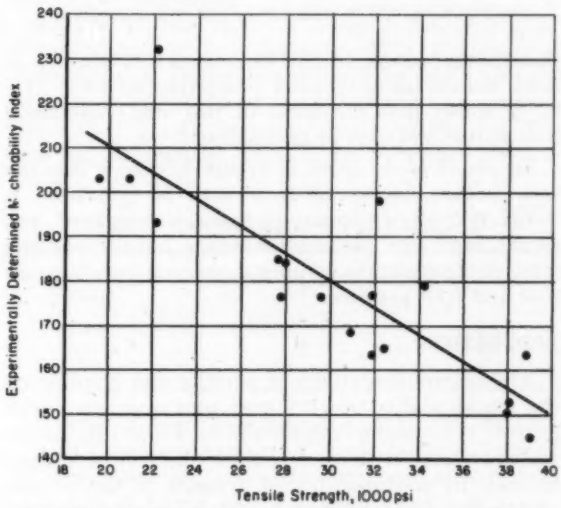


Fig. 7——Relationship of machinability to tensile strength.

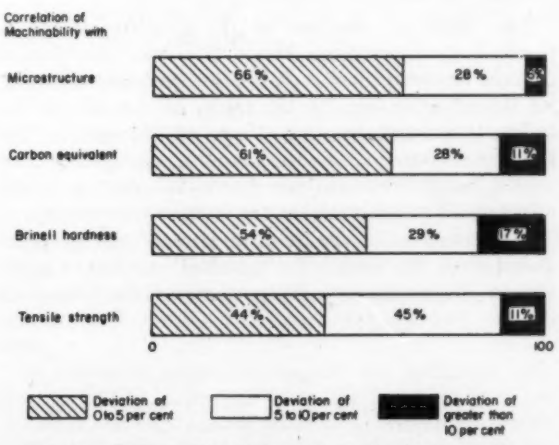


Fig. 8 — Range of deviations of machinability indices determined experimentally from machinability indices calculated from regression equations.

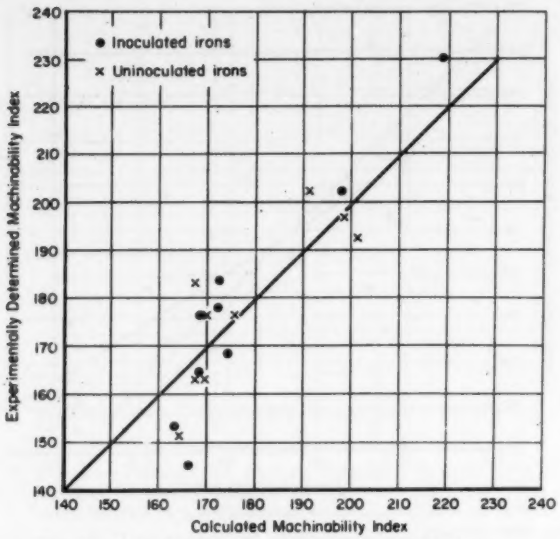


Fig. 9 — Inoculation effect on the relationship between machinability indices determined experimentally and machinability indices calculated from microstructure.

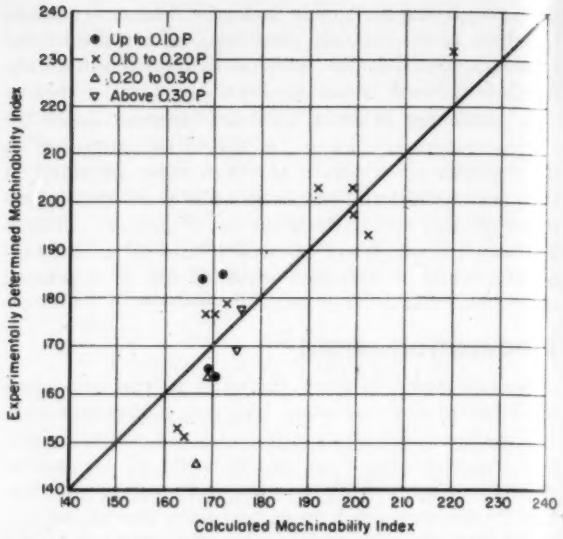


Fig. 10 — Phosphorus content effect on the relationship between machinability indices determined experimentally and machinability indices calculated from microstructure.

phosphorus occurs in the microstructure mainly as hard massive steadite (iron phosphide eutectic). Table 2 shows that the irons in this study contained 0.08 to 0.49 per cent of phosphorus.

Figure 10 is the same as Figure 4, except that the phosphorus content of each iron is indicated. It shows that the relationship between calculated and actual vaues for the machinability indices was not affected significantly by phosphorus contents between 0.08 and 0.49 per cent.

DISCUSSION

The determined effects of pearlite and graphite on the machinability of gray iron were consistent with general principles of machinability. Generally, factors which reduce the strength and/or ductility of a metal increase its machinability. A decrease in the amount of pearlite, or an increase in the amount of graphite, will decrease both the strength and the ductility of gray iron, and each of these effects was found experimentally to increase machinability.

The effect of the size of the graphite flakes is particularly interesting. When a simple correlation is made between machinability and the size of graphite flakes (as in Fig. 3), the effect of the size of the flakes is masked by the effects of changes in the amount of pearlite and the amount of graphite. The multiple regression analysis shows that there is a true effect of the graphite flakes size upon machinability.

Measurement of the relative sizes of the graphite flakes may be useful for another purpose. Coarse (large) flakes may indicate general coarseness of microstructure. This is suggested by Field and Stansbury[3] who show that machinability is improved by coarser microconstituents. This possibility is of particular interest since the coarsening of microstructure studied by Field and Stansbury was brought about by changes in section thickness.

Boulger[4] indicates that machinability may be directly affected by graphite size with larger particles being more effective in decreasing friction. This situation would be analogous to the improvement in machinability brought about by increased size of sulfide particles in a free-machining steel.

ACKNOWLEDGMENT

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REFERENCES

1. Boulger, F. W., Shaw, H. L. and Johnson, H. E., "Constant-Pressure Lathe Test for Measuring the Machinability of Free-Cutting Steels," *A.S.M.E. Transactions, vol. 71,* p. 431, 1949.
2. Howard, R. T. and Cohen, M., "Quantitative Metallography by Point Counting and Lineal Analysis," *A.I.M.E. Transactions, vol. 172,* p. 413, 1947.
3. Field, M. and Stansbury, "Effect of Microstructure on the Machinability of Cast Irons," *A.S.M.E. Transactions, vol. 69,* p. 665, 1947.
4. Boulger, F. W., "Influence of Metallurgical Properties on Metal-Cutting Operations," Annual Meeting, American Society of Tool Engineers, Philadelphia, 1958.

MALLEABLE IRON A MAGNETIC ALLOY

By W. K. Bock

ABSTRACT

Malleable iron is well-known for its strength, toughness and machinability, and is frequently specified where these properties are of prime importance. Malleable iron also has excellent magnetic properties, and although it has been specified for electromagnetic applications designers have not had, in one place, data available to them on the magnetic properties. This paper is presented to correct that situation.

INTRODUCTION

The work considered here will be divided under three general headings:

1) Magnetic theory with definitions and calculations.
2) Magnetic properties.
3) Other properties.

The first section will not be necessary to the designers who work regularly with magnetic problems. However, it is offered, without apology, because some designers may find it handy, and because it gives a background for the argument that malleable iron is worthy of serious consideration as a magnetic material.

The second section, dealing with magnetic properties, will give the important properties of malleable iron and, in order that the reader may have a chance of comparison, will also give properties of other soft magnetic alloys.

The magnetic properties of materials which could be considered for electromagnetic applications are so nearly alike that the final choice is often decided on other properties such as ease of forming, machinability, strength, etc. Therefore, in a review dealing with the use of malleable iron as a magnetic material, it is natural to compare pertinent nonmagnetic properties of malleable and other materials.

MAGNETIC THEORY

Magnetism is a natural force which can be made to perform many useful functions. The two elements in every application are a magnetic field and a magnetic material, that is, a material which can respond to a magnetic influence.

The earliest application of the phenomenon was the use of lodestone for direction finding. Magnetite, the mineral of which lodestone is composed, is a natural permanent magnet, and its unique property is still used in the beneficiation of iron ore. Today, however, in an engineering sense, magnetite or lodestone is not considered a magnetic material.

Present day magnetic alloys are based on three ferromagnetic metals. Ferromagnetic elements are those which have strong susceptibility to magnetic fields. Of the 92 naturally occurring elements, only four are ferromagnetic: iron, cobalt, nickel and gadolinium. Since the last is a scarce rare earth metal, it is of no commercial importance.

Any one of these metals will lose its magnetic quality if heated to a sufficiently high temperature. Gadolinium is ferromagnetic only at low temperatures. Each of the others has a well-defined temperature, called the Curie point, above which they lose their ferromagnetism. These Curie points are given in Table 1.

TABLE 1 — CURIE POINTS OF ELEMENTS

Metal	Curie Point °C	Curie Point °F
Iron	768	1414
Cobalt	1115	2039
Nickel	353	665

In order to react to a magnetic influence, the metal must become a magnet. The mechanism involved has been the subject of much theorizing. At one time, the accepted theory was that each atom was a subminiature magnet, and these magnets were normally oriented randomly and their effects cancelled out. When a magnetic field was applied, these atomic magnets lined up, and their effect was additive and the metal made a magnet.

Domain Magnetic Theory

This theory has largely been superseded by the domain theory, which states that blocks of atoms act together to produce a magnetic effect. The details of the mechanism belong to the field of atomic physics, and are beyond the scope of this paper. There must be some justification of the viewpoint, however, because out of the domain theory came the Heussler alloys which are ferromagnetic, but composed of the nonmagnetic metals aluminum, copper and manganese.

In the space around a magnet, the influence of a magnet can be detected. The simplest demonstration of this is the high school physics experiment in which a sheet of paper is placed over a bar magnet and is sprinkled with iron powder. With little coaxing the iron filings line up to form a pattern, like that in

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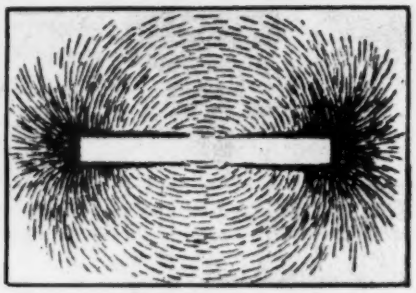


Fig. 1 — Cross-section of magnetic pattern using a bar magnet.

Fig. 1. Of course, the pattern mapped out by the iron is only a cross-section of the whole field.

Demonstrations like this make it easier to introduce the concept of a line of force. The magnetic field can be considered to be made up of a bundle of lines of force, each one running from the North Pole to the South Pole. The lines of iron particles in Fig. 1 show this. Of course, this picture of the magnetic field is not a true one, because the magnetic field is a continuum, but the effect of the field is much like that of a number of lines.

Field Intensity

In order to describe the strength of a magnetic field at any point in space, it is natural to state the number of lines per unit area measured in a plane at right angles to the direction of the lines of force. Such a definition of field strength or field intensity is not practical for magnetic work, and so the unit of field intensity which is commonly used is the "oersted."

A rigorous definition of the oersted would sound too metaphysical for this paper, so it is "defined" here in terms of a common method of producing a magnetic field. A magnetic field can be generated by passing an electric current through a coil of wire or a solenoid. The field intensity (number of oersteds) depends on the number of turns of wire in the coil per inch of its length, and on the number of amperes of current flowing. The number of oersteds can be found by the following expression.

$$\text{Oersteds} = 1.2566 \text{ ampere-turns/cm} = 0.4947 \text{ ampere-turns/in.} \quad (1)$$

If a piece of iron, or other ferromagnetic substance, is placed in the coil, the lines of force will prefer to go through the iron rather than the air. Thus, the iron core will tend to concentrate the magnetic field of the solenoid. A certain flux density will be induced in the iron. By proper laboratory techniques, it is possible to determine how many lines per unit area of iron are induced.

The unit for reporting the result is the gauss, which is defined by:

$$\text{Gauss} = 1 \text{ line/sq cm} = 0.155 \text{ line/sq in.} \quad (2)$$

There may be some question why both the oersted and the gauss are necessary, since both are lines per unit area. The answer lies in the fact that the above "definitions" were not rigorous. They are correct, but not complete.

In any magnetic design a magnetic field of given strength is supplied, generally by means of electric current in a solenoid. A ferromagnetic part, the core (usually of iron), is placed in the magnetic field. The induced magnetism is then used to produce a pulling or turning or other action for which the part was designed. How much useful force will be exerted will depend on how many gausses of magnetic flux will be developed by the number of oersteds of field strength available. It follows that, for design purposes, one piece of information needed is the relationship of induced magnetism to the magnetizing force of the solenoid.

The B-H Curve

In magnetic calculations the magnetizing force is denoted by the letter H, and the induced field by B, so the curve is referred to as the B - H curve. The shape of the B - H curve is shown in Fig. 2.

The shape of the curve is explained by the operation of the domain theory. At low magnetizing forces there are plenty of atoms to form domains, and so a little change in magnetizing force induces a certain change in the induced magnetism. At higher magnetizing forces a greater portion of the atoms are already in domains, and the same change in magnetizing force produces less change in the induced magnetism. Finally, the domain structure is complete and further magnetizing force produces no further strength. The value of the induced magnetism at this point is called the saturation value. This value stated in gausses or kilogausses (1000 gauss) is a characteristic of each alloy.

Another important magnetic property is the permeability, which is a measure of the ease with which magnetic lines of force can pass through the metal. This quantity is usually denoted by μ. By definition:

$$\text{Permeability} = \mu = \frac{B}{H} = \frac{\text{Induced magnetic flux}}{\text{Magnetizing field}} \quad (3)$$

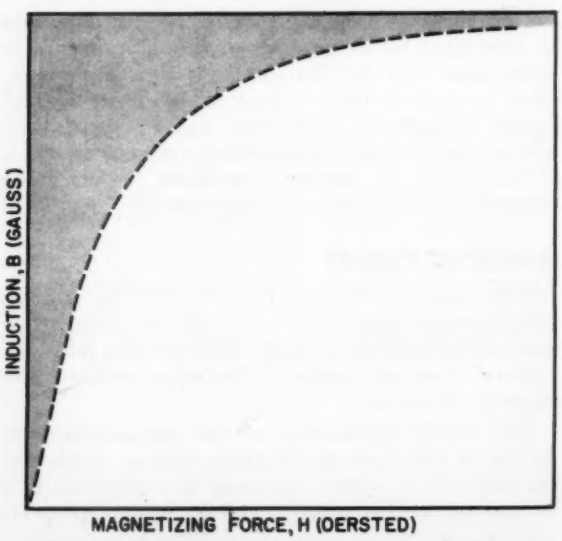


Fig. 2 — The B-H curve used in magnetic calculations. The magnetizing force is denoted by the letter H, and the induced field by the letter B.

So far, the discussion has dealt with the behavior of a magnetic system when a magnetizing force is applied. In any practical use of magnetism the field is applied for a while and then cut off or reversed. For descriptive purposes it is easiest to consider the case where magnetic field is built up in one direction, slowly reduced to zero, built up in the opposite direction, returned to zero and then re-cycled. The entire sequence of events is shown by Fig. 3.

As the magnetizing force is increased to some value H^* oersteds, the induced magnetism will follow a curve, like that of Fig. 2, but when the magnetizing force is decreased the induced strength will not retrace the original curve, but will lag behind so that when H, the magnetizing force, reaches zero there will still be some magnetism in the magnetic material. This is shown by the point B_o. The extent of this retentivity, or retained magnetism, depends on the magnetic material and on the maximum value of the magnetizing force (H^*) which has been used.

Having reached zero the field force can be built up in the opposite direction. The induced magnetic strength will decrease until the magnetizing force reaches a value $-H_o$, where the induced magnetism is reduced to zero. The absolute value of the magnetizing force at this point is the coersive force, which depends again on the metal and on the maximum inducing force used.

As the inducing force goes on to $-H^*$ and back through zero to $+H^*$, the rest of Fig. 3 is traced out.

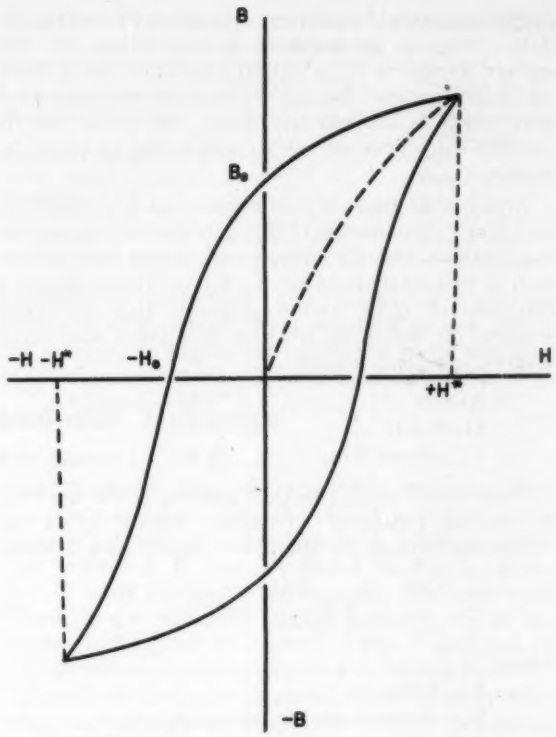


Fig. 3 — Typical hysteresis loop.

Hysteresis Loss

The shape of the loop in Fig. 3 or, more important, its area, depends on the metal and the maximum magnetizing force. Since this area of the loop is so important for efficient operation of some magnetic designs, the hysteresis loss is an important design factor. This is the power lost in the metal, and is proportional to the area of the loop.

Hysteresis loss is usually denoted by the letter h, and the units used are ergs/ccm/cycle. There has been an attempt to relate the hysteresis loss h to the maximum value B of induction in the cycle. For this, it is necessary to use the Steinmetz constant η. The relationship is

$$h = \eta B^n \qquad (4)$$

where the exponent n is about 1.6 for iron and iron-silicon alloys for limited range of magnetization. It is much greater for cast iron. The Steinmetz constant depends on the metal or alloy and runs from 0.0007 for iron-silicon alloy to 0.025 for some steel. Clearly, lower values of the Steinmetz constant are desirable to minimize the hysteresis loss.

Due to limitations of conditions under which the above equation applies the utility of the Steinmetz constant is not great, and it is mentioned here for the sake of completeness.

Another factor of importance in choosing alloys for magnetic work, particularly where alternating current is involved, is the eddy current loss.

It was shown previously that an electric current can produce a magnetic field. In a sense, this process is reversible. If an electrical conductor moves in a magnetic field an electrical current is caused to flow in the conductor. In a generator the magnetic field is stationary and the conductor moves; in an electromagnet the conductor (core) is stationary and the field moves under the action of the alternating current. This causes eddies of electrical current to be set up in the core of the magnet. These currents generate heat and represent a power loss which decreases efficiency.

Eddy Current

The eddy current depends on the area, a design factor, and the resistivity, a material factor. High resistivity is desirable.

Eddy current losses, together with hysteresis loss, make up the core loss which is expressed in watts/lb at a given flux density (10 or 15 kilogausses) at a frequency of 60 cycles/sec. Eddy current losses are proportional to the squares of the thickness, frequency and maximum flux density, and inversely proportional to the resistivity. Thus, the eddy current loss in watts per cc can be

$$P_e = \frac{(\pi\, t\, f\, B_m)^2}{6\, \rho\, 10^{10}} \qquad (5)$$

where

t = thickness of section in cm.
f = frequency in cycles/sec.
B_m = maximum flux density in gausses.
ρ = resistivity in micro-ohm−cm.

Combining this with equation (4) would give the core loss.

There is nothing difficult about using the formulas

given previously. It is merely a question of looking up data, plugging in numbers and grinding out the answer. Equation (5) is usually applied to parts made up of laminations, but can be used for nonlaminated parts also. As the formula shows, the lower the resistivity the higher the eddy current loss, as might be expected.

Armco iron, like all pure metals, has low resistivity and is not recommended for use in alternating current applications. On the other hand, 4 per cent silicon iron is frequently used for such jobs. If we assume a thickness of ¼-in. and a magnetic flux of 10,000 gausses, the following eddy current losses will occur if 60 cycle current is used:

Armco .....5.52 watts/cm²
Malleable 1.85
4% silicon iron.....1.16

Probably of more practical interest is the question of holding power of a magnet. Pierce* gives the following formula for the force exerted by a magnet.

$$F = \frac{B^2A}{72} \times 10^{-6} \qquad (6)$$

where

F = force (lb).
B = magnetic flux (maxwells/in.²).
A = pole area (in.²).

*Mechanical Engineering, vol. 80, no. 4, pp. 64-66, April 1958.

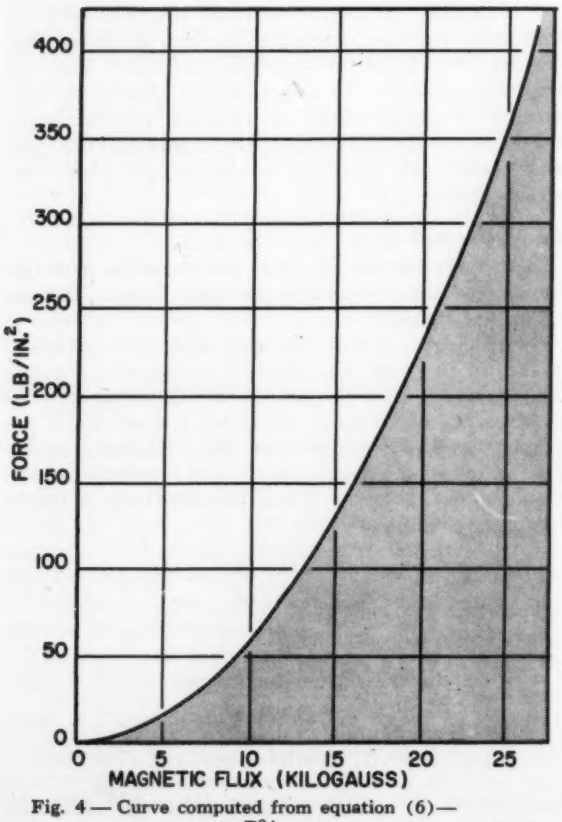


Fig. 4 — Curve computed from equation (6)— $F = \frac{B^2A}{72} \times 10^{-6}$

Here the induced magnetic flux was measured in gausses or kilogausses (1000 gausses) not in maxwells per sq in. Rather than complicate the discussion by introducing new units, the curve of Fig. 4 can be computed from equation (6). Notice that the force is given in lb/sq in. of pole area. In order to show how Fig. 4 can be used, the following illustration will serve.

Example

Suppose a given part is to be made of dynamo steel and of malleable iron. A coil of 100 turns carrying two amperes will energize the mechanism. How large must the pole piece be to exert a force of 150 lb? No air gap is assumed. The coil is 1-in. long.

1) The energizing field will be 200 amp turns which, according to equation (1), is practically 100 oersteds.

2) Using a curve for induced magnetism for each material, the magnetic fluxes can be found. (The required curve for malleable is given in the next section.)

Dynamo steel — 15.70 kilogausses.
Malleable — 14.25 kilogausses.

3) Reference to Fig. 4 gives the following forces:

Dynamo steel — 139 lb/in.²
Malleable — 117 lb/in.²

4) The following areas are needed to produce a force of 150 lb:

Dynamo steel — 1.07 sq in.
Malleable — 1.27 sq in.

An 18 per cent greater area is necessary if malleable is used.

In any electromagnetic application it is desirable to have the force released when the current is shut off. In discussion of the hysteresis loop, it was noted that the magnetic flux does not drop to zero when the magnetizing current drops to zero. This leads to another problem. What is the residual force in the above problem?

Reference to the hysteresis curves of dynamo steel and malleable (also given in the next section) shows the retentivities to be:

Dynamo steel — 10.00 kilogausses.
Malleable — 6.50 kilogausses.

Figure 4 shows the forces corresponding are:

Dynamo steel — 58 lb/in.²
Malleable — 25 lb/in.²

Taking the areas into account, the dragging forces are:

Dynamo steel — 62.5 lb.
Malleable — 29.5 lb.

Figure 4 might also be used for such problems as finding the winding needed to energize the piece given the desired force and the material.

Since this paper is about magnet alloys, examples of calculations of magnetic circuits involving air gaps will not be included. The air gap tends to cut down

the magnetic flux. The ranking of alloys would not change, however, if the calculations are made for circuits with or without air gaps.

METALLURGY OF MAGNETIC ALLOYS

Although nickel and cobalt and some of their alloys have magnetic properties, practically all commercial magnetic materials are iron base.

Pure iron, as represented by armco or electrolytic iron, reacts readily to a magnetic field, and gives a high value of induced magnetism. When the magnetizing current is cut off, the retained magnetism is low. Therefore, for direct current applications pure iron is satisfactory. The electrical resistivity of pure iron is low and, according to equation (5), this means a high eddy current loss in alternating current applications.

In order to overcome this difficulty alloying is indicated. Carbon is the most common alloying element in iron, and carbon will increase the resistivity. Unfortunately, carbon increases the retentivity making the alloy behave like a permanent magnet. It also runs up the Steinmetz constant and, according to equation (4), this means an increase in hysteresis loss. The hysteresis curves of iron-carbon alloys all show the expected increased area. Addition of carbon to iron in effect substitutes one power loss for another. That is why alloys for electromagnets are always low in carbon. The maximum is usually less than 0.10 per cent.

Malleable iron, with its carbon content of about 2.50 per cent, may seem to be an exception. In fully annealed malleable iron all the carbon essentially is in the form of graphite, which is inert magnetically. In effect, all that it does is decrease the active cross-section. The matrix of malleable iron carries all the magnetic lines of force, and the carbon content of this matrix meets the requirements outlined previously.

Iron-Silicon Alloys

Silicon is another common alloying element. It has been found that iron-silicon alloys can be made so that the hysteresis loss and eddy current loss are both low, and still the retentivity and B - H curve are not materially affected. A series of iron-silicon alloys, ranging from 1 per cent to 4 per cent silicon, are commonly used in sheet form for alternating current applications.

Malleable iron, with its silicon content of about 1.25 per cent concentrated in the metallic matrix, fits into this range. Other alloying elements are used for permanent magnets.

Unless a part can be made of sheet or bar stock, there are only two cast alloys in general use. Dynamo steel is low (less than 0.10 per cent) carbon, with about 0.50 per cent silicon. The other cast material is malleable. A comparison of these two cast alloys is given in Table 2. Their magnetic characteristics will be given in detail in the next section.

Current comments can be made on the table. Malleable iron is the only magnetic material which is at the same time an alloy of general application. The magnetic properties of all magnetic alloys vary depending on processing and processing variables. Malleable iron is made by a standard process and is not sensitive to processing variations, so its magnetic properties vary less than those of other alloys.

TABLE 2 — COMPARISON OF CAST MALLEABLE IRON AND DYNAMO STEEL

	Malleable Iron	Dynamo Steel
Magnetic Properties	Relatively constant	Depend on heat treatment
Foundry Characteristics	Good	Poor
Machinability	Good	Poor
Hardness	Soft	Soft
Wear Resistance	Fair to poor	Fair to poor
Availability	Standard alloy	Special alloy

MAGNETIC PROPERTIES

B-H Curve

For malleable iron, like any other magnetic alloy, it is not possible to show the complete picture of its response to magnetic induction with a single graph. The general B - H curve is shown in Fig. 5, which may be used in design calculations. The semi-logarithmic plot is not easily read at its lower end so the arithmetic B - H curve is given in Fig. 6, and can be used in calculations involving small inducing force.

The B - H curves and hysteresis loops of other magnetic materials are seriously affected by processing variables such as rolling or heat treatment. Unless

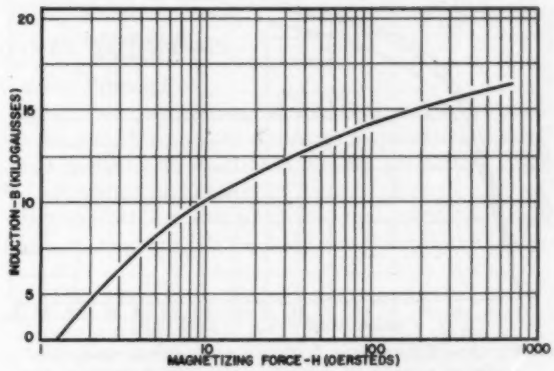


Fig. 5 — General B-H curve for malleable iron which may be used in design calculations.

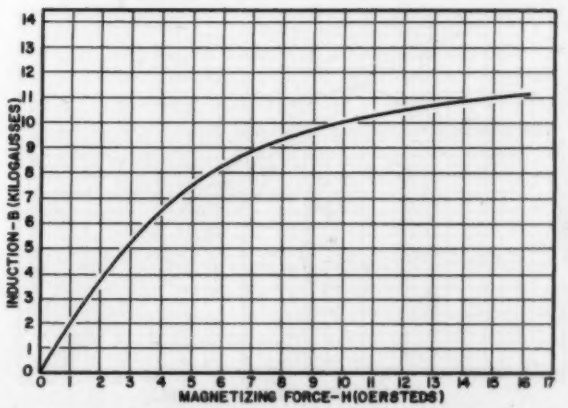


Fig. 6 — Arithmetic B-H curve that can be used in calculations involving small inducing force.

these variables are closely controlled the magnetic properties vary more widely than would be acceptable.

Malleable iron is made for many applications requiring certain properties other than magnetic qualities. In order to obtain the desired properties, a set process with definite controls is followed. When the iron is so produced that it meets the ordinary specifications it will automatically have the best magnetic properties. Since no special handling or treatment is required for malleable iron for magnetic applications, the variability of the product will tend to be less.

Magnetic Variation

There is, of course, some little variation in the magnetic properties of malleable iron due to unavoidable variations in production. However, the effect on magnetic properties is much less than the variability shown by other magnetic alloys. The variation in induced magnetization at low magnetizing forces is practically negligible. At about 30 oersteds, the variation is not more than ±0.5 kilogauss. Even at the highest magnetizing forces the maximum spread is about ±1.5 kilogausses, and this allowance is overgenerous most of the time. Some designers prefer to figure on a range ±1.0 kilogauss.

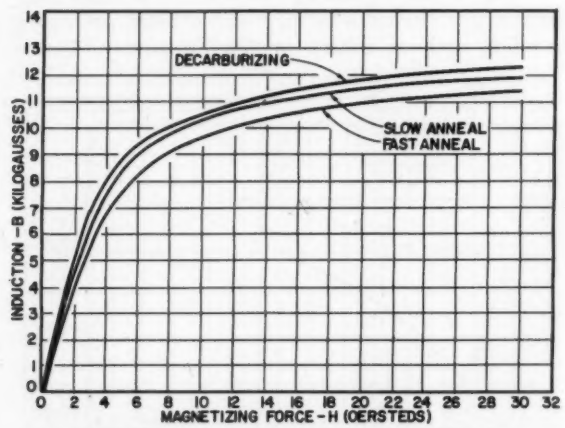


Fig. 7 — Effect of a 2.70 per cent carbon iron, subjected to a series of anneals, which would produce the most extreme annealing conditions possible, on the B-H curve.

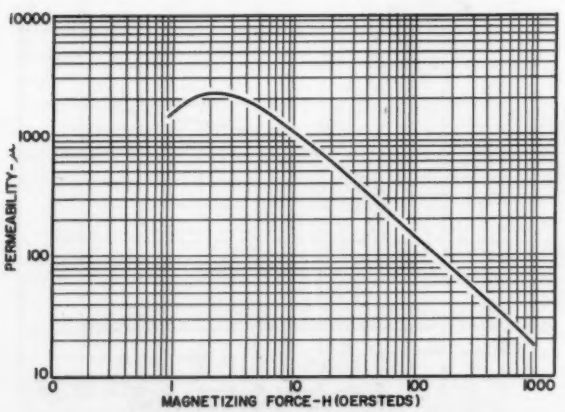


Fig. 8 — Typical malleable iron curve of permeability vs. magnetizing force plotted on log-log paper.

In order to find the sources of the variation in an extended series of studies was made at one time, and it was found that the graphite content and annealing practice were the greatest sources of variability.

At low magnetizing force the effect of carbon will not be noticed. At forces of about 25 oersteds up, the *B* value will decrease about 1.4 kiloguasses for an increase of 1 per cent carbon. The carbon caused variation is independent of the magnetizing force. Since the carbon content of white iron is always controlled within ±0.10 per cent, the contribution of carbon to the variation in induced magnetism is small in commercial malleable iron.

A 2.70 per cent carbon iron (standard malleable iron has about 2.50 per cent carbon) was subjected to a series of anneals which would produce the most extreme annealing conditions possible. The effect on the B - H curve is shown in Fig. 7. The variation in *B* is not great, but the curves do not represent commercial practice. In regular production of malleable iron the variation in annealing practice would be far less, and so the variation in magnetic behavior would be less than that shown in Fig. 7.

Saturation Value

The maximum value of induced magnetism in malleable iron is 18,000 gausses. This value is slightly lower than that for other magnetic alloys for rather obvious reasons. The cross-sectional area of a specimen or part is determined by the dimensions of the piece, but, as a photomicrograph will show, part of this area is taken up by graphite, and the metal which carries the lines of force is less than the total cross-section.

Ordinarily parts are not designed to use the saturation value, and if they are the lower value of malleable iron can be compensated by increasing the cross-sectional area.

Permeability

For any magnetic alloy the curve of permeability vs. magnetizing force plotted on log-log paper rises to a maximum at relatively low magnetizing force. The curve for malleable iron, shown in Fig. 8, is typical.

Since graphite occupies space which does not offer passage to many lines of force, it might be expected that the permeability would be affected by the carbon content. At H = 25 oersteds, an increase of 1 per cent carbon will drop the permeability about 60 points. Since in malleable iron the carbon is allowed to vary only about 0.1 per cent, the variation in permeability will be extremely small.

Hysteresis

One of the advantages of malleable iron for magnetic applications is its low hysteresis loss. Figure 9 shows a typical hysteresis curve. The hysteresis loss depends on the area included in the loop, and this area in turn depends on the maximum magnetization of a cycle. Steinmetz' equation (4) could be used to compute the hysteresis loss at low magnetizations, but it is preferable to measure the loops. Table 3 gives the measured hysteresis losses corresponding to various maximum magnetization.

Another loss encountered in alternating current work is the eddy current loss. One of the quantities needed to compute this loss (equation 5) is the electrical resistivity. For malleable iron this is 32 microohm—cm.

Coercive Force

For any given metal, the coercivity depends on the previous magnetizing force. The greater this force, the greater will be the coercive force for removing the residual magnetism. Coercivity values for malleable iron range from 1 to 3 oersteds for fairly high magnetization.

Retentivity

The retained magnetism is generally undesirable and depends on the previously induced magnetism. Malleable iron is capable of retaining only a small magnetic strength. Figure 10 shows the retentivity as a function of the previously induced magnetism.

Comparison With Other Magnetic Alloys

There is a long list of alloys which have been used for magnetic purposes. Comparison of malleable with each of them is not possible, so only three were selected.

Pure iron (armco or electrolytic iron) is available in sheet form. Its properties depend to a great extent on processing, and those given below are for iron in a state which would be frequently used.

Four per cent silicon iron also is available in sheet and strip. It belongs to the class sometimes called transformer iron. The properties given are not for grain oriented iron.

Dynamo steel is a cast material, and is included because it is the only cast alloy other than malleable iron which is used currently for magnetic applications.

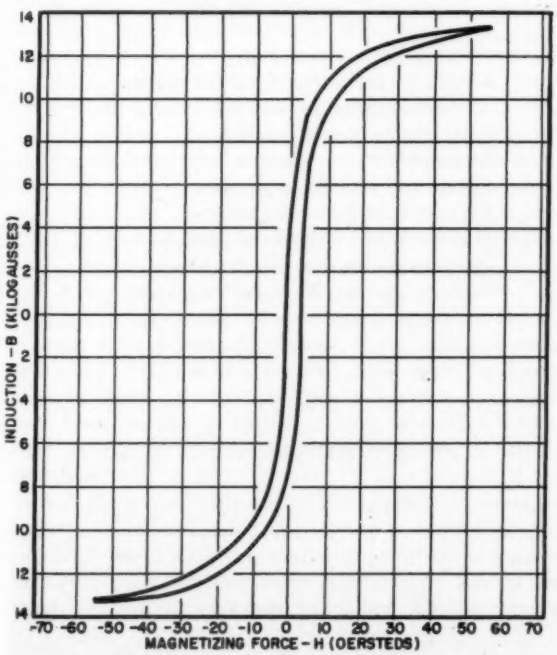


Fig. 9 — Hysteresis curve for malleable iron showing its low hysteresis loss.

TABLE 3 — HYSTERESIS LOSS VS. MAGNETISM

Induced Magnetism B max. (Kilogauss)	Hysteresis Loss (ergs/cc/cycle)
5.45	481
8.63	3752
10.06	4404
11.40	6199
12.70	6862
13.30	7779
14.20	9490
14.60	10542
15.20	11297
15.80	12357

TABLE 4 — MAGNETIC PROPERTIES

Magnetic Property	Malleable	Dynamo	Iron	4% Si Iron
Saturation	18,000	21,400	21,550	19,700
Retentivity (B=15)	6,800	9,000	10,000	—
Max. permeability	2,300	6,000	6,000	7,000
Coercive force	2.0	5.0	1.0	0.5
Hysteresis loss (B=10)	4,400	—	4,000	3,500
Resistivity	32	—	10	60

Some of the magnetic properties of these alloys and malleable iron are given in Table 4.

It can be seen from this comparison that malleable iron has magnetic properties which compare favorably with the properties of other magnetic alloys. No one of these alloys is superior to the rest, and so the final choice of material may depend on other properties which are discussed in the next section.

OTHER PROPERTIES

Tensile Properties

There are two grades of standard malleable iron, but by far the greatest tonnage is produced in the 32510 grade. The minimum tensile properties specified are shown in Table 5.

Dynamo steel in cast form, and most of the wrought forms of magnetic alloys, have tensile properties sim-

TABLE 5 — TENSILE PROPERTIES

Yield Strength, psi	32,500
Tensile Strength, psi	50,000
Elongation, %	10

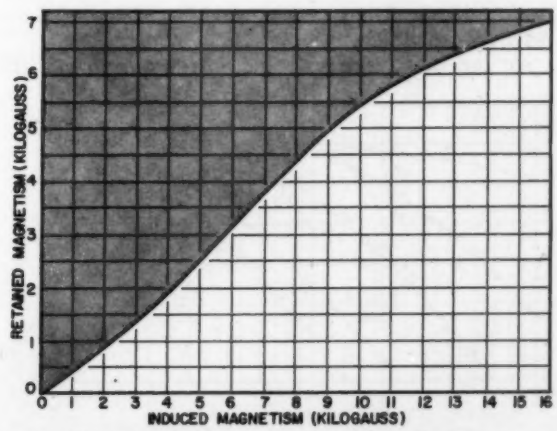


Fig. 10 — Malleable iron retentivity as a function of previously induced magnetism.

ilar to the properties of malleable iron. In fact, design calculations for any of these alloys could be based on a yield strength of 30,000-32,000 psi.

Hardness and Wear Resistance

Since all the magnetic alloys are made up of ferrite, which may or may not contain a little silicon, they are all soft. Malleable iron contains a second phase, graphite, but this does not increase the hardness. All the magnetic alloys have a Rockwell B hardness of approximately 95.

This low hardness means that the wear resistance of magnetic alloys is not particularly good. The hardness of the wrought alloys, and of cast dynamo steel, could be raised by increasing the carbon content and heat treating. The normal heat treatment which produces malleable iron could be changed slightly to produce pearlitic malleable, which is harder than standard malleable. These changes would produce more wear resistant metals, but would do so at the expense of the magnetic properties.

Any designer who has a magnetic application should recognize that the magnetic alloys from which he must choose are all soft, and if wear is a problem he must provide means of protecting against wear.

Malleable iron is the only magnetic material which will respond at all to flame hardening, but the result is obtained by recombining carbon and so the magnetic properties will suffer. For the same reason, carburizing is not recommended.

The use of hard steel wear plates is not a good idea because they form part of the magnetic circuit and act like permanent magnets, and increase core loss and retentivity.

The best solution is to cover the surface with a nonmagnetic, wear resistant material like a hard chromium electroplate.

Foundry Characteristics

There are designs which, because of their shape, are best made as castings. If the alloy must be a magnetic material, the choice would be between malleable iron and cast dynamo steel.

The foundry characteristics of the two metals are quite different. Malleable iron can be produced in much thinner sections than cast dynamo steel. If the casting is not too long malleable iron can be cast $\frac{1}{8}$-in. thick, and the dynamo steel would need to be about twice that thickness.

At the other extreme, there is no limit on the thickness which can be cast in dynamo steel, but malleable is limited by the occurrence of mottling in large sections. However, sections which would mottle are larger than would generally be used for magnetic work, because the eddy currents would be quite high in such large sections.

Aside from the foundryman's problems in dealing with dynamo steel, there is another advantage to malleable iron. The production of malleable iron is a standard thing. Dynamo steel, in contrast, is a tailored steel which requires some special handling.

Machinability

Whether the design is cast or cut from bar stock, there is likely to be some machining. Casting minimizes machining, but even so machinability is important.

Pure iron, the silicon irons and cast dynamo steel have one thing in common—their machinability is terrible. They tend to load the tool and to tear.

Malleable iron, on the other hand, is one of the most easily machined alloys available. It is much easier to machine than the free cutting steels like S.A.E. 1112, which are often used as standards of machinability.

CONCLUSIONS

As far as strength and hardness is concerned, all the magnetic alloys are alike. Malleable iron has the advantage of being a cast form, readily available and easily machined.

APPENDIX

Although the units used in this bulletin are commonly used in magnetic work, there are other units which are preferred by some workers. The following conversion table is given so that the confusion of including these other units in the discussion could be avoided.

Ampere – turns/cm $\times$ 1.257 = oersteds
Ampere – turns/in. $\times$ 0.495 = oersteds
Gilberts/cm $\times$ 1 = oersteds
Lines/cm^2 $\times$ 1 = gausses
Lines/in.^{2} $\times$ 0.155 = gausses
Maxwells/cm^2 $\times$ 1 = gausses
Maxwells/in.^{2} $\times$ 0.155 = gausses
Webers/cm^2 $\times$ 1 $\times$ 10^8 = gausses
Webers/in.^{2} $\times$ 1.55 $\times$ 10^7 = gausses

ELECTRICAL CONDUCTIVITY OF SAND-CAST COPPER-BASE ALLOYS

By D. G. Schmidt and F. L. Riddell

ABSTRACT

The electrical conductivities of over 100 various sand-cast copper-base alloys have been determined. Since many tensile specimens were being tested daily, the standard 0.505 in. diameter test specimen was used for the electrical conductivity measurements. Data list electrical conductivities for the standard American Society for Testing Materials cast copper-base alloys, plus many other cast copper-base alloys. The effect of alloying elements on the electrical conductivities of the various classes of sand-cast copper-base alloys is illustrated.

INTRODUCTION

Copper or alloyed copper castings are used in the electrical industry for their current carrying characteristics. However, sound copper castings, with a minimum of 85 per cent I.A.C.S. electrical conductivity, are difficult to make. The ordinary deoxidizers, such as silicon, aluminum, zinc and phosphorus, cannot be used, because residual amounts lower the electrical conductivities drastically (Fig. 1).

Cast copper is soft and low in strength. Improved mechanical properties with good conductivity (40 to 80 per cent I.A.C.S.) may be obtained with heat treated alloys containing silicon, cobalt, chromium, nickel and beryllium in various combinations. However, these alloys are expensive and less readily available than the standard copper-base foundry alloys, and due to the highly oxidizable nature of their alloying elements (silicon, chromium and beryllium), extra care is required in melting and pouring.

In many instances, where design permits the use of lower electrical conductivities, the standard copper-base foundry alloys may be used. Numerous inquiries, as to the electrical conductivities of the various copper-base foundry alloys, were received, and the writers were unable to give factual data from the available literature.

The *Brass and Bronze Ingot Institute Manual*[2] lists the electrical conductivities for several of the standard alloys, but the authenticity of these figures has never been verified. Some conductivity figures are listed in various manuals published by the manufacturers of copper-base alloys (both in ingot form and as castings). However, no data as to type of specimen, exact analyses or other pertinent information are listed.

Smith and Palmer[3] report the electrical and thermal conductivities of nine various cast copper alloys.

Kura and Lang[4] report the electrical conductivities of three alloys— 85 Cu, 5 Sn, 5Pb, 5Zn; 88 Cu, 6 Sn, 2 Pb, 4 Zn and 80 Cu, 10 Sn, 10 Pb — at five different temperatures. Due to the scarcity of authenticated information in the literature, the authors' company's research laboratory started a program to determine the electrical conductivity of the many commercial copper-base foundry alloys.

RESISTANCE MEASURING EQUIPMENT

A general purpose Kelvin bridge was used to make the resistance measurements. This Kelvin bridge has a range of 0.00001 to 1 ohms, without estimating fractions of scale divisions. By adding a vernier scale to the equipment, readings to the nearest 0.000001 ohm were possible. The resistivity measurements used for this report are estimated to be accurate within ± 0.5 per cent.

Since the measurement of electrical conductivity is a daily routine test in the company's research laboratory, a constant temperature oil bath was not used. Instead, the test specimens were allowed to reach ambient temperature before testing, with the temperature being recorded for each test, and a temperature correction made in the calculations.

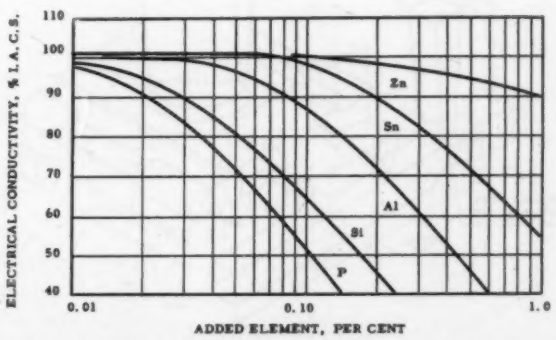


Fig. 1 — Effect of alloying elements on electrical conductivity of copper, ref. 1.

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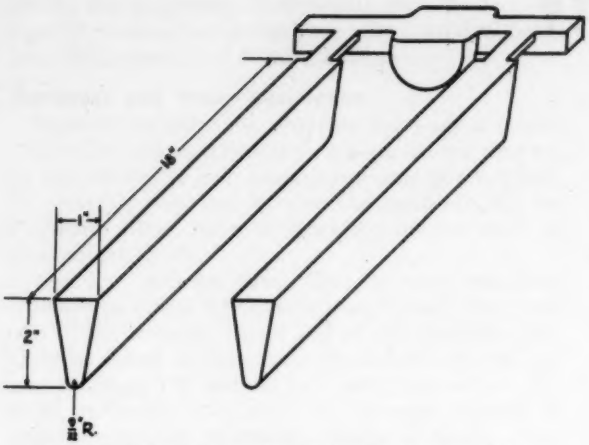


Fig. 2 — Wedge bar casting used for 18 in. x 3/8-in. diameter conductivity specimen.

The formula used for determining the electrical conductivity was:

$$C = \frac{0.017241}{\frac{R\,D^2\pi}{L\,4} + 0.000068(20\text{-}t)} \times 100$$

where:

C = electrical conductivity expressed as a per cent of the International Annealed Copper Standard.

0.017241 = the resistance in ohms of the International Annealed Copper Standard; one meter long and 1 sq mm in cross-section at 20 C (68 F).

R = resistance of the sample between potential points.

D = diameter of the sample in mm.

L = distance between potential points in meters.

t = temp. in degrees C of the sample when the resistance is measured.

0.000068(20-t) = temperature correction for reducing resistance at the temperature of measurement to resistance at 20 C (68 F).

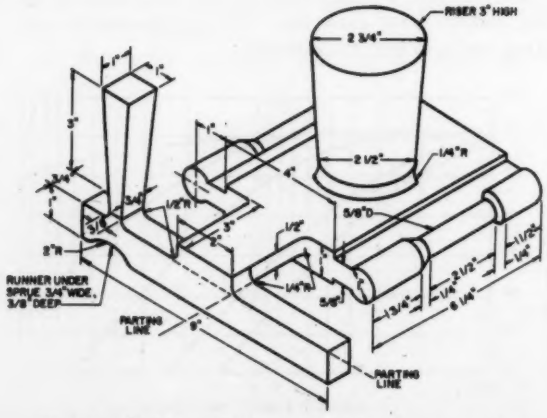


Fig. 4 — Double horizontal full-web type test bars (A.S.T.M. B208-49T).

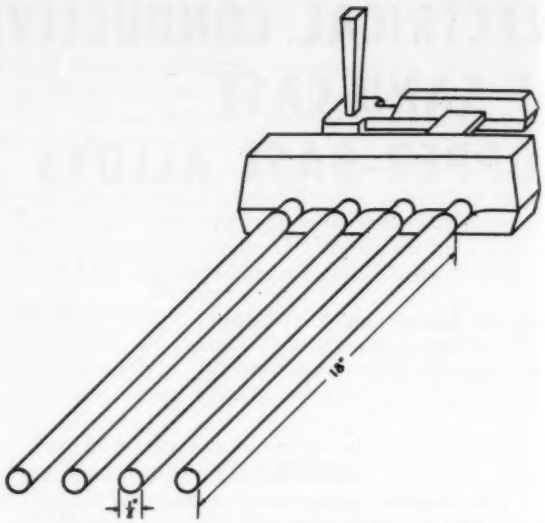


Fig. 3 — Rod bar casting used for 18 in. x 3/8-in. diameter conductivity specimen.

TEST SPECIMEN SELECTION

In order to obtain relatively high resistance readings, a long bar of small cross-section was originally used. The first test bars poured were wedges, 18 in. long, 2 in. deep and 1 in. across the top, with a 9/32-in. radius on the bottom (Fig. 2). The test specimen was then cut from the bottom portion of the casting and machined to a 0.350 to 0.375 in. diameter rod. This wedge-type bar gave results that were not reproducible from one heat to another of the same material. In addition, the test specimen was unwieldy, as far as removing from the wedge casting and then machining.

Figure 3 shows the second type test casting used. This was a 1/2-in. diameter rod, 18 in. long, which was then machined to 0.350 to 0.375-in. diameter rod. Most of these rod castings exhibited center-line shrinkage, and it was felt that for consistent and accurate results, the test specimens should be as sound as possible.

Since one of the functions of the research laboratory is to determine mechanical properties of production heats made for shipment, it was felt that these same tensile test bars could possibly be used for the electrical conductivity measurements. Sound test bars could readily be obtained from the horizontal 5/8-in. Web-Webbert test bar casting, Fig. 4 (Fig. 3 A.S.T.M. B-208-56). This bar is machined to 0.505-in. diameter with a 2 1/2-in. length of uniform cross-section.

To determine if the standard 0.505 in. diameter bar would prove satisfactory, a 3/4-in. diameter and 12 in. long piece of wrought brass, of the following composition was tested for conductivity.

Copper,%	56.88
Tin,%	0.07
Lead,%	0.75
Iron,%	1.13
Nickel,%	1.37
Zinc,%	Remainder

From this 12 in. long rod, two standard 0.505 in.

diameter tensile test bars were machined, and the conductivity measured over a 2-in. length. Table 1 lists the results of this test.

From the close agreement between the conductivity of the longer bar and the standard 0.505 in. diameter test bar, it was decided that electrical conductivity measurements could be accurately determined on the standard 0.505 in. diameter tensile test specimen.

FOUNDRY PROCEDURE

The melting charges consisted of all ingot, or ingot plus foundry returns of known composition. The charges were melted in a No. 60 crucible, in a gas-fired crucible furnace. In all heats, the atmosphere above the melt was of an oxidizing nature.

In the case of the copper/tin/lead/zinc alloys, 15 per cent phosphor copper was used for deoxidation just prior to pouring (usually 2 oz per 100 lb of metal). Zinc additions were made to those alloys requiring replacement for that lost on melting. A proprietary alloy for deoxidation and fluidization of the nickel silvers was used. No special additions were required for the aluminum or silicon bronzes.

A match plate of the test bar pattern, Fig. 4 (Fig. 3 A.S.T.M. B-208-5), was used to make the molds. The molds were made from a natural green sand, having the following average properties:

Green Strength, psi	5.5 - 6.0
Permeability	35 - 40
Moisture, %	5.5 - 6.5

The melts were poured at temperatures consistent with good practice for each alloy, to obtain optimum properties. In almost all cases, the test bar castings were shaken out one-half hour after pouring.

DISCUSSION OF RESULTS

The electrical conductivities, as determined by the research laboratory, and by other investigators, are listed in Table 2. The alloys listed in this table are those standard copper-base alloys, both in ingot form

TABLE 1 — CONDUCTIVITY TEST RESULTS

	0.75 in. diameter bar	0.505 in. tensile bar	0.505 in. tensile bar
Temp., C (F)	24.5 (76.1)	23.3 (73.9)	23.3 (73.9)
Diameter, mm	19.0	12.8	12.8
Length, meters	0.18	0.051	0.051
Resistance, ohms	0.000058	0.000036	0.000036
I.A.C.S., %	18.91	18.95	18.95

TABLE 2 — ELECTRICAL CONDUCTIVITY OF THE STANDARD A.S.T.M. COPPER-BASE ALLOYS

	A.S.T.M. Specification			% I.A.C.S.				
Class	Alloy	Ingot	Castings	H.K. & Co.	B.B.I.I. Manual	Mfg. Literature	Smith and Palmer	Kura and Lang
Tin bronze	88-10-0-2	B30 (1A)	B143 (1A)	10.7		10-12	10.9	
	88-8-0-4	(1B)	(1B)	12.3		14-16	11.6	
Leaded tin bronze	88-6-2-4	B30 (2A)	B143 (2A)	13.9	13-15	13-15		14.3
	87-8-1-4	(2B)	(2B)	12.3		10-12		
High lead tin bronze	80-10-10	B30 (3A)	B144 (3A)	11.0		7-12		10.1
	83-7-7-3	(3B)	(3B)	12.3		10-14		
	85-5-9-1	(3C)	(3C)	14.9				
	78-7-15	(3D)	(3D)	11.7			9.8	
Leaded red & semi-red brass	85-5-5-5	B30 (4A)	B145 (4A)	15.0	13-16	13-16		15.1
	83-4-6-7	(4B)	(4B)	15.3		16.0		
	81-3-7-9	(5A)	(5A)	16.5	18.0	18.0		
	76-3-6-15	(5B)	(5B)	16.6				
Leaded yellow brass	72-1-3-24	B30 (6A)	B146 (6A)	18.7	15-22	15-22		
	68-1-3-28	(6B)	(6B)	19.5	18-25	18-25		
	63-1-1-35	(6C)	(6C)	21.9	20-26	20-24	23.4	
Leaded high strength yellow brass	60,000# T.S.	B30 (7A)	B147 (7A)	18.5	20-24			
High strength yellow brass	65,000# T.S.	B30 (8A)	B147 (8A)	22.7	16-20	16-20	21.6	
	90,000# T.S.	(8B)	(8B)	7.9		12-14		
	110,000# T.S.	(8C)	(8C)	8.0	10-14	10-14		
Aluminum bronze	88-3-9	B30 (9A)	B148 (9A)	12.5	12-14			
	89-1-10	(9B-AC)*	(9B-AC)*	15.0	12-15	12-15		
		(9B-HT)**	(9B-HT)**	13.5	13.8	13.8		
	86-4-10	(9C-AC)*	(9C-AC)*	13.7	12.0	12.0		
		(9C-HT)**	(9C-HT)**	12.0				
	81-4½-10-4½	(9D-AC)*	(9D-AC)*	9.3				
		(9D-HT)**	(9D-HT)**	8.5				
Leaded nickel bronze	55-2-9-22-12	B30 (10A)	B149 (10A)	5.7	5-7		5.9	
	64-4-4-8-20	(11A)	(11A)	5.4	4.5-5.5		5.0	
	66-5-1-3-25	(11B)	(11B)	4.7	4.5			
Silicon bronze			B198 (12A)	6.0		6.0	4.9	
Silicon brass			B198 (13A)	6.5			6.1	
			(13B)	5.5		8.0		

*Sand cast.
**Heat treated — 1650 F (905 C), 2 hr, water quenched, followed by 1100 F (595 C), 1 hr, water quenched.

and as castings, as specified by the A.S.T.M. In most cases, there was agreement between the electrical conductivities determined in this work and those determined by other investigators.

The electrical conductivities of the various classes of sand-cast copper-alloys are listed in Table 3 to Table 10, inclusive.

In a paper by A. L. Norbury,[5] where the effects of several elements on the electrical conductivity were determined, it was found that the specific resistivity was raised approximately 1.3 times for each 1.0 per cent of nickel or tin added to copper. Since tin and nickel have approximately the same effect on electrical conductivity, the data from Tables 3, 4, and 6 are illustrated in Fig. 5, where electrical conductivity is plotted against the per cent tin plus nickel.

For the alloys considered in Fig. 5, the effects of zinc, lead and the minor elements appear to be relatively small, as most of the points lie on or close to the average curve. As the tin and nickel are in

TABLE 3 — ELECTRICAL CONDUCTIVITY OF SAND-CAST TIN BRONZES AND LEADED TIN BRONZES

Alloy Nominal Composition	Composition, per cent								Average % I.A.C.S.
	Cu	Sn	Pb	Zn	Fe	Sb	Ni	P	
89-11-0-0	89.0	10.6	trace	trace	0.03	trace	trace	0.24	9.6
88-10-2-0	87.5	9.3	2.2	0.3	0.02	0.09	0.4	0.02	11.0
88-10-0-2	86.8	10.1	0.2	2.5	0.06	0.03	0.3	0.02	10.9
88-8-0-4	87.6	8.2	0.15	3.8	0.08	0.03	0.1	0.01	12.4
88-6-2-4	88.3	6.0	1.8	3.1	0.08	0.10	0.6	0.01	13.8
88-5-2-5	87.0	5.3	2.2	4.5	0.15	0.03	0.5	0.01	14.1
87-11-1-0-1 (Ni)	87.6	10.2	1.0	0.3	0.01	0.03	0.9	0.01	11.1
87-11-1-0-1 (Ni)	87.0	10.5	1.0	0.3	0.01	0.03	0.9	0.19	10.1
87-11-1-0-1 (Ni)	86.0	10.6	1.1	0.7	0.10	0.03	1.0	0.31	9.2
87-11-0-1-1 (Ni)	85.9	11.5	0.2	1.0	0.03	0.03	1.3	0.01	10.1
87-10-1-2	86.7	9.7	0.9	2.1	0.10	0.10	0.4	0.02	10.8
87-10-2-1	86.4	9.7	1.6	1.6	0.03	0.10	0.5	0.01	11.0
87-8-1-4	87.5	8.0	0.7	3.3	0.15	0.10	0.2	0.02	12.3
88-5-0-2-5 (Ni)	87.1	5.4	0.01	2.4	0.03	0.02	5.1	0.01	11.5
88-5-0-2-5 (Ni)	(Cooled in sand to room temp.)								11.9
88-5-0-2-5 (Ni)	(H.T. — 1400 F - 4 hr - oil quench + 600 F - 5 hr - air cool)								14.8
87-5-1-2-5 (Ni)	86.5	5.1	1.1	2.1	0.10	0.02	5.0	0.02	12.0
87-5-1-2-5 (Ni)	(H.T. — 1400 F - 5 hr - air cool + 600 F - 7 days - air cool)								15.7
85-9-1-0-5 (Ni)	83.6	9.2	1.2	0.4	0.1	0.02	5.2	0.01	10.3
84-16-0-0	83.9	15.4	0.05	0.3	trace	0.02	0.02	0.01	8.5

TABLE 4 — ELECTRICAL CONDUCTIVITY OF SAND-CAST HIGH LEAD TIN BRONZES

Alloy Nominal Composition	Composition, per cent								Average % I.A.C.S.
	Cu	Sn	Pb	Zn	Fe	Sb	Ni	P	
87-4-8-1	86.4	4.1	8.0	0.9	0.02	0.15	0.30	0.01	16.4
85-5-9-1	83.4	4.5	9.8	1.4	0.03	0.20	0.50	0.02	14.9
84-8-8-0	83.3	7.4	8.1	0.5	0.02	0.20	0.25	0.01	11.8
84-4-8-4	84.4	4.0	8.3	2.7	0.10	0.10	0.40	0.01	16.9
83-7-7-3	82.8	6.8	7.5	2.1	0.10	0.15	0.60	0.01	12.4
81-8-9-0-2 (Ni)	82.2	7.0	9.0	0.2	0.01	0.20	1.25	0.01	12.1
80-10-10	79.7	8.8	10.1	0.7	0.01	0.20	0.35	0.01	11.0
78-7-15	77.4	6.8	14.5	0.7	0.03	0.30	0.25	0.02	11.6
75-3-20-0-2 (Ni)	75.0	3.4	18.4	0.3	0.05	0.15	2.20	0.01	14.2
75-13-10-0-2 (Ni)	74.8	13.1	9.4	0.5	0.02	0.15	2.00	0.01	8.6
73-5-22	73.1	4.2	21.9	0.1	0.01	0.05	0.50	0.01	14.1
66-2-32	66.1	1.9	31.0	0.1	0.01	0.05	0.50	0.01	17.8

TABLE 5 — ELECTRICAL CONDUCTIVITY OF SAND-CAST LEADED RED BRASS AND LEADED SEMI-RED BRASS

Alloy Nominal Composition	Composition, per cent								Average % I.A.C.S.
	Cu	Sn	Pb	Zn	Fe	Sb	Ni	P	
93-1-2-4	93.1	0.9	2.5	3.0	0.05	0.05	0.30	0.01	32.4
85-5-5-5	84.6	4.5	5.3	4.6	0.10	0.15	0.65	0.02	15.0
83-4-6-7	82.4	3.8	6.4	6.5	0.20	0.20	0.50	0.02	15.2
83-3-3-11	82.9	3.0	2.9	10.1	0.20	0.15	0.70	0.01	16.7
81-3-7-9	81.2	2.6	7.2	8.0	0.25	0.15	0.50	0.02	16.6
80-5-5-5-5 (Ni)	79.9	4.8	5.4	4.6	0.30	0.15	4.80	0.02	11.1
78-3-7-11-1 (Ni)	79.7	2.6	6.2	10.0	0.25	0.15	1.10	0.01	16.0
76-3-6-15	74.6	3.2	7.3	14.3	0.15	0.10	0.30	0.01	16.6
76-2½-6½-15	76.7	2.4	6.5	13.6	0.30	0.10	0.35	0.02	17.7
76-2-6-16	75.3	2.1	6.9	15.0	0.15	0.10	0.35	0.03	19.0
76-1-6-17	75.9	1.1	7.6	14.8	0.15	0.10	0.30	0.02	21.3

TABLE 6 — ELECTRICAL CONDUCTIVITY OF SAND-CAST YELLOW BRASS

Alloy Nominal Composition	Composition, per cent							Average % I.A.C.S.
	Cu	Sn	Pb	Fe	Sb	Ni	Others	
72-1-5-22	72.8	1.6	4.7	0.4	0.2	0.6		18.6
68-1-3-28	68.1	1.0	2.3	0.3	0.1	0.2		19.6
64-0-0-35-1	64.5	0.05	0.1	0.1	trace	trace	1.1 Si	15.1
63-1-1-35	61.9	0.6	1.0	0.2	0.05	0.3		22.0
63-1-1-35	61.8	0.7	1.1	0.2	0.05	0.1	0.25 Al	21.8
60-1-0-38-1	59.9	1.0	0.04	0.01	0.05	0.3	1.15 Al	23.7
60-0-3-37	61.5	0.1	2.8	0.10	trace	trace	0.06 Al	25.7
60-0-1-38-1	59.5	trace	1.0	0.30	0.01	0.05	1.1 Al	26.5
60-0-0-40	60.9	0.05	0.1	0.4	trace	0.05	0.8 Al	24.9
60-0-2-38	58.7	0.05	2.2	0.1	trace	0.02		26.4
58-1-1-40	58.6	1.0	0.8	0.5	0.01	0.10	0.5 Al	23.3
52-0-0-48	52.4	0.1	0.05	0.01	trace	trace	0.4 Al	35.8

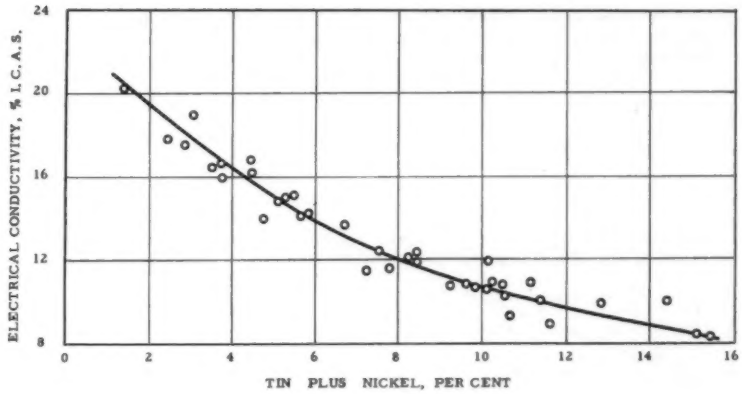


Fig. 5 — Effect of tin plus nickel on the electrical conductivity of tin bronzes, leaded tin bronzes, high lead tin bronzes, leaded red brasses and leaded semi-red brasses.

creased, the electrical conductivity decreases. The data for the 87-11-1-0-1 alloy show that as the phosphorus is increased, the conductivity decreases (Fig. 6).

The electrical conductivities of various sand-cast yellow-brass alloys are listed in Table 6. In the range of copper used in the yellow brasses, i.e., 50 to 75 per cent copper, Fig. 7 shows that as the copper is increased, the electrical conductivity decreases. Here again, tin shows the effect of lowering the conductivity. It is surprising to note that aluminum, which is normally added to many of the yellow brasses for increased fluidity, has not appreciably affected the conductivity.

Table 7 lists the electrical conductivity of the sand-cast high strength yellow brasses. Manganese bronzes, Table 11 and Fig. 8 show the effect of copper on the electrical conductivity of a manganese bronze type alloy. For this alloy, as the copper is increased, the conductivity decreases.

The electrical conductivities for the sand-cast aluminum bronzes, both in the as-cast and heat treated conditions, are listed in Table 8. Heat treating at 1650 F (905 C) for 2 hr, water quenching, followed by 1100 F (595 C) for 1 hr and water quenching, lowers the electrical conductivity. Table 12 and Fig. 9 show the effect of aluminum on the electrical conductivities of three standard sand-cast aluminum bronzes. Within the range of aluminum used, as the aluminum content is increased, the electrical conductivity is also increased.

Table 9 lists the electrical conductivity for the sand-cast nickel-silver and copper-nickel alloys. Due to the many different combinations of alloying elements used in these alloys, little correlation of data is possible. The trend appears to be that as nickel increases, electrical conductivity decreases.

Table 10 lists the electrical conductivities for the sand-cast silicon-bronzes and silicon brasses. Table 13 and Fig. 10 show the effect of silicon on the electrical conductivity of three standard copper-silicon alloys. Within the range of silicon used for these alloys, an increase in silicon decreases electrical conductivity. An increase of approximately 10 per cent zinc has not appreciably lowered the electrical conductivity of the

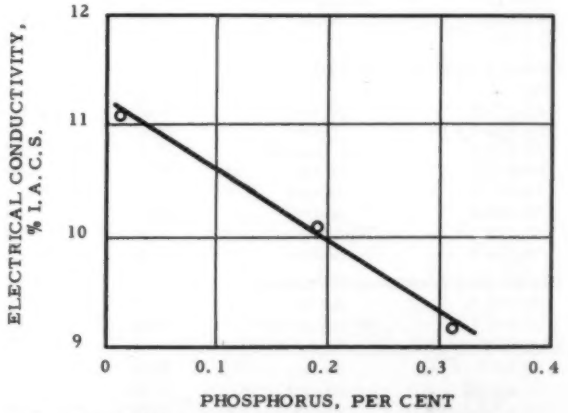


Fig. 6 — Effect of phosphorus on the electrical conductivity of 87-11-1-0-1 leaded tin bronze.

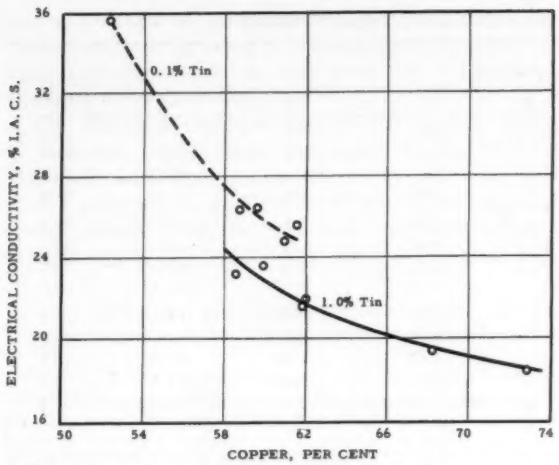


Fig. 7 — Effect of copper and tin on the electrical conductivity of yellow brass alloys.

Fig. 8 — (Right) Effect of copper on the electrical conductivity of a manganese bronze alloy.

TABLE 7 — ELECTRICAL CONDUCTIVITY OF SAND-CAS HIGH-STRENGTH YELLOW BRASS (MANGANESE BRONZE)

Alloy	Composition, per cent					Avera % I.A.C
	Cu	Fe	Al	Mn	Others	
60,000 psi T.S.	58.8	0.9	0.7	0.4	0.6 Sn 0.8 Pb	19.3
65,000 psi T.S.	58.4	1.0	1.0	0.25		21.9
Nickel Manganese Bronze	53.9	1.6	1.3	3.2	3.2 Ni	9.1
80,000 psi T.S.	58.5	1.6	1.7	1.0		16.7
90,000 psi T.S.	59.1	1.6	2.2	1.7		14.5
90,000 psi T.S.	64.0	2.2	3.9	4.0		7.4
90,000 psi T.S.	58.4	2.0	3.1	0.1		24.0
90,000 psi T.S.	66.8	2.5	5.4	3.9		7.4
110,000 psi T.S.	62.5	2.7	6.0	3.8		7.9

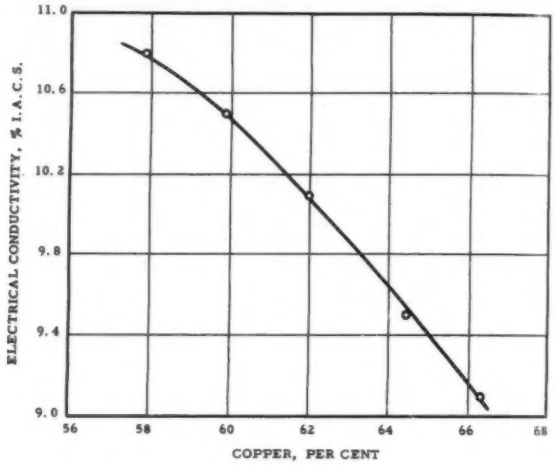


TABLE 8 — ELECTRICAL CONDUCTIVITY OF SAND-CAST ALUMINUM BRONZES

Alloy Nominal Composition	Condition of Bars	Composition, per cent					Average % I.A.C.S.
		Cu	Fe	Al	Mn	Ni	
Copper-Aluminum							
95-5	As-cast	94.8	0.01	5.1	trace	0.03	17.0
90-10	As-cast	88.8	0.10	11.0	0.01	trace	13.6
88-12	As-cast	87.9	0.10	11.8	0.01	0.05	20.3
Copper-Iron-Aluminum							
90-1-9	As-cast	89.1	1.5	9.2	0.10	0.10	12.9
89-1-10	As-cast	88.2	1.4	10.1	0.01	0.03	15.1
89-1-10	Heat treated*	88.2	1.4	10.1	0.01	0.03	12.7
88-3-9	As-cast	87.4	3.4	8.9	0.06	0.20	12.2
86-4-10	As-cast	85.9	3.4	10.4	0.06	0.05	14.6
86-4-10	Heat treated*	85.9	3.4	10.4	0.06	0.05	12.4
84-4-12	As-cast	84.4	3.5	11.8	0.05	0.05	16.8
81-5-14	As-cast	80.8	4.8	14.0	0.20	0.05	10.8
Copper-Iron-Aluminum-Nickel							
88-1-10-1	As-cast	88.1	0.8	9.9	0.01	1.1	13.4
87-1-10-2	As-cast	87.1	0.9	9.8	0.06	2.1	12.2
84-4-10-2	As-cast	83.4	3.9	10.4	0.1	2.1	11.0
84-4-10-2	Heat treated*	83.4	3.9	10.4	0.1	2.1	10.2
81-3-11-5	As-cast	81.5	2.9	10.5	0.1	4.9	9.4
81-4-11-4	As-cast	81.6	4.0	10.4	0.1	3.8	10.3
80-5-10-5	As-cast	79.5	4.8	10.2	0.1	5.2	8.9
80-5-10-5	Heat treated*	79.5	4.8	10.2	0.1	5.2	8.4
76-5-14-5	As-cast	76.7	4.4	14.0	0.2	4.6	12.6
Copper-Iron-Aluminum-Manganese							
85-3-11-1	As-cast	85.3	2.6	10.7	1.1	0.05	10.5
85-3-11-1	Heat treated*	85.3	2.6	10.7	1.1	0.05	9.4
Copper-Iron-Aluminum-Nickel-Manganese							
80-5-9-5-1	As-cast	78.9	4.5	9.6	1.4	5.4	6.9
79-5-9-5-2	As-cast	78.9	4.8	9.1	2.1	4.8	6.5
78-5-9-5-3	As-cast	77.3	4.8	9.6	3.2	4.9	5.8

*Heat Treatment — 1650 F (905 C), 2 hr, water quench, plus 1100 F (595 C), 1 hr, water quench.

TABLE 9 — ELECTRICAL CONDUCTIVITY OF SAND-CAST NICKEL SILVERS AND COPPER-NICKEL ALLOYS

Alloy Nominal Composition	Composition, per cent							Average % I.A.C.S.
	Cu	Sn	Pb	Zn	Fe	Ni	Others	
9% Nickel Silver	50.6	1.4	0.1	35.3	1.1	10.5	1.0 Al	9.4
12% Nickel-Copper	84.8	0.01	0.01	0.01	0.9	12.0	1.2 Al 0.7 Mn	10.1
12% Nickel Silver	64.5	2.6	6.3	14.3	0.6	11.4		6.5
12% Nickel Silver	54.7	1.9	9.9	19.7	1.2	12.4		5.7
15% Nickel Silver	63.1	2.2	7.3	10.5	0.5	16.0		5.4
18% Nickel-Copper-Zinc	64.8	trace	trace	8.5	1.0	17.7	8.0 Al	9.3
20% Nickel Silver	65.0	3.7	3.8	6.0	1.0	20.2		5.0
20% Nickel-Copper-Zinc	57.6	0.3	0.5	21.0	1.2	19.5	0.25 Al	4.7
23% Nickel-Copper-Tin	63.6	10.2	0.05	1.8	0.8	23.4		5.6
25% Nickel Silver	59.5	1.3	2.0	10.9	1.6	24.4		4.2
25% Nickel Silver	65.7	4.7	0.9	2.8	0.8	25.0		4.6
28% Nickel-Copper	66.8	trace	trace	trace	4.3	28.1	0.6 Mn	5.2
30% Nickel-Copper	68.8	trace	trace	trace	0.5	29.3	0.7 Mn 0.7 Si	4.6

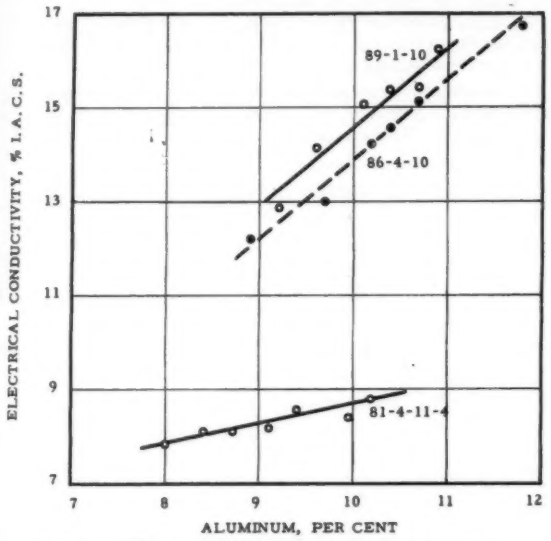


Fig. 9 — Effect of aluminum on the electrical conductivity of aluminum bronze alloys.

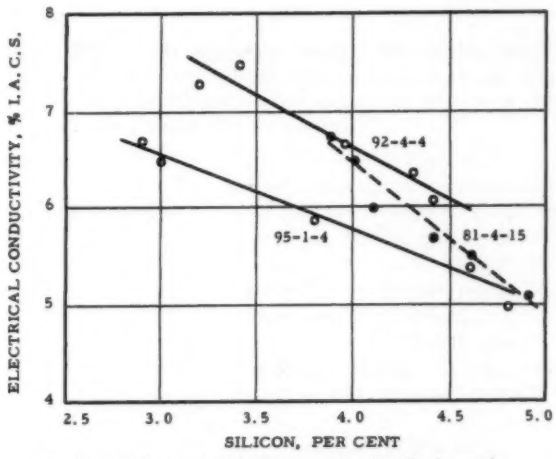


Fig. 10 — Effect of silicon on the electrical conductivity of silicon bronze alloys.

TABLE 10 — ELECTRICAL CONDUCTIVITY OF CAST SILICON BRONZES AND SILICON BRASSES

Alloy Nominal Composition	Composition, per cent						Average % I.A.C.S.
	Cu	Fe	Mn	Si	Zn	Others	
96-1-3	95.8	0.15	1.1	3.0	0.1		6.5
95-1-4	94.7	0.2	1.1	3.8	—		5.9
95-1-4	95.0	0.1	—	3.7	0.3	0.8 Sn	6.6
92-4-4	92.0	0.1	—	4.4	3.5		6.1
91-4-3-1½	90.4	1.2	—	3.2	4.6	0.4 Al	7.4
91-2-7	90.5	0.1	—	2.2	0.2	7.1 Al	8.8
86-2-7-5	85.9	0.1	0.02	2.0	4.9	7.0 Al	7.3
90-4-2-4	90.2	0.2	0.01	1.4	3.6	4.5 Al	10.7
81-4-15	81.9	0.2	0.01	4.0	13.9		6.5

TABLE 11 — COPPER EFFECT ON ELECTRICAL CONDUCTIVITY OF MANGANESE BRONZE

Composition, per cent				% I.A.C.S.
Cu	Fe	Al	Mn	
57.9	2.4	2.9	2.6	10.8
59.9	2.1	2.8	2.5	10.5
62.0	2.0	2.9	2.5	10.1
64.4	1.9	2.8	2.5	9.5
66.3	1.9	2.9	2.5	9.1

TABLE 12 — ALUMINUM EFFECT ON ELECTRICAL CONDUCTIVITY OF SAND-CAST ALUMINUM BRONZES

Alloy	Composition, per cent				% I.A.C.S.
	Cu	Fe	Al	Ni	
89-1-10	89.1	1.5	9.2		12.9
	89.0	1.3	9.6		14.2
	88.2	1.4	10.1		15.1
	87.8	1.5	10.4		15.4
	87.6	1.4	10.7		15.4
	87.3	1.5	10.9		16.3
86-4-10	87.4	3.4	8.9		12.2
	86.5	3.5	9.7		13.0
	86.4	3.3	10.2		14.3
	85.9	3.4	10.4		14.6
	85.7	3.3	10.7		15.2
	84.4	3.5	11.8		16.8
81-4-11-4	81.2	5.2	8.0	5.3	7.9
	81.0	5.1	8.4	5.2	8.1
	81.1	4.9	8.7	4.8	8.1
	80.5	5.0	9.1	5.0	8.2
	80.4	4.8	9.4	4.8	8.6
	80.0	4.8	9.9	5.0	8.4
	79.5	4.8	10.2	5.2	8.8

TABLE 13 — SILICON EFFECT ON ELECTRICAL CONDUCTIVITY OF SAND-CAST COPPER-SILICON ALLOYS

Alloy	Composition, per cent			% I.A.C.S.
	Cu	Si	Mn	
95-1-4	95.8	2.8	1.0	6.7
	95.8	3.0	1.1	6.5
	94.7	3.8	1.1	5.9
	94.0	4.6	1.1	5.4
	93.6	5.0	1.0	4.8
	Cu	Si	Zn	
92-4-4	91.6	3.2	4.6	7.3
	92.2	3.4	4.2	7.5
	91.3	3.9	4.3	6.7
	90.2	4.3	4.6	6.4
	91.0	4.4	4.5	6.1
	Cu	Si	Zn	
81-4-15	80.9	3.8	15.2	6.8
	81.9	4.0	13.9	6.5
	80.5	4.1	15.2	6.0
	82.1	4.4	13.3	5.7
	82.1	4.6	13.1	5.4
	81.4	4.8	13.7	5.1

copper-silicon-zinc alloys (compare the 92-4-4 curve with the 81-4-15 curve, Fig. 10).

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REFERENCES

1. U. S. Dept. of Commerce, National Bureau of Standards, "Mechanical Properties of Metals and Alloys," Circular C447, p. 159.
2. Brass and Bronze Ingot Institute Manual, *Ingot Brass & Bronze*, Jan. 1956.
3. C. S. Smith and E. W. Palmer, "Thermal and Electrical Conductivities of Copper Alloys," *A.I.M.E. Transactions*, Institute of Metals Div., *vol. 117*, p. 225 (1935).
4. J. G. Kura and R. M. Lang, "Mechanical and Physical Properties of Three Low-Shrinkage, Copper Base Casting Alloys," *A.S.T.M. Proceedings, vol. 58* (1958).
5. A. L. Norbury, "Notes on the Effects of Certain Elements on the Electrical Resistivity of Copper," *Journal of the Institute of Metals, vol. 33*, No. 1, p. 91 (1925).

HEAT TREATMENT OF DUCTILE IRON

By W. D. McMillan

ABSTRACT

The heat treatment of ductile iron castings on a tonnage basis is considered. It is the purpose of this paper to cover the three types of treatment set up to process the castings. First, a full ferritizing anneal. Second, a sub-critical draw for machinability retaining the "as cast" tensile properties and third a two temperature normalizing in the interests of machinability for small castings furnished against the "as cast" specifications.

INTRODUCTION

The heat treatment of ductile iron castings on a tonnage basis is considered, rather than on a specific or individual casting basis. The planned as well as expedient measures that have been put into practice during a period of about eighteen months of production are covered.

The foundry property originally produced gray iron castings. The melting equipment consisted of three No. 9½ acid lined cupolas. The molding facilities include three mechanized units, side floors for short runs and four slinger floors.

In February 1957, a 66 in. externally water cooled refractoryless cupola was installed to melt iron of a composition suitable for ductile iron. This is operated with a basic slag in the interests of a low sulfur content. No other major changes in equipment were made at that time. Currently with increase tonnage on schedule, as well as a considerable potential increase, new facilities are being provided.

Two grades of ductile iron are regularly produced:

As-Cast	80-55-06	Brinell 192-241
Annealed	64-45-15	Brinell 156-212

Because ductile iron will show a hardness of 300 Brinell when air quenched from 1650 F, and will also self-anneal in the mold under the right conditions, to show a hardness as low as 166 Brinell, considerable variation in hardness is to be expected. Casting hardness varies with the rate of cooling and the temperature at which they are shaken out.

The process set up to provide a full ferritic structure utilizes existing annealing equipment in the malleable foundry. The furnace is a batch-type unit with a maximum load of 15 tons, although the heats may run as low as 10 tons of castings. It is a radiant tube fired furnace and a high nitrogen atmosphere is used. The castings are not cleaned after annealing. The heat up period will vary depending on the temperature to which the furnace drops between heats. The cycle is basically as follows:

Heat up to	1600 F	8 hr
Hold at	1600 F	3 hr
Cool to	1350 F	3 hr
Hold at	1350 F	3 hr
		14 hr

A complete cycle can be made in about 17 hr including loading and unloading.

The mechanical properties based on results of the Y Block tests show the following:

Hardness	90% between 170 and 187 Brinell. Maximum hardness 196.
Yield Strength	90% between 49,000 psi and 57,000 psi. Values above 58,000 are associated with higher silicon. Yield of 49,000 psi is obtained with 2.40 silicon.
Tensile Strength	90% between 68,000 psi to 75,000 psi. Higher values associated with higher silicon.
Per Cent Elongation	90% between 18 and 22%. 10% between 16 to 18% elongation.

Casting specified annealed fall into two groups.

1) Those requiring higher ductility because of adjustment in assembly and because of service conditions.
2) Small gears that because of the freezing rate of light sections will show carbides are annealed for machinability. These gears are machined and heat treated to a hardness of about 50 Rockwell C.

DUCTILE IRON COMPOSITION

As mentioned earlier, the composition of the ductile iron is such that the castings will air harden when dumped hot.

The manganese ranges from 40 to 50 points, and reflects the manganese content of the steel scrap which establishes the manganese level.

The copper content ranges from 6 to 12 points. This level also is established by the copper residual in the steel scrap.

The nickel content is maintained at 40 points, and results from the use of nickel bearing treatment alloy. The nickel content contributes about 2500 psi to the annealed iron, and constitutes a safeguard against a yield point below the 45,000 psi specified.

W. D. McMILLAN is Supvr. of Met., International Harvester Co., McCormick Works, Chicago.

With the high air hardenability and a fast shake out, castings of an intermediate section size and intermediate cooling rate show hardness values higher than 241 Bhn, ranging up to 321 Bhn.

Castings in this group are usually made up of ½-in. to 1 in. sections. Location and size of gates and feeders influence the amount of retained heat which affects the ultimate hardness.

The freezing rate of castings in this group is such that no carbides are present, and the hardness results from the rate of cooling or air quenching rather than from the freezing rate of the metal. The temperature from which the casting is air cooled is the controlling factor. These castings are processed by a sub-critical anneal consisting of 3 hr at 1200 F and air cooled. This practice is an expedient rather than a planned process, and has become necessary on certain castings as a result of the hardness developed by hot dumping. This treatment is carried out in a push-type, open-fired furnace which will handle about 700 to 800 lb per hr. Being open fired, the castings are airless-blast cleaned after annealing. This operation is also performed in the same type of furnace as used for the full anneal, the temperature being held at 1200 F for 3 hr.

This is essentially a ferritizing treatment. At 1200 F, ferritizing is not complete and a hardness level maintained of approximately 200 Bhn. At this hardness, the mechanical properties of tensile, yield and elongation will meet the requirements of the as-cast grade.

The 1200 F temperature has been established for the metal produced in this foundry, and takes into account the character of the iron as influenced by the melting operation, residual elements and the manganese, phosphorus and nickel content.

For metal melted under different conditions and of different composition, a higher or possibly a lower sub-critical temperature would be necessary to maintain a hardness at a given level.

BRINELL HARDNESS

The two specifications produced combined show a total spread of hardness of 156 to 241 Bhn. Actually there is an over lap of 20 points of Brinell hardness 192 to 212, which could apply to either the as-cast or the annealed grade. However, from the standpoint of elongation and formability, and sub-critical anneal does not produce the properties or characteristics obtained by an anneal involving a full austenitizing. It is not a satisfactory alternate treatment if the properties obtained by a full anneal are required. It is satisfactory as a treatment where the anneal is performed only to provide a structure of higher machinability.

A third heat treatment process has been set up. This is for castings specified as-cast, but which because of section size, and consequently, freezing rate, will show some carbides. Light section ¼-in. to $\frac{3}{16}$-in. castings fall into this group, and fortunately constitute only about 3 to 5 per cent of the tonnage. Wall thickness is not necessarily the determining factor. "Dead ends," particularly in the drag, are more liable to show carbides than similar sections where the metal can flow through the cavity.

This treatment consists of a two-temperature normalize. It is essentially 2 hr at 1650 F to break up the carbides, cooling to 1450 F and air quenching after 2 hr at this temperature.

Obviously, a sub-critical treatment will be ineffective in removing carbides, and since the specifications require a minimum of 80,000 psi the full annealing treatment cannot be used. This is the reason for the two temperatures normalizing practice.

Castings showing carbides may run as high as 340 Bhn, and as low as 255 Bhn. On heavy section castings, the hardness may be as high as 285 Bhn, and show no carbides. A hardness of 295 or 302 Bhn is regarded as suspicious from the standpoint of carbides. The shape of the casting, size and location in the mold, with respect to the flow of the metal, is better basis for singling out castings that will show carbides than a Brinell hardness test. This is also a lot more practical as many light castings are too small to obtain Brinell readings.

TENSILE PROPERTIES

This heat treatment results in lowering the hardness from the vicinity of 300 Bhn to a range of 217 to 241 Bhn. The tensile properties range from 84,000 to 100,000 psi tensile strength, 63,000 to 68,000 psi yield strength with elongation 10 to 12.5 per cent. These values are generally higher than obtained on as-cast bars at the same hardness levels.

This process is an expedient measure but appears to be of a permanent nature. It relates to a condition that is inherent to the metal composition suitable for ductile iron. However, the 1200 F process is of a less permanent nature, and could be eliminated or at least to great extent by lowering the shake out temperature, which is the dominating factor in causing high as-cast hardness.

The as-cast tensile properties based on Y Blocks, cooled in dry sand molds, reflect the variations in composition, principally silicon and carbon, as the other elements are fairly uniform.

A survey of two representative months production shows that 90 per cent of the results were in the following range:

Tensile Strength	84,000 to 90,000
Yield Strength	56,000 to 67,000
% Elongation in 2 in.	6 to 15 – Ave. 10.9

It is the purpose of this paper to record the three types of heat treatment set up to process ductile iron castings produced at this operation. These represent the measures taken and pertain to this specific operation. They would not necessarily apply without some modification to the product of other foundries.

The economic virtue of ductile iron lies in as-cast ductility. It is referred to as "the cast iron that can be bent," which should be qualified as to how much bend and at what hardness.

There lies ahead further study of composition, character of the charge materials and modification of the facilities in the interest of increasing the tonnage that can be used as-cast.

RADIOGRAPHY, MICROSTRUCTURE AND MECHANICAL PROPERTIES OF CAST MAGNESIUM–THORIUM–ZIRCONIUM ALLOY HK 31A

a correlation

By T. R. Bergstrom and R. G. Bassett

ABSTRACT

Radiographs of cast HK31A magnesium parts exhibit sweeping dark lines, light areas of both circular and angular outline, light areas similar in shape to shrinkage cracks and mottled or banded areas. These indications are excellent evidence of macrosegregation in the castings. An evaluation of the effect of this segregation on the mechanical properties of the cast material was essential to continued use of the cast parts. Several castings were selected on the basis of x-ray indications to be cut into tensile specimens representative of the segregation conditions.

These specimens were tensile tested at room temperature and at 500 F. Autoradiographs, radiographs, micrographs and spectographs were made of the specimens to investigate variations in composition and associated effects. It was determined that the x-ray indications were caused by segregation, and that the segregation had a definite effect on the microstructure. Severe segregation was found to lower mechanical properties. Regions deficient in alloying elements had reduced elevated temperature mechanical properties, but had normal room temperature properties.

Areas of high alloying element content were found to have reduced mechanical properties at both room and elevated temperature. Radiographic standards were established on the basis of the evaluation.

INTRODUCTION

The demand for light alloy castings capable of retaining relatively high strength at elevated temperatures led to the development of magnesium alloys containing the rare earth metals, thorium and zirconium. These alloys have been useful in aircraft and missile design, and their increased use is almost certain. Although there are several commercial casting alloys of this type, the following discussion is based on testing and evaluation of the magnesium alloy designated HK31A (chemical composition: Th 2.5-4 per cent, Zr 0.5-1.0 per cent, Mg remainder).

Thorium, with an atomic weight of 232 and a density of 11.5, is more opaque to x-ray than magnesium, with an atomic weight of 24 and a density of 1.74. Zirconium, with atomic weight of 91 and a density of 6.5, is also more opaque than magnesium (the absorption coefficient of x-rays increases with the atomic number of the absorbing atoms, the number of atoms per unit volume and the x-ray wave length).[1] As a result, portions of a casting that contain higher than normal concentration of thorium or zirconium will absorb greater amounts of radiation and appear as light areas on x-ray films. Conversely, deficient regions will absorb less radiation and appear darker.

These conditions manifest themselves in the radiograph as sweeping dark lines, light circular or angular areas, mottled or banded areas and as light areas similar in shape to shrinkage cracks. They are easily observed and provide excellent indications of gross variations in chemical content. They are also a constant source of perturbation to personnel in the x-ray laboratory.

The determination of the effect of these anomalies on the room and elevated temperature properties of HK31A-T6, and the subsequent adoption of radiographic standards to establish permissible limits, was deemed essential to the use of cast HK31A magnesium as a production material.

INVESTIGATION AND TEST RESULTS

A power-pack door panel casting, as shown in Fig. 1, is used on an interceptor missile. This casting measures approximately 27¼-in. x 20½-in. The configuration has walls as thin as 0.22 in. and one mass roughly 4x1x1½-in. The panel doors are given complete x-ray inspection, and on this basis several were selected for the testing and evaluation reported herein. The selection of specimens from the castings required repeated dissecting and x-ray checks to insure selection of the exact conditions desired.

Unfortunately, many of the best illustrations of the conditions in question occurred where the configuration was such that tensile specimens could not be obtained. The shape and size of the tensile specimen selected can be seen in Figs. 2, 3 and 4. For room temperature testing the specimen thickness was as near the thickness of the casting as possible, but for the elevated temperature testing it was necessary to standardize on two thicknesses (0.220 and 0.160 in.) to permit a uniform heating rate. Altogether, a total of 39 specimens was tested, 25 at room temperature and 14 at 500 F (260 C).

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The elevated temperature testing was performed in a 10,000 lb, radiation-type, quick heat test machine, by bringing the samples to 500 F in less than 30 sec, holding at temperature for 30 sec, and loading to cause failure in 3 to 10 sec, or a rate of 4 to 6 ksi p[er] sec. The room temperature loading rate averaged approximately 6 ksi per sec. The test results, together with the condition observed in the gage area, are presented in Tables 1 and 2. Figures 2 and 3 are negative prints of radiographs of the specimens tested at 500 F.

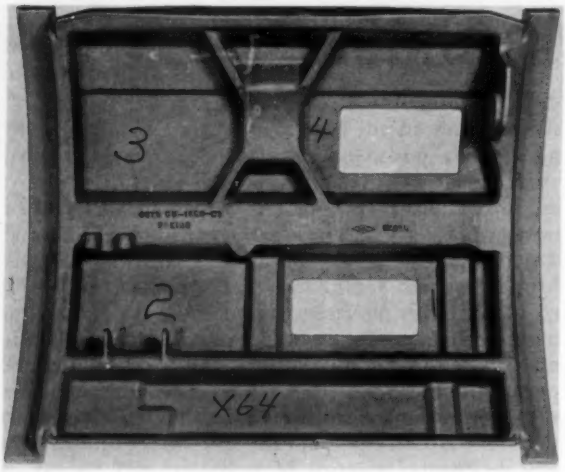


Fig. 1 — Power pack door panel casting.

Figure 4 is a similar print of selected specimens tested at room temperature. Specimens S-3, S-30 (Fig. 2) and F-3 (Fig. 3), respectively, are examples of the spherical type of segregation, the angular type of segregation and the deficiency condition.

It has been possible using radiographic, spectrographic and metallographic examination to determine with fair certainty that a compositional variation is responsible for the effects observed. Thorium, being radioactive, will expose industrial x-ray film in a period of several days. Figure 5 is a positive print of an autoradiograph of specimen S1-4. Figure 6 is a negative print of an x-ray of the same specimen. The fact that segregate particles appear light on the positive print of the autoradiograph indicates a higher thorium content.

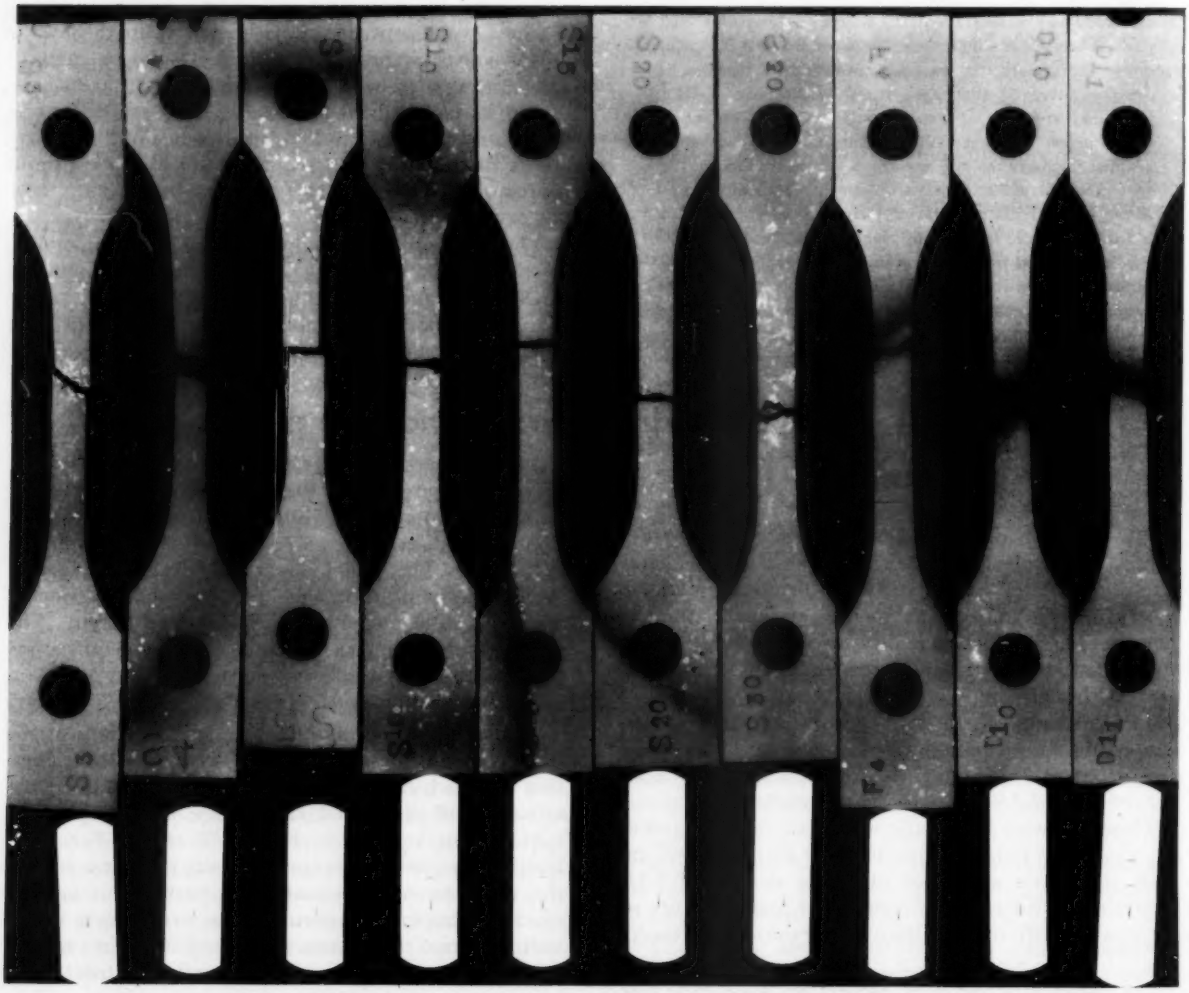


Fig. 2 — Negative print of radiograph showing condition present in elevated temperature samples and location of failure. Reduced slightly in reproduction.

The light appearance of the particles on the negative print of the x-ray indicates that the particles are more opaque to radiation. This would be expected from particles of high thorium content. Similarly, the sweeping dark band evident in Figs. 5 and 6 indicates a low thorium content.

Five of the tensile specimens were examined by emission spectroscopy to compare the chemical composition of the fracture surface with some other area of the specimen. The results of composition analysis by emission spectrographic method are given in Table 3.

The thorium results may vary 0.2 to 0.3 per cent from true values, especially on the fracture surfaces. The zirconium analysis should be accurate to ± 0.05 per cent where near the mean value of 0.6-0.7 per cent. The two high zirconium readings are about twice the value of our highest quantitative standard.

Figure 7 is a photomicrograph of a section of one of the smaller spherical segregate particles evident in the gage area of specimen S-17, Fig. 4. All observed

TABLE 1 — ROOM TEMPERATURE TEST RESULTS

Spec. No.	Condition Present in Gage Area	Ultimate Tensile Str., psi	0.2 Per cent Offset Yield Str., psi	Elong. Per cent in 1 in.	Remarks
CT-1	Clean	30,700	16,000	8	Control specimen
CT-2		30,300	14,600	10	Control specimen
CT-3		33,700	16,300	10	Control specimen
CL-1		29,700	20,100	8	Control specimen
CL-2		33,500	16,600	8	Control specimen
CL-3		31,400	15,500	9	Control specimen
S1-1		30,700	14,600	7	Control specimen
S1-2		29,300	14,700	7	Control specimen
S-32		30,400	16,000	8	Control specimen
	Averages	31,077	16,044	8.3	
S2-3	Heavy segregation	26,900	15,500	8	Failed in segregated zone.
F-1		26,200	14,600	7	Failed in segregated zone.
S-14		26,600	16,500	3	Failed in segregated zone.
S-17		26,300	16,200	3	Failed in segregated zone.
	Averages	26,500	15,700	5.25	
S1-3	Moderate segregation	29,000	14,600	6	Failed in clean zone.
S1-4	Moderate segregation	28,100	15,600	7	Some segregation at break.
S-11	Moderate segregation and deficiency	28,300	14,100	5	Failed in segregated zone.
S2-1	Light segregation	30,600	15,400	8	One segregate particle in break.
S2-2	Light segregation	30,900	14,800	8	Failed in clean zone.
	Averages	29,380	14,900	6.8	
F-2A	Th deficiency	31,000	14,900	10	Failed 1/8-in. from sharp indication.
F-3	Th deficiency and light segregation	29,000	17,300	8	Failed in segregation zone 1.
F-5A	Th deficiency	29,300	14,400	9	Failed 1/4-in. from sharp indication.
F-2	Th deficiency	35,000	14,700	6	Failed 1/2-in. from sharp indication.
F-5	Th deficiency	30,900	14,100	5	Failed 1/4-in. from sharp indication.
F-6	Th deficiency	34,700	16,100	7	Failed in deficient zone.
	Averages	31,650	15,250	7.5	
F-4	Th deficiency and dross	26,600	14,600	6	Failed in dross.

TABLE 2 — 500 F QUICK HEAT TEST RESULTS

Spec. No.	Condition Present in Gage Area	Ultimate Tensile Str., psi	0.2 Per cent Offset Yield Str., psi	Elong. Per cent in 1 in.	Remarks
S-4	Clean	23,700	13,100	18	Control specimen
S-15	Clean	26,000	*	10	Control specimen
**	Clean	22,100	12,114	18.4	
S-3	Heavy segregation	20,800	12,800	10	
S-30	Heavy segregation	16,900	13,400	6	
S-10	Medium segregation	22,300	*	6	See Figs. 2 and 3 for positions of failure.
S-20	Light segregation	23,400	13,300	8	
S-2	Light segregation	21,000	11,800	9	
S-31	Light segregation	19,500	12,700	7	
	Averages	20,650	12,800	7.7	
F-3	Th deficiency	22,600	13,200	20	
F-4	Th deficiency	22,100	12,100	10	See Figs. 2 and 3 for positions of failure.
D-11	Th deficiency	18,800	13,900	16	
D-10	Th deficiency	17,500	9,600	10	
	Averages	20,250	12,200	14	
S-9	Mottled type segregation	22,000	14,400	9	See Fig. 3 for position of failure.
L-1	Light stress segregation	21,800	13,000	9	Specimen failed in bolt hole.

*Curve not suitable to obtain yield strength.
**Average of eight other specimens of HK31A tested similarly.

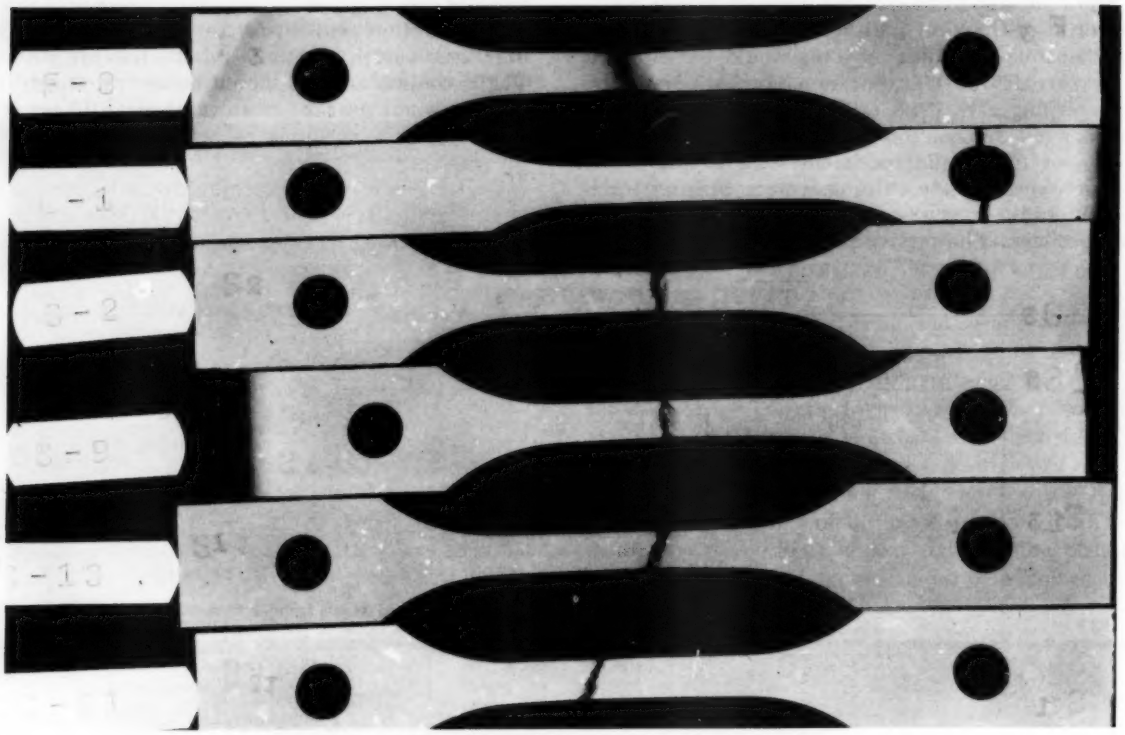


Fig. 3 — Negative print of radiograph showing condition present in elevated temperature samples and location of failure.

TABLE 3 — FRACTURE SURFACE CHEMICAL COMPOSITION

Specimen	Thorium Fracture, %	Thorium Specimen Surface, %	Zirconium Fracture, %	Zirconium Specimen Surface, %
F- 3 (Fig. 3)	1.2	3.5	0.43	0.66
D-11 (Fig. 2)	2.0	2.2	0.60	0.67
S-11 (Fig. 4)	1.7	2.4	0.71	0.70
S-30 (Fig. 2)	3.9	2.5	1.25	0.68
S-17 (Fig. 4)	3.1	2.4	1.00	0.65

particles of this type had a clearly defined periphery, and most had a void in the center. Figure 8 at 1000 × was taken of the inside of the particle, and shows a lamellar eutectic-type structure similar to that reported by Yamamoto, Klimek and Rostoker.[2] The ternary system magnesium-thorium-zirconium has been examined with nominal zirconium contents of 0.5, 0.75 and 1.0 per cent, and thorium contents ranging from 0.5 to 6.5 per cent. Reference 2 is the report on this work.

A tentative vertical section diagram at constant 1 per cent zirconium has been constructed. This diagram (Fig. 9) indicates that in solidifying, an alloy with 3 per cent thorium would produce only magnesium solid solution. Subsequent cooling to below about 1000 F (538 C) would start the precipitation of the intermetallic compound Mg_5Th. Diffusion in this alloy is reported to be quite slow due to the size of the thorium atom. It is then probable that as the casting cools below the liquidus curve a considerable deviation from equilibrium exists.

This deviation or undercooling effect will cause the precipitating magnesium solid solution to contain less thorium than indicated. Therefore, the last portion of the liquid will be of such a composition that a third phase, zirconium solid solution, will form. The zirconium solid solution formed is then consumed in a peritectic reaction with the melt, producing magnesium solid solution and the intermetallic compound Mg_5Th.

The liquid remaining below the temperature of the peritectic reaction will then solidify over a temperature range to produce a eutectic-type structure consisting of magnesium solid solution and Mg_5Th. The spherical shape of the particles, observed in Figs. 2, 3 and 4, their void center, and their eutectic-type structure, suggests that they are being formed while the casting is partially molten, possibly where the metal contains gas pockets.

Figures 10 and 11 were taken respectively on the dark and light sides of the band running roughly parallel to the length of sample S-15, Fig. 2. A sharp difference was noticed in the rate of etching of the grain boundaries in the respective areas. Figure 11, taken on the side of high thorium content, has clearly defined fine grains indicating normal precipitation at the grain boundaries. The fine grained structure can be attributed to a normal concentration of zirconium, which is added as a grain refiner.

Figure 10, taken only 0.062 in. (center to center) from Fig. 11, is on the side of the band with lower thorium content. The grain boundaries are not as readily etched, and the grains are considerably coarser, which would tend to confirm the suspected lack of thorium and zirconium.

The x-ray of specimen S-9 (Fig. 3) reveals what

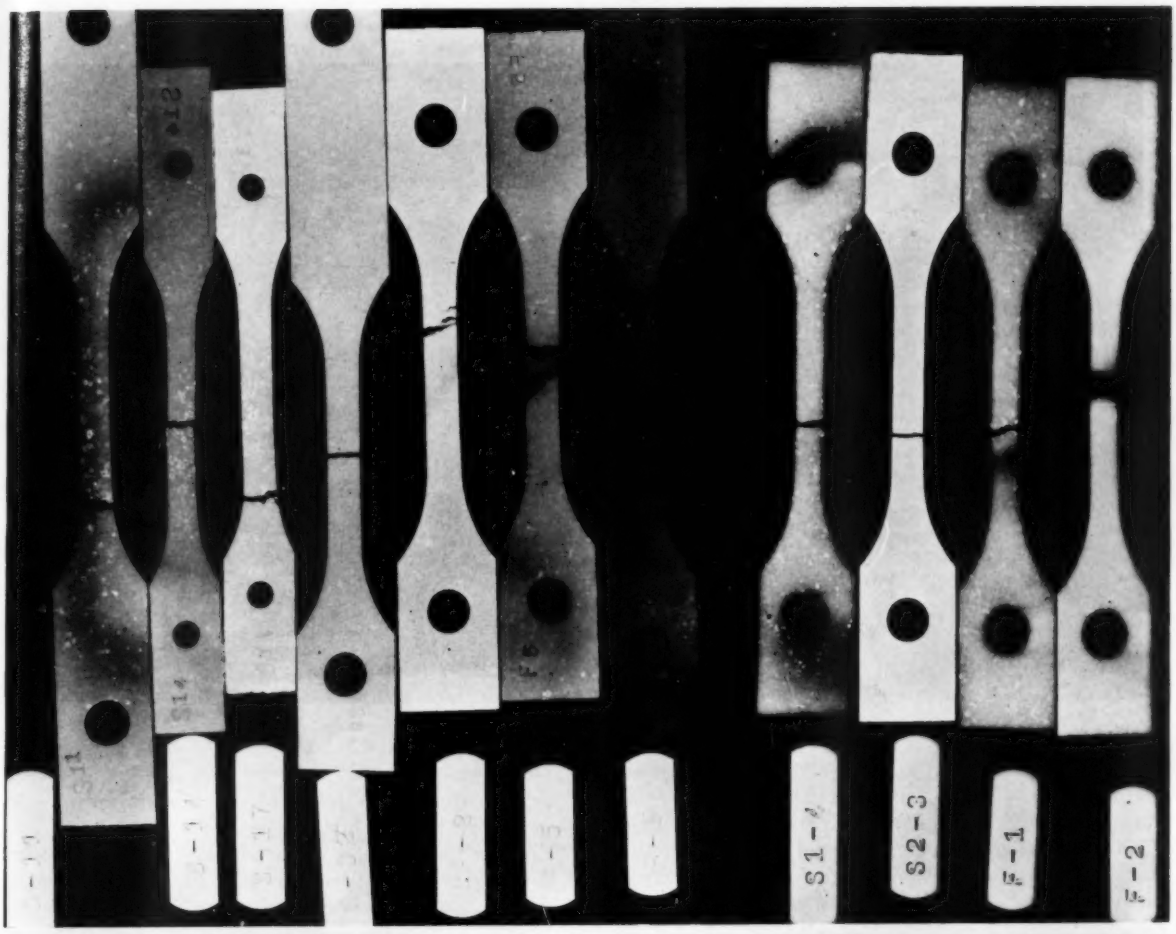


Fig. 4 — (Above) Negative print of radiograph showing condition present in room temperature samples and location of failure. Reduced slightly in reproduction.

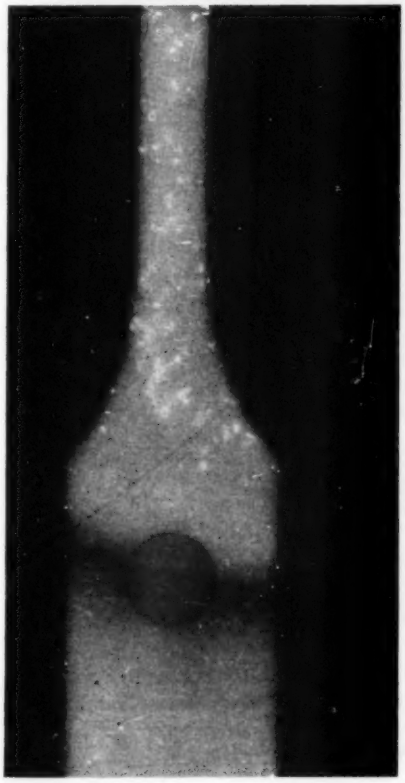


Fig. 5 — (Left) Positive print of an autoradiograph of specimen S1-4. 1.7 ×.

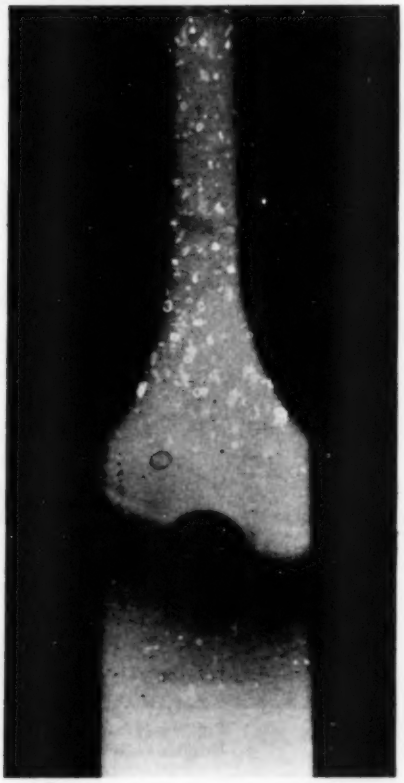


Fig. 6 — (Right) Negative print of radiograph of specimen S1-4. 1.7 ×.

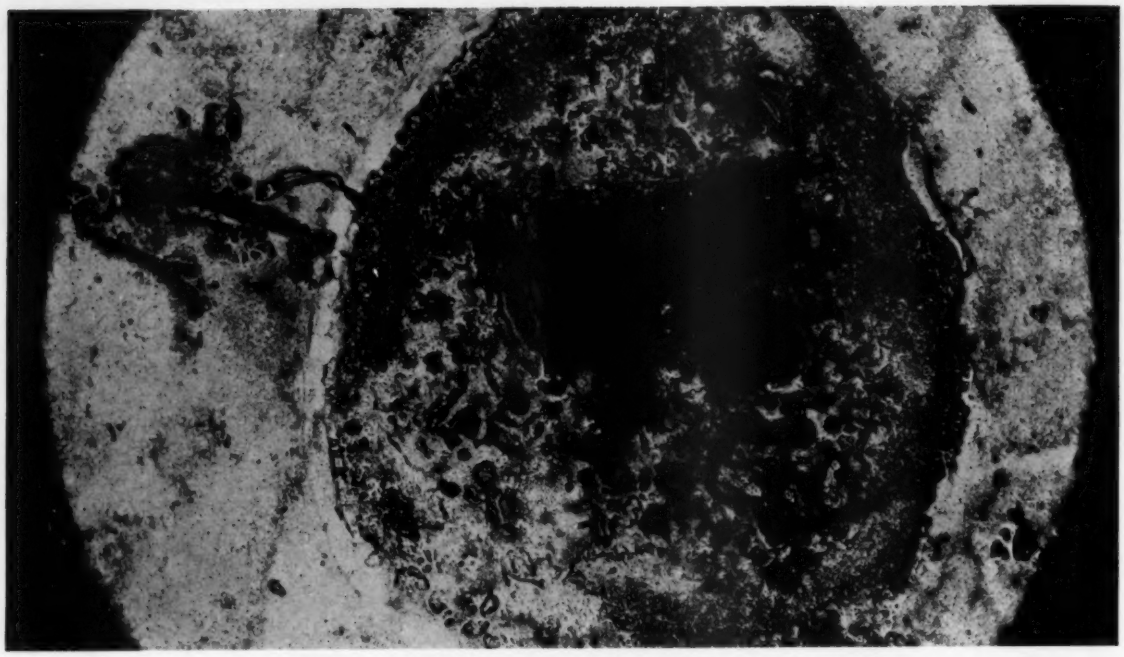


Fig. 7 — Spherical particle observed in specimen S-17. Unetched. 50 ×.

might be termed a mottled or banded appearance. Figure 12 is a photomicrograph of this area, showing grains which contain clusters of precipitated particles. This greater than normal amount of precipitate results in the mottled appearance of the x-ray. Similiar radiographic indications have been observed in EZ 33 castings (chemical composition: rare earth-2.5-4.0, Zn-2.0-3.1, Zr-0.5-1.0, remainder Mg). These indications were associated with microshrinkage being fed by near eutectic material. However, precipitation within the grains was not noted.[3]

Figure 13 is a photomicrograph of one of the angular indications of specimen S-30, Fig. 2. This type defect meanders through the metal in a pattern that might be expected of dross inclusions. The defects are lined with material of eutectic-type structure, although

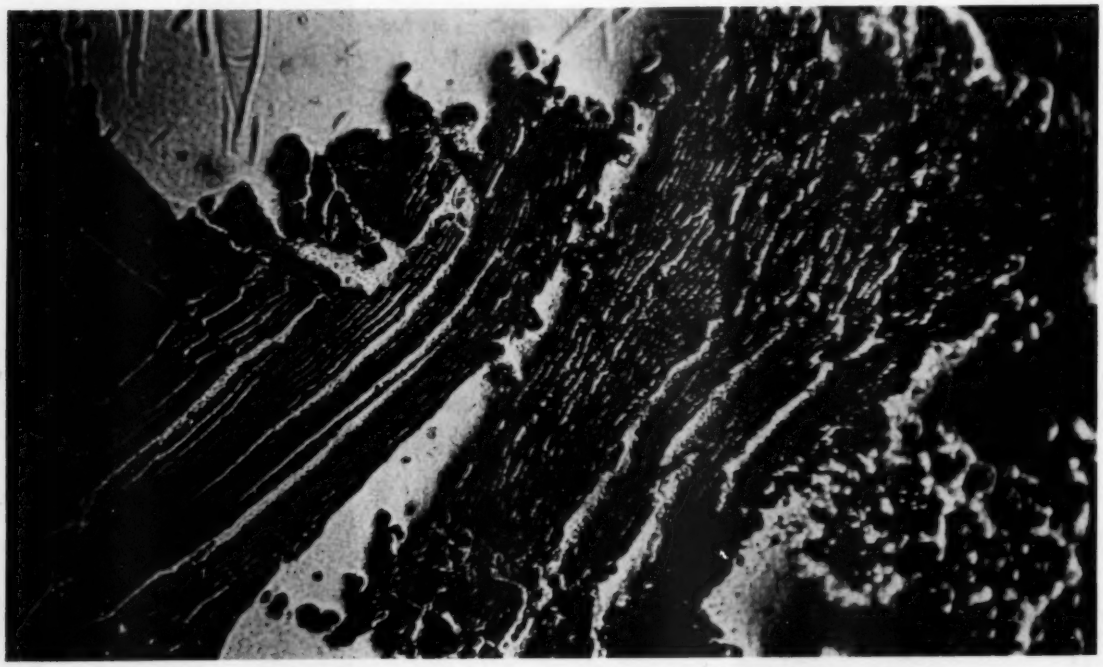


Fig. 8 — Lammellar eutectic-type structure observed in spherical particle shown in Fig. 7. Ethylene glycol 75 ml, distilled water 24 ml, concentrated HNO_3 1 ml, etch. 1000 ×.

the clearly defined lamellar eutectic-type structure of Fig. 8 could not be seen. Figure 14 is a photomicrograph at 1500 diameters of one region of this defect. These defects are similar to the spherical type in that they have a void region in the center.

DISCUSSION AND CONCLUSIONS

An examination of Table 1 indicates that the deficiency condition does not affect the room temperature mechanical properties of the material. It will be noted that the room temperature specimens of Fig. 4 did not fail in regions of deficiency. Conversely, testing at 500 F shows that the deficiency condition definitely promotes failure. This can also be observed from the location of the applicable fractures in Figs. 2 and 3. The values of Table 2 show that this condition reduces average ultimate tensile strength by about 8 per cent, and the per cent elongation by about 24 per cent.

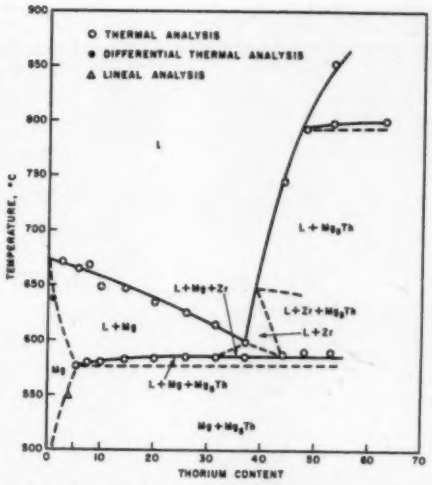


Fig. 9 — Tentative vertical section of the magnesium-thorium-zirconium ternary at 1 per cent Zr. From WADC TR-411, ref. 2.

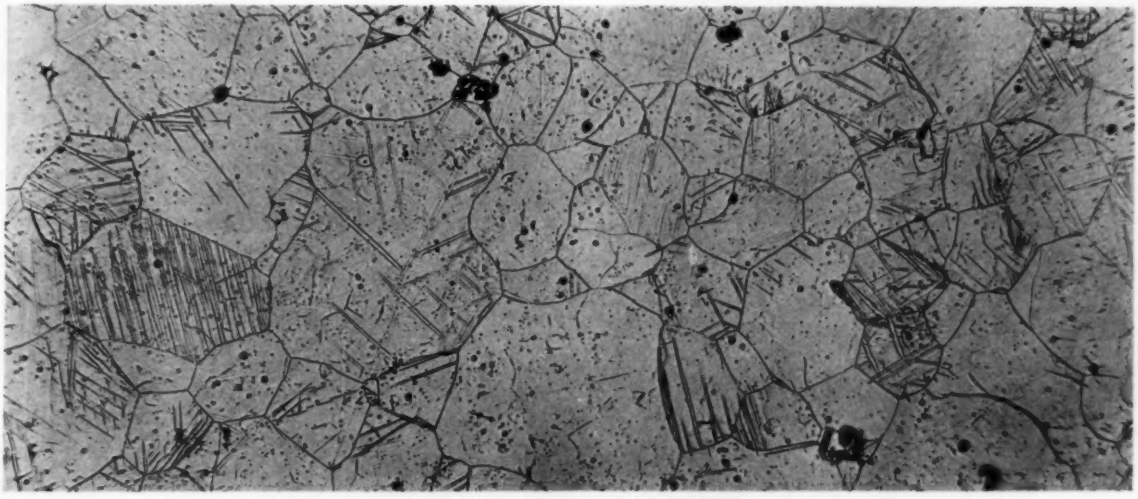


Fig. 10 — Grain size observed on the dark side of the band appearing in specimen S-15, Fig. 2. Ethylene glycol 75 ml, distilled water 24 ml, concentrated HNO_3 1 ml, etch. 100 ×.

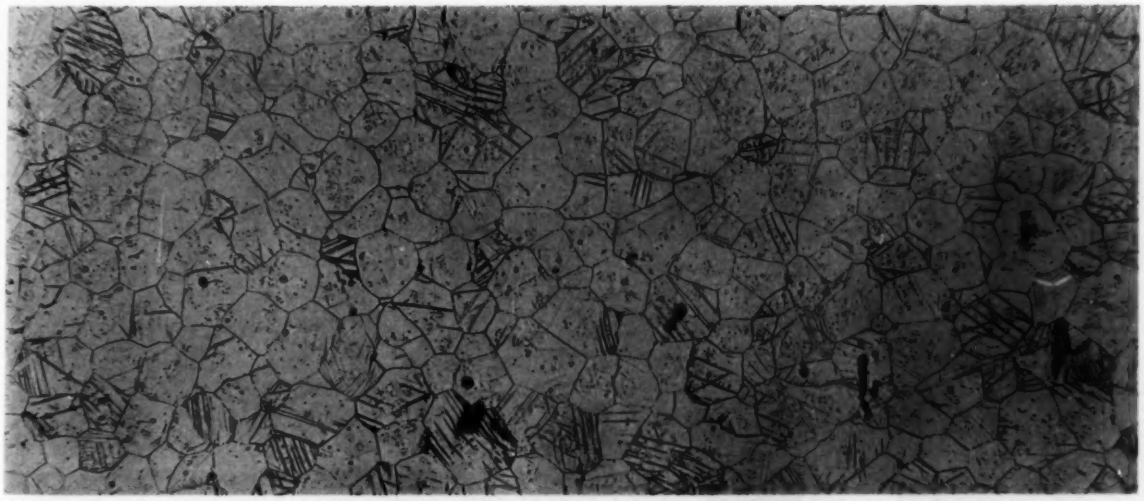


Fig. 11 — Grain size observed 0.062 in. from Fig. 10 (horizontal center line distance), or on the light side of the band of specimen S-15, Fig. 2. Ethylene glycol 75 ml, distilled water 24 ml, concentrated HNO_3 1 ml, etch. 100 ×.

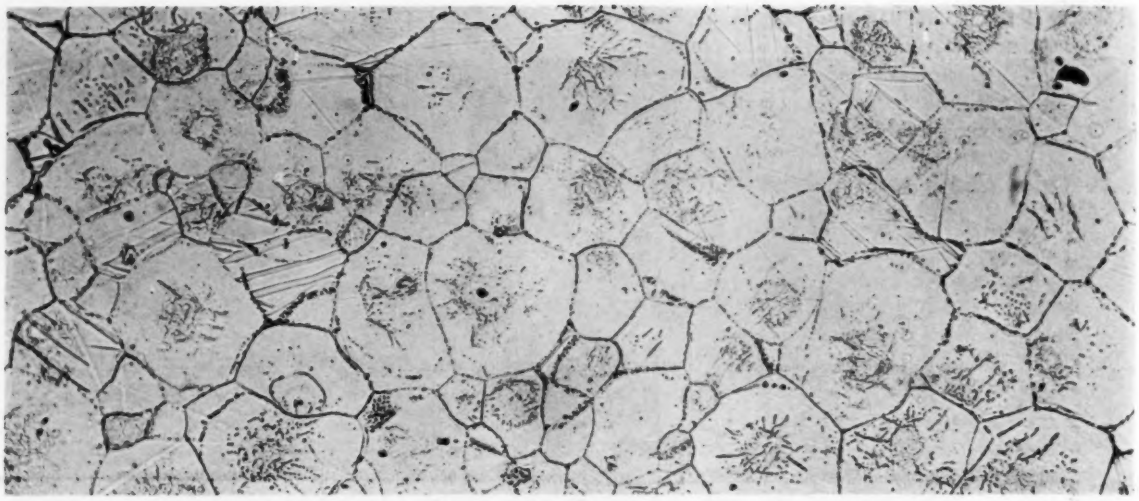


Fig. 12 — Photomicrograph of mottled area of specimen S-9, Fig. 3. Ethylene glycol 75 ml, distilled water 24 ml, concentrated HNO_3 1 ml, etch. 250 ×.

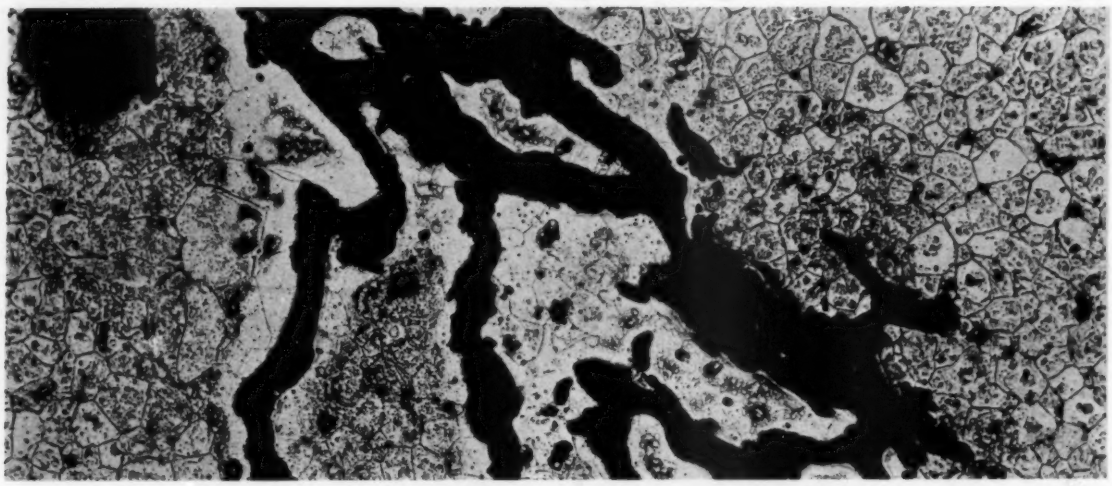


Fig. 13 — Gross imperfection that caused one of the angular indications appearing in specimen S-30, Fig. 2. Ethylene glycol 75 ml, distilled water 24 ml, concentrated HNO_3 1 ml, etch. 75 ×.

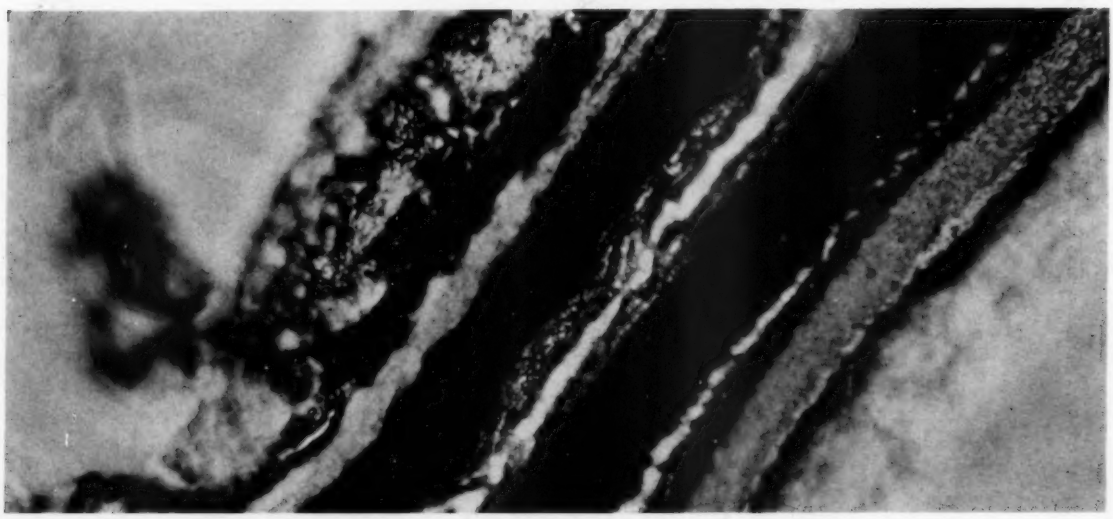


Fig. 14 — Structure observed in region of angular type indication appearing in specimen S-30, Fig. 2. Unetched. 1500 ×.

This reduction would be expected considering that thorium is added to impart high temperature strength. The coarse grain size observed in association with deficient regions is believed to be significant in contributing to the loss of ductility and ultimate strength.

The condition of spherical and angular segregation has the effect of reducing ultimate tensile strength and per cent elongation at both room and elevated temperature. Room temperature testing of specimens with heavy segregation showed an average loss in ultimate tensile strength of 15 per cent, and a loss in per cent elongation of 37 per cent. The average 500 F ultimate tensile strength and per cent elongation values of all specimens with spherical and angular type segregation were found to be 7 and 58 per cent, respectively, lower than observed in the normal samples.

Although only two specimens containing the angular appearing segregation were tested, their performance and the nature of the defect indicates that they are more detrimental than the spherical type. Neither segregation nor deficiency had any significant effect on the 0.2 per cent offset yield strength, or the proportional limit of the material at room or elevated temperature.

All of the specimens tested met the mechanical property requirements of Federal Specification QQ-M-56a for specimens cut from the casting. The authors' company has, however, found it necessary to design castings on the basis of higher mechanical properties. Certain critically stressed regions of the casting, shown in Fig. 1, are required to have the following room temperature mechanical properties: 27,000 psi minimum ultimate tensile strength, 13,000 psi minimum yield strength at 0.2 per cent offset, and 4 per cent minimum elongation.

These regions are expected to have properties of 19,000 psi minimum ultimate tensile strength and 10,700 psi minimum yield strength when tested at 500 F. In order to be relatively certain that these castings will perform to expectation, it was deemed necessary to add an additional inspection requirement to the already thorough inspection procedure.

This requirement is that regions of castings that are delineated as being critically stressed may not exhibit segregation or deficiency in excess of that illustrated in the three x-ray standard film blocks, shown in Fig. 15. Although these standards do not cover all eventualities, they are a step in the right direction, and they will be used by the authors' company until replaced or supplemented.

The trend in casting design for missile applications is becoming exceedingly clear. The extensive use of castings for missile structure application will depend on the casting industry progressing to where it can offer the missile designer the following advantages:

1) Guaranteed strength and ductility in the casting. Separately cast test bar values will not be acceptable.
2) The maximum in consistency and reliability throughout the parts of any lot, and between all lots of any part. Castings to be competitive cannot be penalized by any casting factor. Accurate analysis of heats is an absolute necessity.

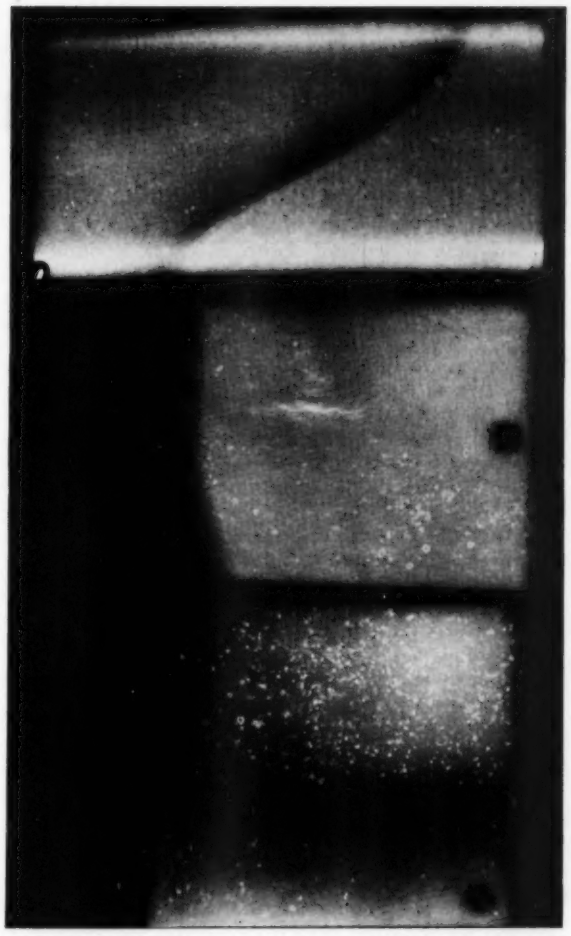


Fig. 15 — Negative print of x-ray standard radiograph for use in critically stressed regions of cast thorium magnesium.

3) Strength and ductility equal to that obtained in wrought products. Castings will have to be designed on the basis of the ultimate in wrought properties.
4) Increased dimensional accuracy permitting the maximum use of as-cast surfaces.
5) Specifically formulated casting alloys to compete with the wrought superalloys.

The airframe industry is usually limited to investigations of the type represented by this discussion. The foundry, because of its nature, must perform the basic research and development needed to eliminate the various causes of defects, and provide the missile industry with sound, strong, reliable castings.

REFERENCES

1. Kent R. Van Horn, "Radiography of Metals," A.S.M. *Metals Handbook*, p. 141, 1948 edition.
2. A. S. Yamamoto, E. J. Klimek anl W. Rostoker, "The Constitutional and Aging Characteristics of Magnesium – Thorium and Magnesium-Thorium Ternary Alloys," W.A.D.C. Technical Report 56-411, ASTIA Document No. 118107.
3. H. M. Skelly and D. C. Sunnucks, "Segregation In Magnesium-Rare Earths-Zinc Zirconium Alloys," AFS TRANSACTIONS, *vol. 62*, p. 481, 1954.

GRAY IRON PERMANENT MOLDING

By H. U. McClelland

ABSTRACT

Although the tonnage of gray iron castings produced in metal molds is relatively small when compared to the tonnage produced by all casting methods, it is felt by the author's company and others producing quality castings by this method that their contribution to the users of castings is important.

The purpose of this paper is to present a short history of the origin and development of gray iron permanent molding, a description of the process and the methods used in the design and fabrication of the molds. The metallurgy and dimensional tolerances of the castings produced by the process will be discussed as well as the scope of the process and its advantages and disadvantages. It is hoped that the paper will give the producers of castings by other methods a better understanding of the procedures used, and that the users of gray iron castings will be given a picture that may help in their purchasing decisions for product applications that can be served by this process.

INTRODUCTION

Although the use of permanent molds in the casting of metals is centuries old, the first successful method for molding gray iron castings in production quantities is that which will be discussed here. The process is approximately 35 years old, having been conceived and developed by a manufacturer of automotive carburetors who was having difficulty procuring gray iron castings, free of porosity, and who, though having no foundry experience or facility, decided to enter this entirely new field. As in most ventures of this nature, there were many disappointments and failures, and several times during the early development the project was almost abandoned.

The key to ultimate success was the formulation of an adequate mold coating which permitted molten metal to be poured into metal molds of similar analysis without fusing. The facility was operated as a captive foundry for several years, casting carburetor components exclusively and manufacturing permanent mold machines for sale. The decision was then made to enter the commercial casting field and discontinue the sale of machines domestically. In 1932, the author's company acquired ownership of plant and patents, and has continued to expand and develop the process. Several installations are in operation under license in the United Kingdom and continental European countries.

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In the United States, at the present time, there are three other foundries producing gray iron castings by this method and in the size range of those produced by the author's company; two are captive, one commercial.

The method to be described is unique and different from other means of producing gray iron castings only in respect to the molding material used, the manner in which the molds are conveyed, the products used in the manufacture of the molds and the preparation of the molds for pouring. After the mold is poured and the casting removed, subsequent operations required to prepare the part for shipment and ultimate use are similar to those inherent to all foundries.

Mold Transport Technique

Present day production foundries, with few exceptions, use various handling devices and techniques to transport molds through the making and pouring cycle. In the permanent mold process, the conveyor used is a turntable type of machine which carries the mold through the following cycle:

1) Mold preparation where a thin coating of carbon is deposited on face.
2) Core setting (if cores are required).
3) Pouring.
4) Ejection or shake-out of the casting.

This machine, approximately 14 ft in diameter, consists of a hub from which project 12 hollow arms to which are attached the outer or stationary heads used to support one-half of the mold. The inner or movable heads are actuated by air cylinders whose movement is controlled by cam-operated valves. Both heads have air passages through which approximately 500 cfm is drawn across the back of each half mold for the purpose of cooling. This air passes from the heads, through the hollow radial arms to an exhaust stack. The speed of rotation is controlled by an infinitely variable speed reducer which guarantees a constant cycle speed, determined by the solidification rate of the castings being produced.

All operations are performed while the machine is in motion. It is extremely important that the continuity of operation be maintained at all times with a minimum of interruptions for any reason. Quality and uniformity of product depend greatly on this continuity. The advantages gained, volume-wise, are quite obvious. Shut downs of short duration for mold

cleaning or alignment and minor mold repairs are necessary and permissible, but are rigidly controlled.

The metal mold consists of two shallow open boxes, the closed sides containing the casting cavities and the gating system, the narrow sides forming air chambers in which are integrally cast cooling pins and/or fins to assist in the dissipation of heat from the mold. The metal used to cast these mold halves has essentially the same analysis as that of the castings produced by them, which means that worn-out molds can be remelted as regular returns. Other metals and ferrous alloys have been tried, but this is the analysis that the author's company has found to give the longest life and best operating conditions. Castings for the mold halves are cast by conventional means, either in oil sand or green sand.

The CO_2 process has also been used in making certain types of mold castings, and has been found to have definite advantages. When the contour of the casting to be produced is complicated, and it is difficult and costly to form the mold cavities by machining, it has been found economically sound to purchase mold castings from a producer of precision castings. A minimum of machining is then necessary, and any number of exact duplicates can be made. These molds will be as dimensionally accurate as the pattern from which they are made.

Mold and Casting Drawings

Before the mold is designed and fabricated, a casting drawing of the part to be produced is submitted to the customer for approval of any minor changes which are considered desirable from a foundry standpoint, and which will not be detrimental to processing or ultimate function of the casting. It is often possible to recommend design changes which reduce weight, increase strength or eliminate processing operations, thereby reducing costs and assisting the customer in the engineering of his casting. Complete mold drawings are made after receipt of customer approval of the casting drawing. Since all molds are standard in size, with a maximum face area of 16x20 in., placement of cavities to secure the maximum potential is most important.

Ordinarily center sprues are used, but where size or weight does not allow multiple cavities offset sprues are employed. Various gating systems are used, the determination of the type employed is largely dependent on past practice, type and size of casting and placement of cavities. In multiple-cavity molds the risers of the lower cavities serve as gates of those above. The standard gate is 0.065 in. deep with the width variable, but in certain applications gates as small as 0.025 in. in depth have been successful. The gate size, the size and shape of sprue and runners are important in the elimination of slag and impurities in the casting cavities and the control of pouring time.

As in all casting processes, pouring time is one of the most important elements in the casting process, and in permanent molding this function of the gating system is most significant. Inasmuch as a metal mold has no permeability, and in this process all molds are poured in a vertical position, time must be allowed to exhaust the air contained in the cavities through the risers which act as vents. Where gases or air is entrapped means must be provided for removal. This is accomplished by the insertion, where necessary, of small round plugs from which segments 0.015 in. deep have been removed.

This size opening allows air or gas to escape, but restricts the entrance of metal. Since the unique properties of the castings produced by this process are dependent on rapid solidification, anything that can be done to expedite quick freezing is desirable. To achieve optimum results, special consideration is given to the metal thickness and contour of the model backing, the placement of cooling pins and fins to assist in the rapid dissipation of heat from the mold during its cycle. The thickness of the mold backing varies between 0.50 in. and 1.25 in. depending on the shape, size and mass of the casting to be produced.

Core Prints

Where the use of cores is necessary, core prints must be designed not only for accurate location but for suspension in the vertical mold prior to closing. Both oil sand and shell cores are used, the choice depending on economics, the requirements of the casting in dimensional accuracy, complexity of the cavity to be formed in the casting and availability of equipment. In many instances, although the initial cost of shell cores may be greater than that of cores made from oil sand, it has been found that net cost reductions either in the casting itself, or in the ultimate part made from the casting, can be realized.

In other cases, complexity or fragility is such that it is impossible to produce the core by any other method. Considerable savings are also possible in the maintenance and repair of core boxes. Shell cores, being hollow, also expedite the removal of core gases and assist in a more rapid solidification of the metal in the cored areas, especially desirable in this process.

At the author's company, iron of but one type is produced with an analysis of T.C.–3.45-3.65; Si–2.45-2.65; Mn–0.70-0.90; S–0.10 max.; P–0.25 max. This is the analysis that produces castings with the desired physical properties of strength, hardness and machineability with the lowest scrap and highest yield. Since all castings produced by this process are annealed to eliminate all chilled area induced by casting in a metal mold, the resulting microstructure consists of ferrite and finely dispersed eutectiform graphite, with faint traces of pearlite sometimes present. This metal is sensitive to section size, although not to the degree found in other casting methods, and so may contain some fine flake graphite in heavy sections.

The Bhn range is 170-207 and tensile strength 30,000 psi and up, depending on section size. The material is amenable to hardening, either by the induction heating method or the more conventional oil or water quench and draw process. Hardness of R_c 50-55 can be readily obtained with tensile strengths in excess of 50,000 psi.

Tolerances and machining finishes in any casting are always of great interest and importance, and in

these days of automated machining and intense programs of cost reduction, these two characteristics assume even greater significance. Although each casting must be considered individually, the usual tolerance maintained on exterior dimension is ±0.025 in. This degree of tolerance is necessary on dimensions across the parting line due to some mold spread which occurs during the solidification process. However, on small castings of comparatively thin section, tolerances of ±0.015 can be guaranteed.

Locating points and chucking surfaces can be maintained consistently uniform from casting to casting as rigid mold walls produce constant dimension repetitively. Insofar as finish allowances are concerned, customers require different allowances ranging from zero finish to a maximum of 0.090 in., depending on size, warping tendencies of the casting, method of machining and, of course, ultimate use of the casting. It is safe to say that the average finish allowance is 0.060 in. to 0.070 in., with grinding finishes as low as 0.025 in. possible on small flat parts such as refrigeration valve plates.

Maximum Face Dimensions

Inasmuch as the standard production mold has maximum face dimensions of 20x16 in., it is obvious that the size of casting that can be produced is limited to that which can be contained in this face area. Casting weights range from a few ounces to 25 lb, the latter being exceeded when contour or section thickness of the casting permit practical use of the process. Section sizes range from a minimum of 0.190 in. to a maximum of 2 in. Straight parting lines are desirable from a mold cast standpoint, but are not absolutely necessary.

Small radii tend to decrease mold life while decreasing casting strength by abrupt section change. Castings should have a minimum of three degrees draft and, wherever possible, small pockets or deep cavities should be eliminated to increase mold life and maintain uniformity of casting contours.

All casting processes have disadvantages as well as advantages, but the users of the method being discussed feel that the latter definitely outweigh the former. Obviously the greatest disadvantage is the size limitation imposed by mold dimensions. Equipment to increase the scope of the process is being developed at the present time. Since one type of metal is normally used, it follows that certain castings cannot be produced due to analysis and physical property specifications.

Casting Users

Other castings whose complexity of contour or production volume do not permit practical mold life or the initial mold cost are also eliminated. However, there is a large field whose requirements can be more than adequately met by the process. Permanent molding has been proved especially advantageous to those users of relatively small castings, who demand a dense tight metal without porosity which is easily machined. These users include the automotive industry as well as refrigeration, air conditioning, washing machines, miscellaneous appliances, vee-belt sheaves and hydraulic applications.

These users require uniformity of dimension, tolerances and quality as well as a competitive cost. Their acceptance, and continued use of castings made by the process, speaks eloquently of the scope and advantages of the process. Today, the author's company is producing 1200 different permanent mold castings in gray iron for more than 100 customers.

Let it be said in summation, that the producers of castings by the permanent mold process are proud of their past achievements, and look forward to the future with confidence that they will continue to supply quality castings at a competitive price to casting users everywhere.

GREEN TENSILE AND SHEAR STRENGTHS OF MOLDING SANDS

By R. W. Heine, E. H. King and J. S. Schumacher

ABSTRACT

When a pattern is drawn or a mold is handled, shear and tensile forces are exerted on the molding sand. Compressive forces may be applied to the sand by mold weights, closing over onto cores, the weight of the casting itself and, of course, by squeezing in molding. However, if the mold cracks during pattern drawing, molding or handling, it is commonly due to failure of the sand in tension or shear.

Knowledge of the tensile and shear strengths of molding sands, and principles controlling these properties, is therefore necessary to the best utilization of green sand. The principles relating to tensile and shear properties, their relation to other green sand properties and their connection with molding problems are discussed.

MOISTURE PER CENT EFFECT

The moisture content of a particular green sand mixture largely controls its green shear and tensile strengths. To illustrate, consider the data in Table 1. The sand mixture reported in Table 1 is composed of 8.0 per cent southern bentonite and 92 per cent sand (85 AFS). Sand batches weighing 4500 grams were mixed 10 min in a vertical wheel muller to produce the data listed in Table 1, and plotted in Fig. 1. Tensile tests were performed with a green tensile testing machine modified to extend its range to higher strength levels. Figure 1 shows that green tensile and shear strengths decrease as moisture content increases.

Decreasing tensile and shear strength occurs beyond that moisture content which is required for coating the sand grains and clay.[1] If the moisture content is much below that required for coating the sand and clay, the tensile and shear properties decrease and are erratic, and the sand is too dry and brittle for molding. Moisture contents beyond the highest given in Table 1, 4.4 per cent, were considered too wet for molding this mixture and were therefore not studied.

According to principles of mechanics, green tensile and green shear strengths should be related to each other. Thus, the data in Table 1 and Fig. 1 may be replotted, as in Fig. 2, to show this relationship. Figure 2 shows that throughout the range of usable moisture contents, the two properties are related, both decreasing with increasing moisture content. Thus, it might be inferred that testing for green shear strength of a molding sand will adequately reflect the tensile strength (and moisture content) of the sand. However, later it will be shown that other

TABLE 1 — EFFECT OF MOISTURE ON GREEN SHEAR AND TENSILE STRENGTH*

Mixture	Moisture, %	Tensile Spec. Wt., gm.	Green Tensile Str., psi	Green Shear Str., psi
1. 8% S. bent.	2.7	154	3.8	5.1
92% sand,	3.2	149.5	3.5	6.3
85 AFS	3.6	148	3.1	5.5
	3.8	148	2.8	5.2
	4.3	150.5	2.5	4.5
2. 8% S. bent.	2.6	156	3.1	5.2
92% sand,	3.0	150.5	3.8	6.5
85 AFS	3.4	149	3.1	5.6
	4.0	148.5	2.5	4.9
	4.4	151	2.4	4.5

*Standard AFS specimen.

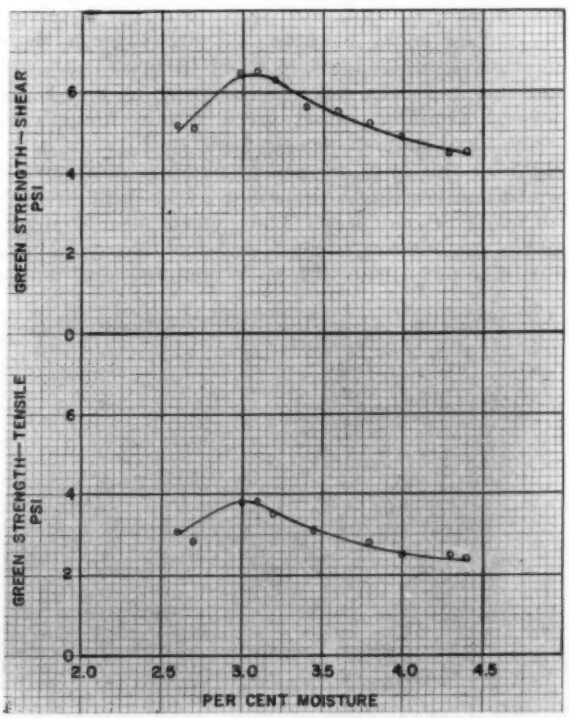


Fig. 1 — Moisture content effect on green tensile and shear strength of a sand bonded with 8.0 per cent southern bentonite. Data of Table 1.

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factors can prevent this simple relation from being true.

Other sand mixtures show a relationship of green tensile and shear strength to moisture content similar to that in Figs. 1 and 2 for a specific southern bentonite sand mixture. Thus, the effect of moisture is to control the shear strength of the plastic mixture of clay and water in the aggregate which, in turn, largely determines the other green properties such as tensile and compressive strengths and mold hardness.

RAMMING EFFECT

The ramming effect on green tensile and shear strength was studied for mixtures of various clay contents and clay types. Variation in ramming was achieved through the use of AFS rammers equipped with two different ramming weights, 14 lb and 2 lb, dropped a distance of 2.0 in. Density of the 2.0 in. x 2.0 in. diameter specimen was varied from a lower to a higher value by varying the specimen weight and number of rams with each weight. Table 2 reports the green properties of mixtures of 8 per cent western bentonite, 92 per cent sand mixtures at 3.3 to 3.8 per cent H_2O, mixtures 3-8. Table 2 also lists 8 per cent western bentonite mixtures containing cellulose additives, mixtures 9 and 10. Three different western bentonites were used in the mixtures, as listed in Table 2.

The relationships of the properties are shown in Figs. 3 and 4. Figure 3 reveals that both green tensile and green compressive strengths are proportional to green shear strength up to a value of about 6 psi green shear strength. Figure 4 shows that green compressive strength and bulk density are related to average mold hardness. Obviously green shear strength is also related to mold hardness, although this has not been plotted. Referring to Fig. 3, increased ramming causes green tensile strength to reach a maximum of between 3 and 4 psi.

Above this level, increased ramming causes erratic green tensile strength to develop as the shear strength increases beyond 6 psi. Thus, green shear strength, green compression strength and mold hardness may continue to increase with more ramming after tensile strength reaches a maximum of about 3 to 4 p in the 8 per cent western bentonite mixtures.

The reason for the erratic values of both gree tensile and green compression strengths above 6 p green shear strength in the 8 per cent western be tonite mixture is associated with the density achieve during ramming. Below 6 psi shear strength, increa ing ramming to 3 standard rams is accompanied b shear strength of the clay-water bond, the expressio of voids, compaction of the sand and greater effecti contact area of the clay-water bond within the spec men. Above 6 psi green shear strength (or 154 gra weight and 3 standard rams in this case), fitting to gether of sand particles and particle to particle co tact occurs.

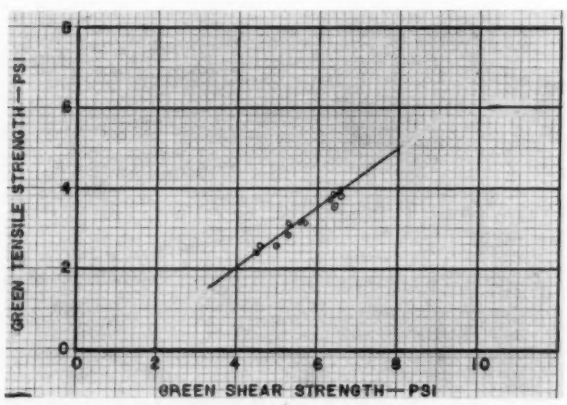


Fig. 2 — Relationship of green tensile strength to green shear strength of 8.0 per cent southern bentonite bonded sand at 2.60 to 4.40 per cent H_2O. Data of Table 1.

TABLE 2 — GREEN PROPERTIES OF 8% WESTERN BENTONITE MIXTURES

Mixture and Symbol	Spec. Wt. grams	No. Rams*	Green Strength, psi Comp.	Tens.	Shear	Mold Hardness**
3-o 8% W. bent.	148	2	15	2.65	4.6	88
92%-85AFS	154	3	18.6	3.50	5.9	90
sand,	161	7	32.5	4.3	6.8	94
3.3% H_2O	164	10	36.5	4.3	8.0	94.5
	167	12	41.0	4.1	9.1	95.5
4-● 8% W. bent.,	142.5	27*	11.1	1.9	3.6	83
92%-85AFS	147.5	45*	15.9	2.65	4.7	87.5
sand,	153.5	3	16.4	2.80	5.0	89.0
3.8% H_2O	160.5	8	30.0	3.50	7.1	93.5
	164.5	11	32.0	3.70	7.2	94.5
	167.5	14	35.0	3.60	8.2	95.0
5-△ 8% low gel						
(21 sec)	138	18*	8.8	1.7	3.1	80
W. bent.,	148	2	12.2	2.3	4.0	86
92.0%-85AFS	154	3	16.5	2.5	4.7	89
sand,	162	7	27	3.5	6.5	93.5
3.5% H_2O	167	11	32.5	3.7	7.1	94.5
	170	15	35.8	3.7	8.1	95.0
6-▲ 8% high gel						
(30 sec)	138	18*	9.5	1.7	3.20	81.5
W. bent.,	148	2	14.1	2.5	4.4	87.5
92.0%-85AFS	154	3	18	3.0	5.10	89.5
sand,	162	7	28.8	3.8	6.8	94
3.5% H_2O	167	11	35.5	3.8	7.6	94.5
	170	15	40.5	3.7	8.1	95.5
7-□ 8.0% W. bent.	142	15*	8.6	1.7	3.0	82.5
92%-49AFS						
Wedron	151.5	2	14.0	2.3	4.2	88
silica sand	156.5	3	18.4	2.8	5.2	90.5
3.40% H_2O	164.5	7	25.5	3.3	7.1	94.0
	168.5	12	30.5	3.2	7.0	94.5
8-■ 8% W. bent.	139	22*	10.5	1.95	3.6	81.5
92%-91AFS						
Portage	147.5	2	12.7	2.2	4.2	86.5
silica sand	153	3	15.1	2.55	4.6	88
3.40% H_2O	161	8	24.5	3.0	6.7	93
	166	15	32	3.1	7.2	95
9-× 8% W. bent.	139	22*	13.8	2.4	4.6	86.5
91%-85AFS sand	147	2	18	3.1	5.3	90
1.0%-carbonized						
cellulose	153	3	24	3.65	6.5	92
3.40% H_2O	161	8	41.5	5.55	8.5	95.5
	165	10	46.5	4.9	10	96
10-+ 8% W. bent.	141	21*	10.5	1.8	3.3	81.5
91%-85AFS sand	147	2	13.7	2.5	4.8	87
1.0% wood flour	153	3	19.0	3.1	5.3	90
3.9% H_2O	161	6	29	3.8	7.3	93
	168	13	39	3.7	7.7	95

*No. rams marked with asterisk refer to 2.0 lb-2.0 in. rams, others refer to standard 14 lb-2.0 in. ram.
**Ave. of 6 readings, 3 top and 3 bottom.

Particle packing with increased ramming is accompanied by a rise in green compressive strength, but no rise in green tensile strength at the 8 per cent clay level in these mixtures. The rise in green compressive strength is due to particle friction on shear planes. The lack of further rise in green tensile strength is due to the fact that maximum contact area of clay-water bond has occurred when the 6 psi shear strength value is reached, and further ramming or density increase is unable to develop more contact area. The ability to develop more contact area for shear strength hinges on the clay content of the sand.

With less clay, the shear strength for maximum tensile strength will be lower, and with higher clay contents a higher shear strength and tensile strength will be reached. This principle is demonstrated in 10 to 12 per cent western bentonite and 6 per cent western bentonite mixtures listed in Tables 3 and 4, respectively.

The data for 10 and 12 per cent western bentonite mixtures are shown in Figs. 5 and 6. Figures 5 and 6 show that green tensile strength increases uniformly as the shear strength increases due to ramming to higher bulk densities and mold hardness. At the 10 to 12 per cent western bentonite level, tensile strength does not level off at a fixed value as in the case of the 8 per cent western bentonite, even when rammed to average mold hardness of 95. The higher clay content mixtures are regarded as clay-saturated

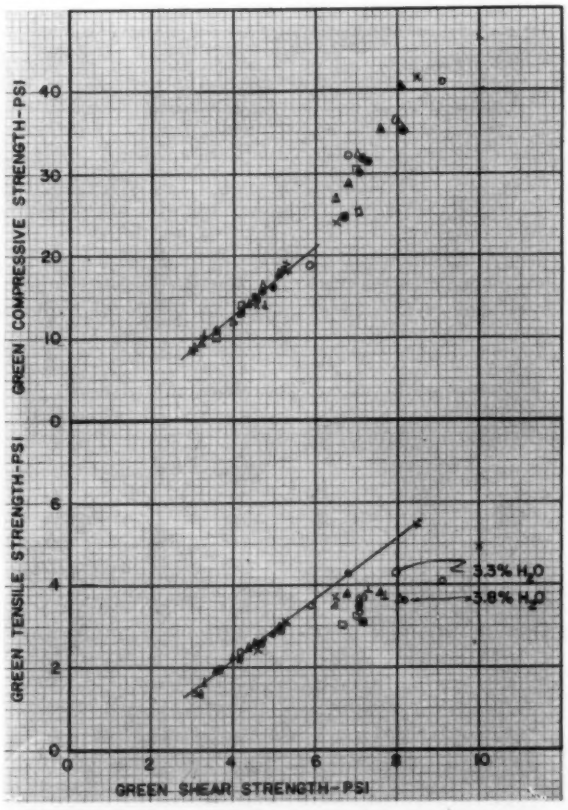


Fig. 3 — Relationship of green tensile and compressive strength to green shear strength in a sand bonded with 8 per cent western bentonite. Data of Table 2.

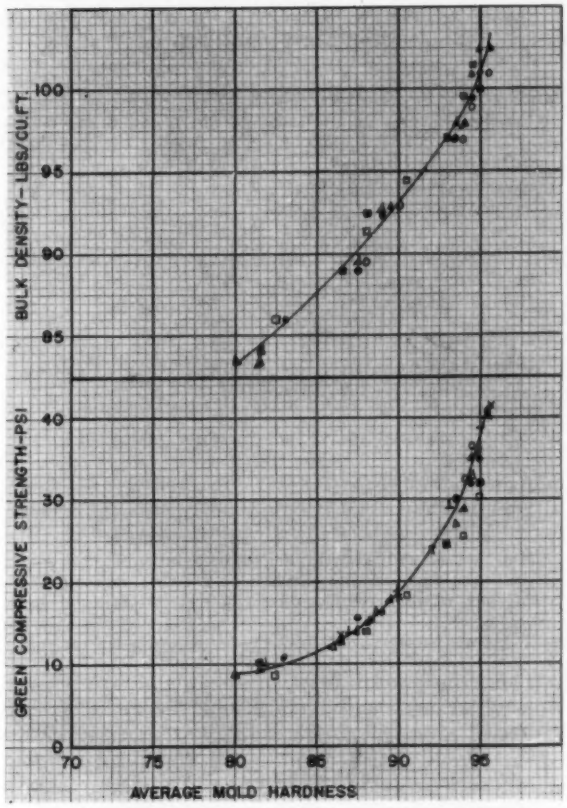


Fig. 4 — Relationship of green compressive strength and bulk density to mold hardness of sands, listed in Table 2 and Fig. 3.

sands, i.e., sands in which the addition of more clay does not produce a further increase in strength at a given hardness. This concept has been defined and demonstrated.[2]

TABLE 3 — GREEN PROPERTIES OF 10 AND 12% WESTERN BENTONITE MIXTURES

Mixture and Symbol	Spec. Wt., grams	No. Rams*	Green Strength, psi			Mold Hardness**
			Comp.	Tens.	Shear	
12-o	141.5	23	10.7	2.0	4.2	82
12% W. bent.	149.5	2	14.4	2.4	5.1	86
88%-68AFS Silicon	155.5	3	18.5	2.9	6.2	88
sand	163.6	6	30	4.2	8.7	92
4.8% H_2O	168	8	33.5	4.8	9.9	93.5
	172	12	39	5.4	10.2	95
13-●	131	13*	8.6	N.G.	3.2	78
12% W. bent.	135	20*	11.5	2.0	4.0	85
88%-85AFS	139	27*	14.1	2.4	4.9	85
sand	147	2	18.0	2.8	6.0	88
4.5% H_2O	153	3	22.5	3.4	7.2	90
	161	5	31.5	5.2	10.1	93
	165	7	40.5	5.8	11.0	94
14-▲	140	28*	13.0	2.2	4.5	83.5
10% W. bent.	147	2	15.1	2.4	5.3	86
90%-85AFS sand	153	3	19.5	3.1	6.4	89
4.1% H_2O	162	7	34	4.8	9.4	93.5
	164	9	38.5	5.0	9.7	94.5
	170	11	5.3	5.3	10.7	N.D.

*Rams marked with asterisk refers to number of rams with a 2.0 lb weight dropped 2.0 in.; Other refer to standard 14-lb ram.

**Ave. of 3 readings on top and 3 readings on bottom.

TABLE 4 — GREEN PROPERTIES OF 6% WESTERN BENTONITE MIXTURES

Mixture and Symbol	Spec. Wt., grams	No. Rams*	Green Strength, psi			Mold Hardness**
			Comp.	Tens.	Shear	
15-×	148	2	16.5	2.8	4.7	87.5
6.0% W. bent.	154	3	18.4	2.8	6.0	91
94.0% 85AFS sand	165	11	36	2.9	9.5	95
2.60% H_2O	167	15	38	2.65	10.5	95.5
	170	19	41	2.35	11.5	95.5
16-o 6.0% W. bent.	145	28*	11.8	2.65	3.5	85.5
94.0% 85AFS sand	155	3	13.8	2.3	4.1	89
2.90% H_2O	165	12	26.6	2.6	7.3	95
	160	14	28.6	2.6	8.7	95
17-● 6.0% W. bent.	156	3	11.5	1.85	3.2	86.5
94.0% 85AFS sand	161	6	15.4	2.2	4.0	90.5
3.50% H_2O	166	9	17.3	2.3	4.6	92
18-△ 6.0% W. bent.						
94.0% 85AFS sand	157	3	11.0	1.85	3.1	86.5
3.8% H_2O						
19-▲ 6.0% W. bent.	157.5	3	9.5	1.65	2.9	85.5
94.0% 85AFS sand	167.5	10	15.2	2.1	3.5	91
4.1% H_2O	171	15	18.5	2.1	4.5	92

*Rams marked with asterisk refer to the number of 2.0 lb-2.0 in. rams; others refer to the standard 14 lb-2.0 in. rams.

**Ave. of 3 readings on top and 3 readings on bottom of specimen.

The relationship between maximum green compressive strength and average mold hardness in clay-saturated sands has been proved.[2] This relationship of maximum green compressive strength to mold hardness is satisfied in the 10 and 12 per cent western bentonite mixtures in Fig. 6 (lower curve), but n in the 8 per cent western bentonite mixture of Fig. (lower curve).

Since the higher clay content sands are clay-satu ated, it is possible to achieve maximum contact ar for shear strength within the mass and, therefo higher tensile strength is obtained at the higher de sities. The effect of higher clay content in clay-satu ated sands then is to permit the sand to be ramme to higher tensile and shear strengths and mold har ness.

A comparison of lower clay content sands wit those of higher clay contents can be made with th data in Table 4 for 6 per cent western bentonite mix tures. These data are plotted in Figs. 7 and 8. Figur 7 shows that tensile strength is proportional to shea strength, but reaches a maximum at about 4.5 to 5.0 psi shear strength or below, and then levels off. In Fig. 7, a solid line curve is reproduced from Fig. 3 for 8 per cent western bentonite mixtures. Comparison of the data on Fig. 7 show that at low shear strength, tensile strength for the two different clay levels is virtually the same.

However, as green shear strength increases with increasing ramming, the tensile strength levels off at between 2.0 and 3.2 psi tensile strength beyond 4.5 psi shear strength for the 6 per cent western bentonite mixture. At 8 per cent western bentonite, tensile strength levels off at 3.0 to 5.3 psi beyond 6 psi shear strength. In mixtures containing 10 to 12 per

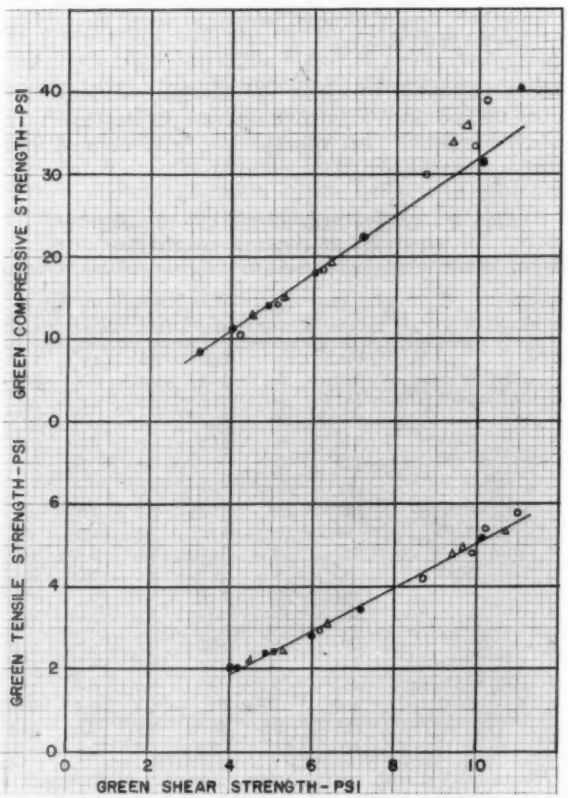


Fig. 5 — Relationship of green tensile and compressive strength to green shear in sand bonded with 10 to 12 per cent western bentonite. Data of Table 3.

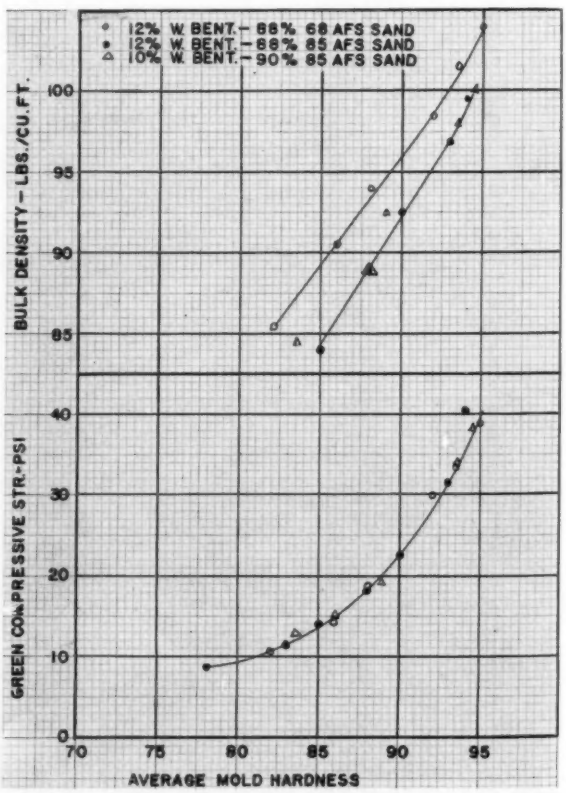


Fig. 6 — Relationship of green compressive strength and bulk density to mold hardness of sands, listed in Table 3 and Fig. 5.

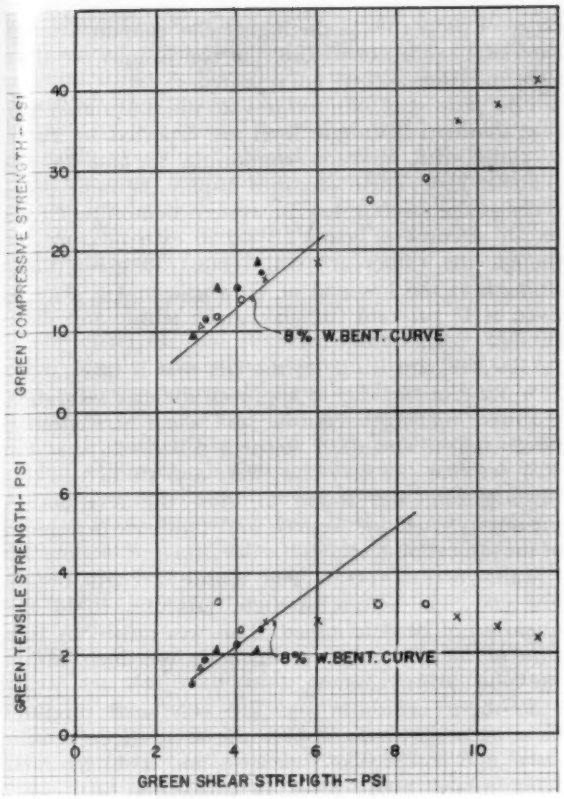


Fig. 7 — Relationship of green tensile and green compressive strength to green shear strength of sands bonded with 6.0 per cent western bentonite. Data of Table 4.

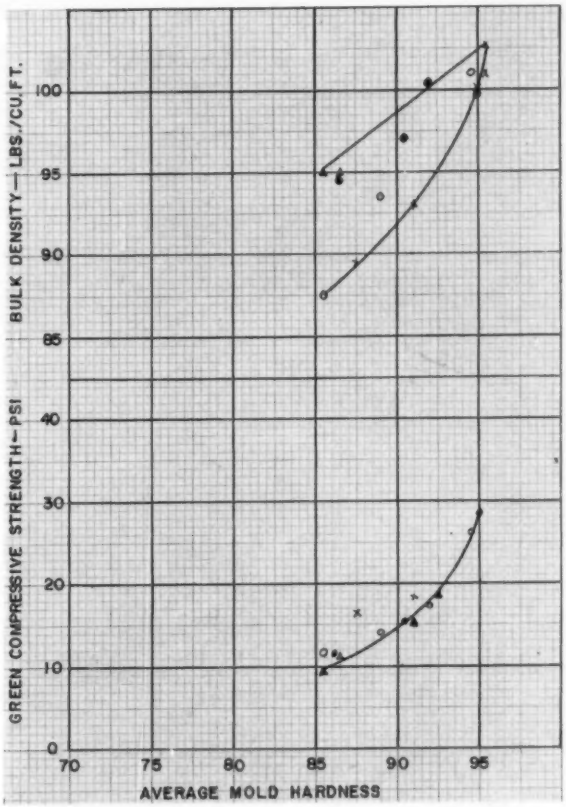


Fig. 8 — Relationship of green compressive strength and bulk density to mold hardness of sands, listed in Table 4 and Fig. 7.

cent western bentonite, tensile strength does not level off even at 5.8 psi tensile strength and 11.0 psi shear strength, the limit of these tests. Further, it may be noted in Fig. 7, that tensile strength of the 6 per cent western bentonite mixtures actually falls off beyond 6 psi shear strength with increasing ramming.

This effect is due to elastic rebound, caused by ramming when particle packing and particle contact begins within the specimen. In fact, some test specimens rammed to high density will actually fail in tension during ramming. Elastic rebound causes the specimen to fracture during the ramming operation, and it is not possible then to get a tensile strength test result. Thus, the effect of ramming cannot be considered independently of the clay content of the sand.

CLAY TYPE EFFECT

Southern bentonite and fire clay bonded sands were studied in the same way. Table 5 reports data for several mixtures containing 8 per cent southern bentonite, alone and with additives. The data in Table 5 are plotted in Figs. 9 and 10. Figure 9 shows the same relationship reported earlier for 8 per cent western bentonite mixtures, Fig. 3. Since the relationships of green shear, tensile and compression strengths, mold hardness and bulk density are the same as those reported for western bentonite earlier, the conclusions are the same and will not be repeated.

Some 10 and 12 per cent southern bentonite mixtures with 85 AFS sand were also tested. The points of data relating green shear strength with tensile and compressive strength are plotted in Fig. 11, and com-

TABLE 5 — GREEN PROPERTIES OF 8% SOUTHERN BENTONITE MIXTURES

Mixture and Symbol	Spec. Wt., grams	No. Rams*	Green Strength, psi			Hardness**
			Comp.	Tens.	Shear	
20-△.	132	15*	9.3	1.6	3.2	79
8% S. bent.	135	1	10.1	1.8	3.5	80.5
92% 85 AFS sand	149.5	3	21.3	3.6	6.3	91
3.2% H_2O	160	7	33.7	5.4	8.5	94.5
21-▲.	142	2	13.0	1.9	3.8	84.5
8.0% S. bent.	155	3	14.0	2.3	4.1	85.5
92% 85 AFS sand	165	10	29.5	3.3	6.9	94.0
4.60% H_2O						
23-o.	136	26*	12.0	2.0	3.8	84.5
8.0% S. bent.	144	2	14.1	2.3	4.3	86.5
91% 85 AFS sand	149	3	20.5	2.9	5.8	90
1% carbonized	154	9	34	4.15	8.3	94
cellulose	164	13	43	4.20	8.4	95.5
3.8% H_2O						
24-●.	135	19*	10	1.7	3.3	83
8.0% S. bent.	142	2	14.7	2.3	4.2	87
91% 85 AFS sand	147	3	18.3	2.7	4.9	89
1% wood flour	155	5	24.5	3.2	5.9	92.5
4.1% H_2O	160	10	35	3.5	8.2	94.5
	165	17	42.5	3.4	8.7	95.5

*Rams marked with asterisk refer to number of 2.0 lb-2.0 in. rams; others refer to standard 14 lb-2.0 in. rams.

**Average of 3 readings on top and 3 readings on bottom of specimen.

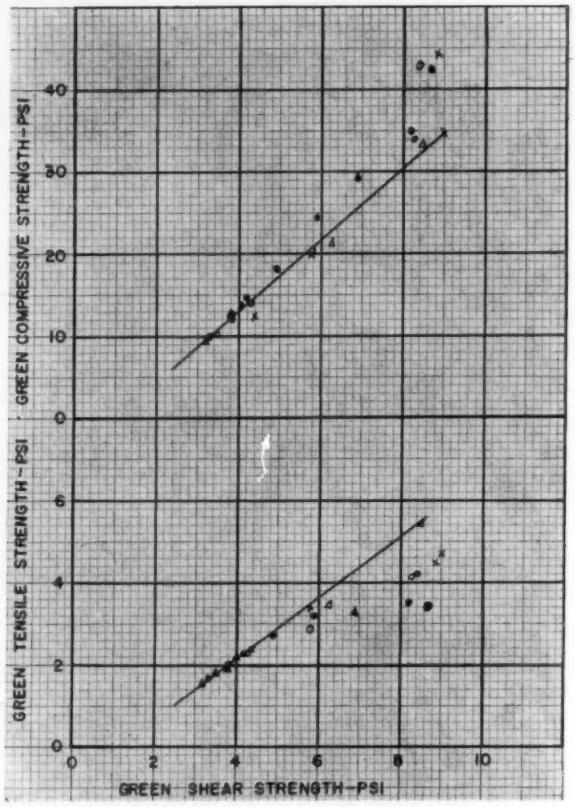


Fig. 9 — Relationship of green tensile and green compressive strength to green shear strength of sands bonded with 8 per cent southern bentonite. Data of Table 5.

pared with the curves for western bentonite. Figure 11 shows that this higher clay content southern bentonite mixture displays the same relationships and

TABLE 6 — GREEN PROPERTIES OF SAND MIXTURES CONTAINING SOUTHERN BENTONITE OR FIRECLAY

Mixtures	Spec. Wt., grams	No. Rams*	Green Strength, psi Comp.	Tens.	Shear	Ave. Mold Hardness**
25. 4% S. bent.						
96% 85AFS sand	155	3	10.8	1.4	2.5	87.5
2.50% H_2O	165	12	18.7	1.6	3.7	93.5
26. 4% S. bent.	152	2	15.5	2.0	3.1	90
96% 85AFS sand	157	5	19.5	2.4	3.7	93.5
2.0% H_2O	160	8	22.5	2.2	3.85	94
27. 6% S. bent.						
96% 85AFS sand	150	3	18	2.3	4.1	89
2.6% H_2O	155	5	22	2.5	4.6	92
28. 10% fire clay	168	3	9.5	Rebound failure	2.4	86
90% 85AFS sand						
3.8% H_2O	178	12	17	1.5	3.2	93
29. 10% fire clay	170	3	6.6	Rebound failure	1.8	81
90% 85AFS sand						
4.7% H_2O	180	9	10.3	1.5	2.4	87
30.	177	3	7.2	Rebound failure	2.7	84.5
15% fire clay						
85% 85AFS sand	184	6	9.5	1.5	3.4	87.5
5.8% H_2O	189	11	15.9	1.7	3.6	90

*Number of standard 14 lb-2.0 in. rams.
**Ave. of 3 readings on top and 3 readings on bottom of specimen.

range of properties as the higher clay content western bentonite sands. Therefore, the conclusions are held to be the same, and are not repeated.

The properties of mixtures at the 4.0 and 6 per cent southern bentonite level are reported in Table 6, along with mixtures containing 10 and 15 per cent fire clay. These mixtures display the same properties as the mixture low in western bentonite. Green tensile strength is below 2.5 psi in each mixture, although the values are still related to green shear strength at the low end of the curve in Fig. 3 or 9. Even at 15 per cent fire clay, green tensile and shear strengths remain low.

However, this is due to the fact that it takes about 25.0 per cent fire clay to approach the clay saturation required for higher tensile and shear strengths.[2,3] Green tensile and shear strength relationships at the 25.0 per cent fire clay level were also studied in this investigation, and found to show the relationship previously cited for western and southern bentonites at the saturation level.

Table 6 also reveals that tensile strength of the low clay, low tensile strength mixtures could frequently not be determined. This was caused by rupture of the 2.0 x 2.0 in. diameter specimen within the tube during ramming.

Southern bentonite and fire clay sand mixtures have been shown to display the same green tensile, shear and compression strengths and mold hardness relationships, as discussed in connection with west-

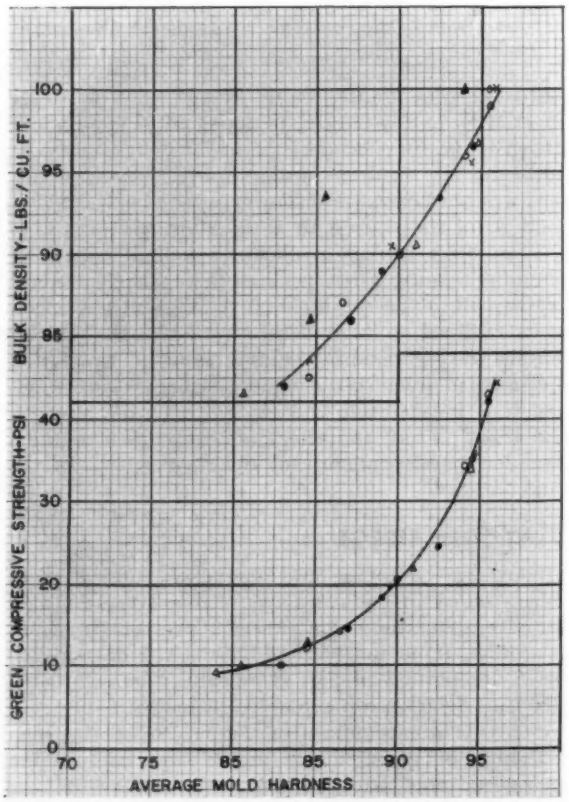


Fig. 10 — Relationship of green compressive strength and bulk density to mold hardness of sands, listed in Table 5 and Fig. 9.

ern bentonite. Southern and western bentonite display almost identically equivalent green properties, as pointed out in Figs. 9-11. This equivalence is also displayed by fire clay, but requires a substantially higher clay percentage. A total of ten different commercially available clays were tested, and found to show the same relationships. However, other investigators[4] have stated that there are some clays which display different relationships although the principles are the same.

BASE SAND EFFECT

The effect of sand fineness and sieve analysis on green tensile and shear strengths was investigated within present practical limits. Sands varying in fineness from AFS 49 to AFS 85, are represented in Tables 1 through 6. Silica and bank sands are also included. However, all these sands were of the 3 to 5 screen sieve distribution and sub-angular shape common to many foundry sands. Data on the standard AFS 50 sands have been reported,[4] and these data show that the same principles apply. However, no sands with an abnormally high percentage of fines were studied.

With reference to fineness and sieve analysis, no effects on the relationship of green tensile, shear strength, etc., were noted. This seems likely, especially at 6 per cent or more of bentonite, since the green properties are largely determined by the shear strength of the clay-water mass separating the sand grains. The dominating effect of the clay-water cement increases at higher clay contents. However, at low clay content the sieve analysis might have a noticeable effect. The low clay content mixtures at 4.0 per cent and below were not studied for such an effect.

TENSILE TEST LOADING RATE

The rate of loading can affect the values obtained in tensile testing. Two rates were studied in the investigation. The data in Tables 1 through 6 and Figs. 1-11 were obtained with a loading rate of 0.32 lb/sq in./sec, using a machine of the make shown on p. 92 of the AFS SAND HANDBOOK. A lower rate of loading of 0.062 lb/sq in./sec was found to develop tensile strengths about 10 per cent higher than those reported in this work. Slippage of the specimen in the specimen tube occurred during the tests with the slower rate of loading.

CELLULOSE ADDITIVES EFFECT IN FOUNDRY SANDS

Mixtures 9, 10, 23 and 24, reported in Tables 2 and 5, contained cellulose additives. Mixtures 9 and 23 both contained 1 per cent carbonized cellulose and 8 per cent western or southern bentonites, respectively. Mixtures 10 and 24 both contained 1 per cent wood flour and 8 per cent western or southern bentonites, respectively. The green tensile, shear and compression strengths and mold hardness relationships were not affected by the additives present. However, the carbonized cellulose permitted a higher tensile strength to be reached by ramming.

A number of foundry sands containing additives also were studied, with no deviation from the principles studied. Typical values for a foundry sand based on a mixture containing 6 per cent bentonite, 2 per cent sea coal, 1 per cent carbonized cellulose and 85 AFS sand are cited (Table 7). Comparison of the data in Table 7 for a foundry sand should be made with the data for a new 8 per cent western bentonite mixture in Figs. 3 and 4. Such comparison shows that the combination of sea coal and carbonized cellulose has the effect of causing a 6 per cent western bentonite mixture to have properties similar to an 8 per cent mixture.

This is probably due to the contribution of the additives to extending or increasing the effective amount of plastic bonding from the clay-water additive mixture. Generalizing, the additives studied do not alter the principles of green property relationships in foundry sands.

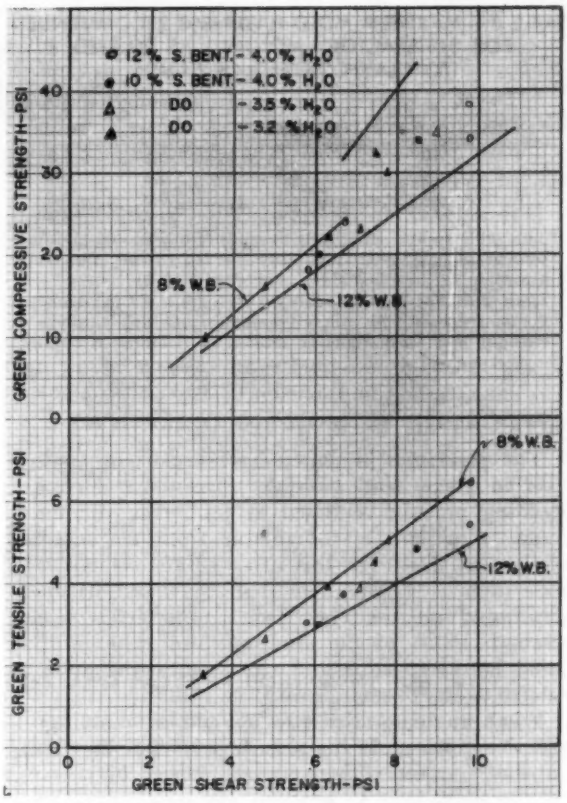


Fig. 11 — Relationship of green tensile and green compressive strength to green shear strength of sands bonded with 10 to 12 per cent southern bentonite.

TABLE 7 — GREEN PROPERTIES OF A FOUNDRY SAND

Mixture	Spec. Wt., grams	No. Rams*	Green Strength, psi			Ave. Mold Hardness
			Comp.	Tens.	Shear	
6% W. bent.	134	20*	9.8	1.55	3.0	82.5
2.0% sea coal	142	2	12.1	2.1	3.9	87
1.0% carbonized	148	3	16.5	2.6	4.4	88.5
cellulose	156	7	27.5	3.6	6.5	93
3.3% H_2O	159	9	33'	3.6	7.3	94
	162	15	40.5	3.6	8.7	95.5

*No. of 2 lb-2.0 in. rams; all others are 14 lb-2.0 in. rams.

TABLE 8 — MAXIMUM MOLD HARDNESS RECOMMENDED FOR SATISFACTORY MOLDING AS RELATED TO CLAY CONTENT OF SAND*

Clay, %		Limiting Mold	Limiting Strength, psi	
Bentonites	Fire clay	Hardness**	Shear*	Tensile*
4	10	80 - 85	3.5	1.8
6	15	85 - 89	4.0	2.5
8	—	89 - 91	6.0	3.6
10-12	—	93 - 95+	11.0	5.8

*This table applies specifically to sands of low enough moisture content to have good flowability (i.e., 10 to 30% free water according to the calculation method of Ref. 1).

**Limit does not refer to standard AFS 2.0 x 2.0 in. diameter, 3 ram specimen, but refers to ramming to this hardness or strength level with whatever amount of ramming is required in specimen or mold.

APPLICATION TO MOLDING OPERATIONS

The principles presented have important application in green sand molding operations. Currently, molding practice is aimed at producing the maximum mold hardness obtainable from the molding equipment. However, this investigation demonstrates that there is a maximum desirable mold hardness, depending on the clay content (or green tensile strength limit) of the sand mixture. The limiting mold hardness, shear and tensile strengths limit, as related to clay content, is listed in Table 8.

When ramming (by any method of molding) is practiced to a mold hardness higher than that listed in Table 8, the consequence is danger of cracking of the mold due to residual stress or spring back from the applied molding forces. For example, if a molding sand containing 4.0 per cent clay is molded on powerful molding equipment to hardnesses over 90, the mold is likely to crack, drop or hang-up on the pattern. Unless such a low clay content sand is used at a high moisture content where it is quite sticky (50 to 100 per cent more than calculated moisture[1]), these molding problems will develop.

At high moisture contents it is not possible to achieve the higher mold hardnesses, because the mixture will not develop sufficient compressive strength (or shear strength) to cause a high hardness regardless of the amount of power applied. The sand may then have sufficient deformation to relieve residual stresses, and therefore, the mold may not crack although drops can still occur due to low tensile strength.

On the positive side, the data in Table 8 show that a clay-saturated sand is one that is best suited for producing high hardness molds (90 to 95 mold hardness) and high strength molds. Powerful molding equipment can then be used to develop high strength molds, which can be drawn from patterns without mold failure. The sand can develop adequate tensile strength for pattern drawing, and residual stresses from molding forces will not cause cracking of the mold.

There are of course other reasons for an adequate amount of clay used in the sand, which are concerned with the behavior of the mold when the metal is poured. However, from the standpoint of green tensile strength, there is no gain in ramming low clay content sands to high mold hardness, since no significant increase in tensile strength is obtained. This ramming limitation does not apply to the high clay content sands.

REFERENCES

1. R. W. Heine, E. H. King and J. S. Schumacher, "How to Determine the Moisture Requirements of Molding Sands", AFS TRANSACTIONS, *vol. 65*, p. 118, 1957.
2. R. W. Heine, E. H. King and J. S. Schumacher, "Mold Hardness: What It Means", AFS TRANSACTIONS, *vol. 65*, p. 123, 1957.
3. A. H. Zrimsek and R. W. Heine, "Clay, Fines, and Water Relationships for Green Strength in Molding Sands", AFS TRANSACTIONS, *vol. 63*, p. 575, 1955.
4. A. H. Zrimsek, private discussion and Magnet Cove Barium Corp. Technical Data No. 3: "Compression, Shear, or Tensile?", a private company bulletin.
5. AFS, FOUNDRY SAND HANDBOOK, 1952 Edition.

HOT CRACKING TEST FOR LIGHT METAL CASTING ALLOYS

By E. J. Gamber

ABSTRACT

A simple and reproducible test for determining the relative resistance to hot cracking of aluminum and magnesium casting alloys is described. This test is capable of assessing a wide range of cracking susceptibility, yet it possesses excellent sensitivity to small differences in cracking tendency. It is based on the fact that the probability for hot cracking to occur is greatest at a sharp internal angle, and diminishes as fillet radii of increasing sizes are provided.

Small specimens, termed "U" castings, are cast in a dry sand mold. The castings have the shape of thin elongated channel sections with short upright ends. Individual molds containing eight castings having variations in fillet radius or casting length produce a wide range of conditions governing hot cracking susceptibility.

Relative hot cracking ratings obtained from tests on various commercial aluminum and magnesium casting alloys are in good agreement with foundry experience, and appear superior to ratings based on cooling curve analysis or various ring mold tests employed by other investigators.

INTRODUCTION

A potential hot cracking problem exists whenever contraction is restrained while castings cool in molds. Both designers and foundrymen recognize the problem, and exercise known preventive measures such as providing generous fillet radii and draft angles, avoiding abrupt changes in section sizes and controlling solidification patterns and thermal gradients. Stresses that result from restrained contraction can be minimized in sand casting operations by selecting proper mold materials, and in permanent mold or die casting operations by controlling removal or ejection times.

Although design and operational precautions are effective, they will prove inadequate unless the proper alloy has been selected. Consideration also must be given to the cracking susceptibility of alloys.

Since hot cracking has been a problem frequently encountered by foundrymen, it has been the subject of numerous investigations. Various testing methods have been devised to study the influence of contributing factors.

The existing tests had limited ranges, low sensitivity or provided poor correlation with experience, therefore, they were inadequate. A test was desired that would enable the evaluation of many alloys having a wide range in resistance to hot cracking, yet one that would possess the sensitivity required for evaluating the effects of composition variations in specific alloys.

A testing method was developed which satisfied these requirements. It has been used successfully to evaluate the resistance to hot cracking to both aluminum and magnesium casting alloys, and has provided a reliable and useful tool for alloy development work.

TEST AND PATTERN EQUIPMENT DESCRIPTION

The hot cracking test selected consists of casting eight small specimens in a single dry sand mold. In Fig. 1, a test casting, termed a "U" casting, is shown. The rectangular cross-sections are ¾-in. wide and ⅜-in. thick, and the length of the castings can be varied from 2 to 8 in. Fillet radii are ¾-in. at one end, and vary at the other end from ¾-in. to a sharp corner. The severity of the test depends upon both the fillet radius and the length of the casting. Tendency to crack is more pronounced with smaller radii and greater length.

Figure 2 shows the design of the patterns for test castings. A total of 14 aluminum patterns are used. These consist of eight that provide an 8-in. length but variable fillet radii, and six that provide a fillet radius of ¾-in. but variable lengths ranging from 2 to 7 in. Although only eight patterns are used for each mold, the greater number available permits selection of combinations depending upon the anticipated hot cracking resistance of alloys. Since eight castings are produced from each mold, a range of conditions can be surveyed rapidly and conveniently.

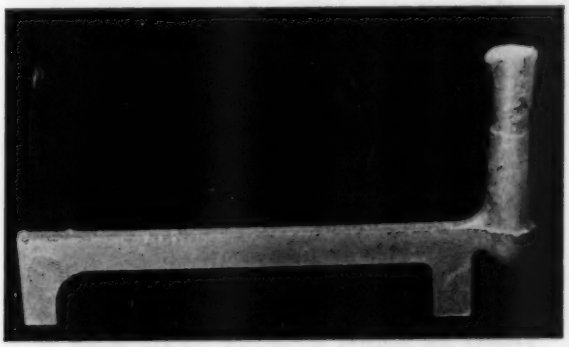


Fig. 1 — A "U" casting hot cracking specimen.

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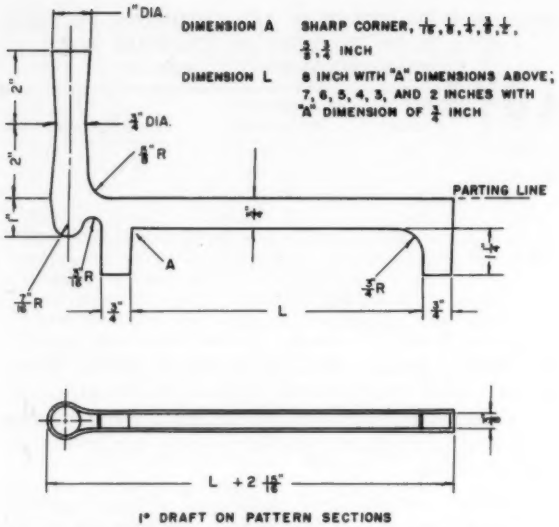


Fig. 2 — The design of patterns for the hot cracking test castings.

Figure 3 shows the patterns mounted 2 in. apart in an 18 in. long, 12 in. wide and 4 in. deep aluminum core box. The equipment was designed so the test castings are produced in the drag section of the mold. A sprue that also serves as a riser is located in the cope. The patterns can be easily removed from the box and are interchangeable, so that combinations including any eight of the 14 patterns can be used.

Test Procedure

Molds are made from carefully controlled baked core sand. The core mix selected should be strong to resist contraction stresses, yet have good collapsibility so castings can be easily removed from the mold without causing further cracking. A mix consisting of round grained washed silica sand bonded with urea formaldehyde is used. The sand should have an AFS grain fineness number of about 85. After baking, the core sand molds should exhibit a surface scratch hardness of about 90.

For the tests to be described, molds were prepared on a jolt, roll-over and draw machine. Use of the machine provided hard uniform packing of the sand and assured a straight even draw that was essential to reproduce the fillet radii accurately.

During preparation of the drag, small cast iron chills are inserted at the side of the casting cavity. These cool the area at the intersection of the cross arms and the ends located away from the sprue. Through chilling in conjunction with the pouring and risering system employed, a solidification pattern was established which concentrated contraction stresses at the variable radii fillet. Since the combination sprue and riser is located near the variable fillet, that section remains hottest and therefore weakest during the solidification process.

Figure 4 shows an assembled mold. Usually three molds were cast of each alloy. Each of the test castings was individually poured. Heat losses during pouring were minimized by filling the cavities in successive order as rapidly as possible.

Aluminum Base Alloy Castings

Aluminum-base alloy castings were poured at 1350 F, and magnesium-base alloys were poured at 1425 to 1450 F. Castings were allowed to cool to room temperature before they were removed from the mold and then inspected for cracks. Occurrence of cracks was detected with the aid of a liquid dye penetrant, and also with a low power magnifying glass. A rating indicating resistance to hot cracking was assigned to the alloys on the basis of the most severe condition tolerated without cracking. No consideration was given to the length of the cracks.

The numerical system, shown in Table 1, was adopted to eliminate the necessity of specifying length and radius dimensions in reporting test results. Ratings were based on experience that indicated greater

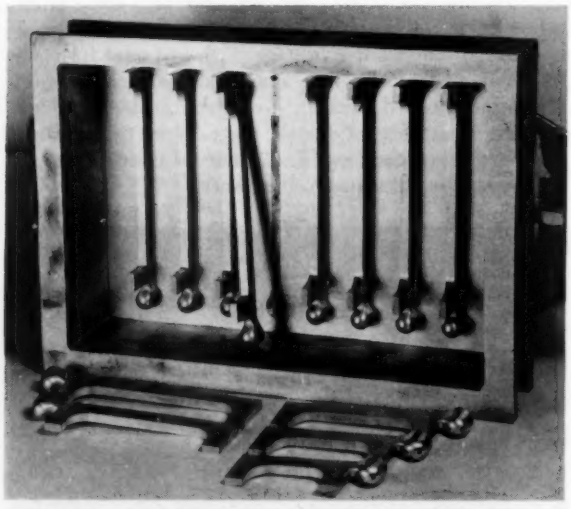


Fig. 3 — Core box used for production of the drag section of the hot cracking test mold. Interchangeable patterns are mounted in the core box and appear in the foreground.

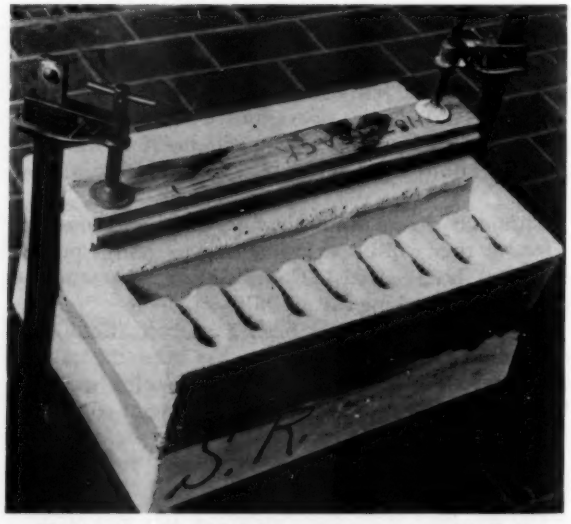


Fig. 4 — An assembled hot cracking test mold. Each of eight test castings was individually poured. The overflow basin adjoining the sprue cups was provided to facilitate rapid pouring.

differences in resistance to hot cracking were depicted by varying fillet radii from ¾-in. to a sharp corner than by changing casting lengths from 8 to 2 in.

Test Results

Laboratory test results demonstrated the usefulness of the "U" casting test for comparing the hot cracking resistance of various commercial aluminum and magnesium casting alloys, as well as determining the effect of composition variations on specific alloys. The results, shown in Fig. 5, illustrate the ability of the test to detect the effect of composition variations in a specific alloy. Progressive improvements in hot cracking rating were obtained when silicon content of 195 alloy was increased from 0.2 to 3.0 per cent.

The improvement shown is in good agreement with foundry experience with sand casting alloy 195 (Al; 4.5 per cent Cu; 0.8 per cent Si) and permanent mold alloy (Al; 4.5 per cent Cu; 2.5 per cent Si). The higher silicon content of B195 alloy imparts greater hot cracking resistance for permanent mold castings. Composition limits shown by shaded areas, in Fig. 5, indicate the silicon content of the two alloys.

Ratings obtained and the nominal compositions of commercial aluminum and magnesium casting alloys subjected to "U" casting hot cracking tests are summarized in Tables 2 and 3. The alloys tested represent most of the types that are used in present-day foundry operations. The results demonstrate the ability of the test to differentiate between alloys with a wide range of resistance to hot cracking.

Of the aluminum alloys listed in Table 2, alloys 43, 333, 355, 356 and F132 displayed the greatest resistance to hot cracking, and alloys 195 and C612 exhibited the lowest resistance. The hot cracking ratings of aluminum alloys containing copper and silicon, 195, 108, B195, 319, 333 and F132, showed progressive improvement with increasing silicon content. In Fig. 6, the relationship between the hot cracking ratings for those alloys and their silicon content is shown.

Among the magnesium casting alloys tested, alloys EK41A and EZ33A exhibited the greatest resistance to hot cracking, whereas alloys AZ63A and ZK61A exhibited the lowest resistance. Of the magnesium alloys

TABLE 1 — HOT CRACKING RATING SYSTEM

Smallest Fillet Radius Not Cracked, in.	Span Length, in.	Hot Cracking Rating
Sharp Corner	8	1.0
1/16	8	2.0
1/8	8	3.0
1/4	8	4.0
3/8	8	4.5
1/2	8	5.0
5/8	8	5.5
3/4	8	6.0
3/4	7	6.3
3/4	6	6.6
3/4	5	7.0
3/4	4	7.3
3/4	3	7.6
3/4	2	8.0
3/4 Cracked	2	9.0

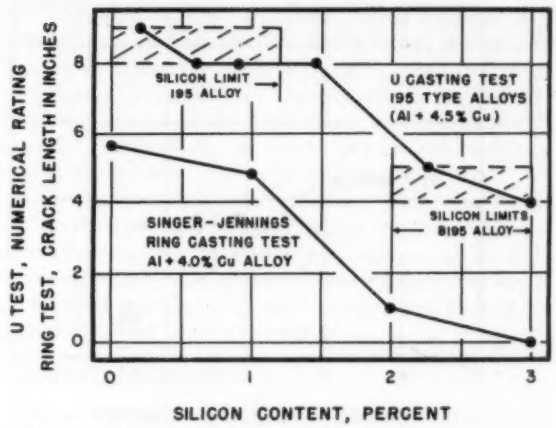


Fig. 5 — Silicon content effect on resistance to hot cracking of aluminum alloys containing 4.0 or 4.5 per cent copper measured by two test methods.

containing aluminum, AM100A was rated somewhat better than AZ92A and considerably better than AZ-63A. Alloy HK31A was rated slightly better than AM100A and considerably better than HZ32A.

DISCUSSION

The "U" casting hot cracking test provides a simple and reliable means for evaluating the relative resis-

TABLE 2 — ALUMINUM CASTING ALLOY HOT CRACKING RATINGS AND NOMINAL COMPOSITIONS

Alloy	Composition, %* Cu	Si	Mg	Zn	Ni	Hot Cracking Rating**
43	—	5.0	—	—	—	1.0
F132	3.0	9.0	1.0	—	—	1.0
356	—	7.0	0.3	—	—	1.3
333	3.8	9.0	—	—	—	1.3
355	1.3	5.0	0.5	—	—	1.7
220	—	—	10.0	—	—	2.7
319	3.5	6.3	—	—	—	3.5
108	4.0	3.0	—	—	—	4.5
B195	4.5	2.5	—	—	—	4.5
142	4.0	—	1.5	—	2.0	5.0
C612	0.5	—	0.35	6.5	—	6.6
195	4.5	0.8	—	—	—	8.0

*Alloying elements — Aluminum, and normal impurities constitute the remainder.

**Smallest number indicates greatest resistance to hot cracking.

TABLE 3 — MAGNESIUM CASTING ALLOY HOT CRACKING RATINGS AND NOMINAL COMPOSITIONS

Alloy	Nominal Composition, %* Al	Mn (Min.)	Zn	Zr	Rare Earths	Th	Hot Cracking Rating**
EK41A	—	—	—	0.6	4.0	—	3.0
EZ33A	—	—	2.5	0.6	3.3	—	3.5
HK31A	—	—	—	0.7	—	3.2	4.2
AM100A	10.0	0.10	—	—	—	—	4.4
AZ92A	9.0	0.10	2.0	—	—	—	4.9
HZ32A	—	—	2.1	0.7	—	3.2	5.5
AZ63A	6.0	0.15	3.0	—	—	—	7.5
ZK61A	—	—	6.0	0.7	—	—	7.5

*Alloying elements — Magnesium, and normal impurities constitute the remainder.

**Smallest number indicates greatest resistance to hot cracking.

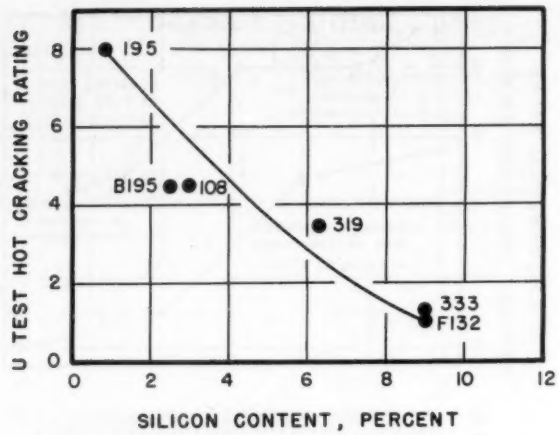


Fig. 6 — Silicon content effect on the resistance to hot cracking of commercial aluminum alloys containing 3.5 to 4.5 per cent copper.

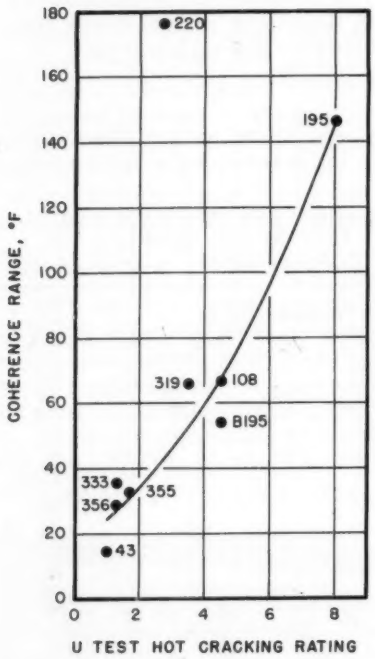


Fig. 7 — Comparison between "U" test ratings for commercial aluminum casting alloys and ratings based on cooling curve analyses.

TABLE 4 — "U" CASTING RATINGS AND MODIFIED RING TEST RATINGS COMPARISONS, MAGNESIUM CASTING ALLOYS

"U" Casting Tests		Modified Ring Tests**	
Alloys in Order of Resistance To Hot Cracking	Hot Cracking Rating*	Alloys in Order of Resistance To Hot Cracking	Proposed Classification*
AK41A	3.0	AM100A	1A
EZ33A	3.5	EK41A	1B
HK31A	4.2	AZ92A	1B
AM100A	4.4	EZ33A	2
AZ92A	4.9	HK31XA	2
HZ32A	5.5	AZ63A	2
AZ63A	7.5	HZ33A	3
ZK61A	7.5	ZK61A	3

*Smallest number indicates greatest resistance to hot cracking.
**Ref. 5.

tance to hot cracking of both aluminum and magnesium casting alloys. The ability to survey a wide range of testing conditions with one group of castings offers a distinct advantage. Since the test castings are produced in dry sand molds, and complete sets can be poured rapidly, variables such as fluctuations in mold and pouring temperatures and changes in alloy composition are avoided. The results obtained have been found reproducible, and have displayed little experimental scatter.

Hot Cracking Ratings

Relative hot cracking ratings are based entirely upon whether or not castings are cracked without giving consideration to the length of cracks. This provides a convenient and positive method for assessing resistance to hot cracking. Also the method is realistic because it conforms to foundry inspection standards, which usually specify that acceptable castings must be free from cracks.

The relative hot cracking ratings assigned to commercial alloys are in good agreement with foundry experience. Alloys having the best ratings are used to produce intricate castings. Greater susceptibility to hot cracking has been observed with alloys that are rated inferior by the test. Comparisons between ratings obtained from the "U" casting test and results reported by other investigators are of interest. Figure 7 compares ratings obtained from the "U" casting test with relative hot cracking ratings based upon cooling curve analyses.[1]

The cooling curve method was based upon the premise that metals form a coherent network before they solidify completely. By determining the difference in temperature between the formation of the coherent network and complete solidification, values termed coherence ranges were established for various alloys. Alloys having the smallest coherence range were considered to have maximum resistance to hot cracking.

Figure 5 shows data plotted from published results[2] of tests performed with the Singer and Jennings ring mold,[3] in which the effect of silicon and copper on the cracking susceptibility of aluminum alloys was determined. The curve shows the effect of silicon contents ranging from 0 to 3 per cent in an Al + 4 per cent Cu alloy. Comparison with "U" test ratings, shown in Fig. 5, reveals that the two methods give similar results, but the "U" casting test is capable of evaluating alloys that have greater resistance to hot cracking.

"U" Test Versatility

The versatility of the "U" test permitted the evaluation of magnesium casting alloys. Work performed by R. A. Dodd, W. A. Pollard and J. W. Meier[4] showed that the Singer and Jennings ring mold test was not severe enough for most commercial magnesium alloys. They developed modified ring tests which involved solidifying castings around steel cores or around carbon dioxide hardened sand cores. Table 4 compares ratings obtained from the "U" casting test, and classifications based on hot cracking sensitivity determined with the Dodd-Pollard-Meier test.

The two methods agreed on comparisons between alloys of the same general type. Both rated AM100A slightly superior to AZ92A and noticeably better than AZ63A, EK41A superior to EZ33A and HK31A superior to HZ32A. However, the two methods differed in the overall ratings of resistance to hot cracking. The "U" casting test rated alloys EK41A and EZ33A better than alloys AM100A and AZ92A. The Dodd-Pollard-Meier modified ring test rated alloy AM100A the least susceptible to hot cracking.

Alloys EK41A and AZ92A were rated equal and noticeably better than alloy EZ33A. The modified ring test placed both alloys EZ33A and AZ63A in the same general classification. The magnesium alloys containing thorium exhibited higher ratings with the "U" casting test than with the modified ring casting test.

Foundry experience has indicated that ratings from the "U" casting hot cracking test reliably indicate the hot cracking susceptibilities of alloys. Although significant differences between alloys AM100A and AZ92A are not found in sand casting operations, alloy AM100A is considered to be less susceptible to hot cracking when poured in permanent molds. Alloys EK41A and EZ33A can be used for intricate permanent mold castings.

However, sufficient comparisons, on a production scale and in similar castings, have not been made to substantiate definitely the superiority of the rare earth containing alloys over Mg-Al and Mg-Al-Zn alloys that was indicated by the test.

SUMMARY

The "U" casting test provides a rapid and convenient method for surveying an extensive range of testing conditions. Alloys exhibiting a wide range of hot cracking susceptibilities can be assessed, yet minor changes in resistance to hot cracking resulting from composition variations in specific alloys can be determined. Since variables affecting hot cracking can be closely controlled, the results are reliable and can be reproduced with little experimental scatter.

The test is suitable for assessing both aluminum and magnesium-base alloys. Ratings depicting the relative hot cracking resistance of commercial as well as experimental alloys are in good agreement with foundry experience, and agree more favorably than ratings obtained by other methods.

ACKNOWLEDGMENT

The author acknowledges substantial contributions by R. C. Boehm during the development of the hot cracking test, the advice and guidance received from R. C. Lemon during the preparation of the paper and the assistance of E. F. Fischer and W. C. Newhams in supplying information and test data concerning magnesium casting alloys.

REFERENCES

1. E. E. Stonebrook and W. E. Sicha, "Correlation of Cooling Curve Data With Casting Characteristics of Aluminum Alloys," AFS TRANSACTIONS, *vol. 57*, p. 489-496, 1949.
2. R. H. Jennings, A. R. E. Singer and W. I. Pumphrey, "Hot Shortness of Some High Purity Alloys in the Systems Aluminum-Copper-Silicon and Aluminum-Magnesium-Silicon," *Journal of the Institute of Metals, vol. 74*, p. 227-248, 1948.
3. A. R. E. Singer and P. H. Jennings, "Hot Shortness of the Aluminum-Silicon Alloys of Commercial Purity," *Journal of the Institute of Metals, vol. 73*, p. 197-212, 1946.
4. R. A. Dodd, W. A. Pollard and J. W. Meier, "Hot Tearing of Magnesium Casting Alloys," AFS TRANSACTIONS, *vol. 65*, p. 100-117, 1957.

MARTENSITIC WHITE IRONS FOR ABRASION-RESISTANT CASTINGS

By T. E. Norman, A. Solomon and D. V. Doane

ABSTRACT

Martensitic white iron castings are extensively used to resist abrasive wear in a wide variety of applications. There are numerous variables in composition, production control, heat treatment and final structure which can influence the wear resistance and mechanical properties of such castings. It is important, therefore, in the interests of obtaining optimum service life, to recognize the effects of these variables and take the steps necessary for their proper control.

This paper deals with a number of the important characteristics of the martensitic white irons, together with studies which have been made by the authors on a number of the more important variables influencing their structure, wear resistance, mechanical properties and resultant service life.

Definition of Martensitic White Irons

To be technically accurate, the term "martensitic white iron" should probably be confined to those irons whose structure consists essentially of martensite and iron or alloy carbides. However, practically all such irons contain substantial quantities of retained austenite in their structure, and usually there is more austenite than martensite present. Some of these irons contain only austenite and carbides.

This retained austenite is normally metastable in character; therefore, it can transform to martensite or a martensite-like structure at the wearing face of the castings. This is brought about by the action of the abrasive forces, which cause plastic deformation and resultant transformation of some of the austenite at the wearing surface. Under the circumstances, it appears to be reasonably accurate to classify all such white irons as martensitic.

To be sufficiently inclusive, the authors suggest that martensitic white irons are those irons which contain a graphite-free structure of iron or alloy carbides and austenite or one of the low temperature transformation products of austenite. This definition will apply to all of the martensitic white irons dealt with in this paper.

MARTENSITIC AND PEARLITIC WHITE IRONS COMPARATIVE PROPERTIES

Composition and Structure

All white irons are basically iron-carbon alloys which solidify with a portion of the carbon in the form of primary carbides. In the iron-carbon system, this occurs when the carbon exceeds about 2.0 per cent. Practically all of the alloying elements, when added to this system, progressively lower the concentration of carbon necessary to form the primary carbides.

In white iron produced specifically for abrasion-resistant castings, the carbon normally exceeds 2.5 per cent, and is usually in a range of 3.0 to 3.6 per cent. Silicon contents are normally held under 1.0 per cent to avoid graphitization, and certain other alloying elements such as chromium are frequently added to further suppress formation of graphitic carbon in the structure. The structure of these irons immediately after solidification consists of primary carbides and austenite.

As the iron cools, this austenite transforms to pearlite, or to mixtures of ferrite and pearlite. This provides the basic structure of pearlitic white irons. Figures 1 and 2 show typical microstructures of pearlitic white irons. Figure 1 is an unalloyed iron, while Fig. 2 contains 5 per cent chromium.

Addition Alloys Effect

When certain alloying elements, notably nickel and molybdenum, are added in sufficient concentrations to white iron compositions, the transformation of austenite to pearlite during cooling of the iron after casting is completely suppressed. As a result, the austenite remains stable down to temperatures below which pearlite can be formed. During further cooling, the austenite in these alloyed irons may partially transform to low-temperature bainite and to martensite, although substantial proportions of this austenite normally are retained in the structure even after the casting has reached room temperature.

Such structures form the basis for the martensitic white irons. Typical microstructures of nickel-chromium white irons of this type, both sand and chill cast, are shown in Figs. 3 and 4, respectively.

It will be noted that in Figs. 1 through 4 the matrix (the principal phase in which another constituent is embedded) of the structure is iron carbide. This carbide has the formula Fe_3C or $(Fe,Cr)_3C$, and will be designated type C_1 for the remainder of this paper. Other carbide-forming alloying elements, such as manganese and molybdenum, also partially enter these type C_1 carbides. Type C_1 carbides have a relatively high hardness (800 to 1000 DPH under a 25-gram load), which contributes

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to abrasion resistance. Data from various sources indicate that these carbides are approximately as hard or slightly harder than quartz, which is one of the most common abrasives encountered in commercial service.

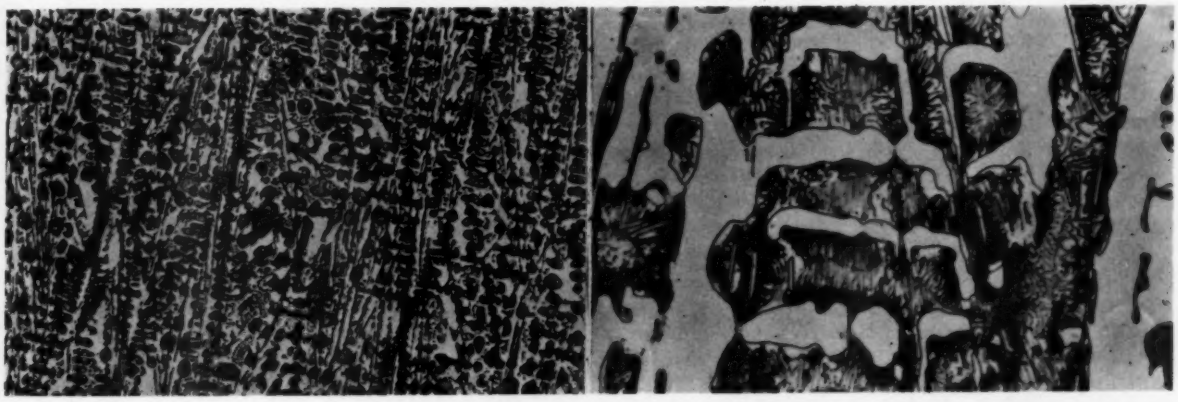
By increasing the chromium content of white iron compositions to concentrations over about 10 per cent, a new type of primary carbide with the formula $(Cr,Fe)_7C_3$ is formed in the structure.[1] This carbide, which will be designated type C_2 for the remainder of this paper, no longer forms the matrix phase, but instead is itself contained in a pearlite or austenite-martensite matrix. Figures 5 through 7 illustrate the typical structures obtained from compositions of this type.

Figure 5 is from a sand-cast test bar with a pearlitic matrix. Figure 6 is from a sand-cast test bar with a matrix which is largely austenite with small amounts of pearlite and martensite. Figure 7 is a similar composition except that it is chill cast. It shows extremely fine-grained primary carbides well surrounded by a matrix of austenite, with possibly some martensite.

By comparing the structures in Figs. 1 through 4 with Figs. 5 through 7, it is noteworthy that the matrix structure of the high chromium irons provides a much more favorable basis for good potential strength and toughness in the irons. In addition, the type C_2 carbides in Figs. 5 through 7 tend to be much finer grained than the type C_1 carbides in Figs. 1 through 4. This factor should further enhance the potential strength and toughness of high-chromium irons. Hardness of the type C_2 carbides is in the range of 1300 to 1800 DPH, which is well above the hardness of quartz. Consequently, high chromium irons should provide outstanding abrasion resistance in those types of service where quartz is the principal abrasive.

Commerical Compositions

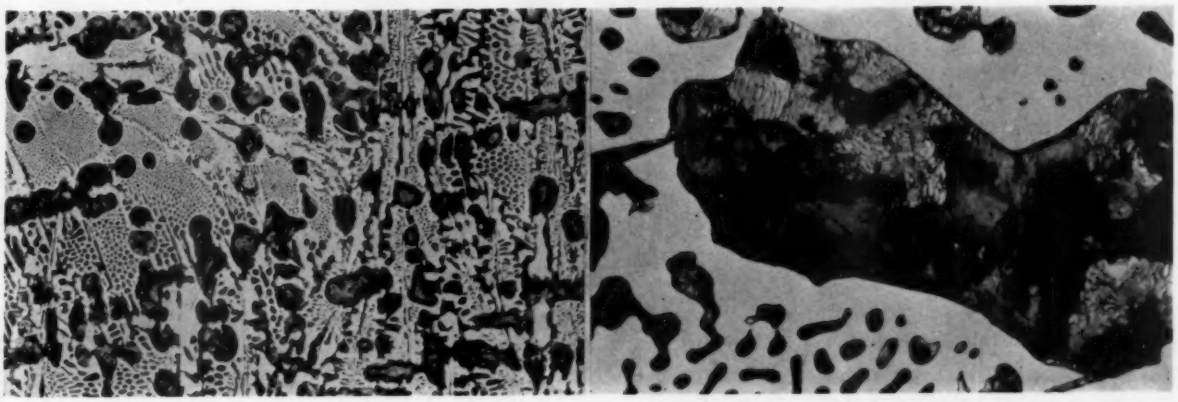
White irons containing type C_2 carbides form the basis of a series of commercially produced compositions containing between about 12 per cent and 30 per cent chromium. When the chromium content is near the lower end of this range, the addition of other alloying elements, such as molybdenum, is normally necessary to produce satisfactory suppression of pearlite in the matrix of these irons. When the chromium content is in a range of about 25 to



4 per cent picral etch. 100 ×.

4 per cent picral etch. 1000 ×.

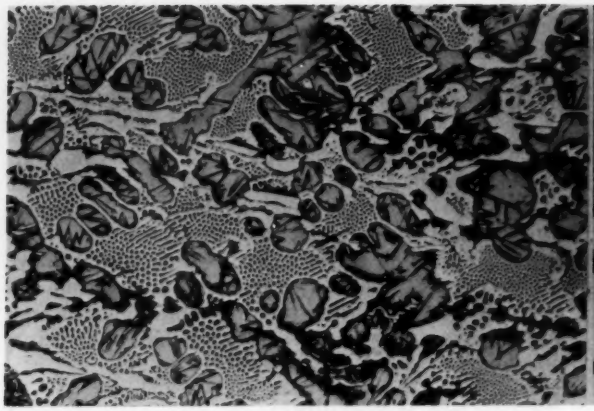
Fig. 1. — Unalloyed white iron containing 3.6 per cent carbon, 0.7 per cent silicon, 0.8 per cent manganese. Structure shows coarse lamellar pearlite and ferrite in a type C_1 carbide matrix. This iron was chill cast to prevent graphite from forming in the structure.



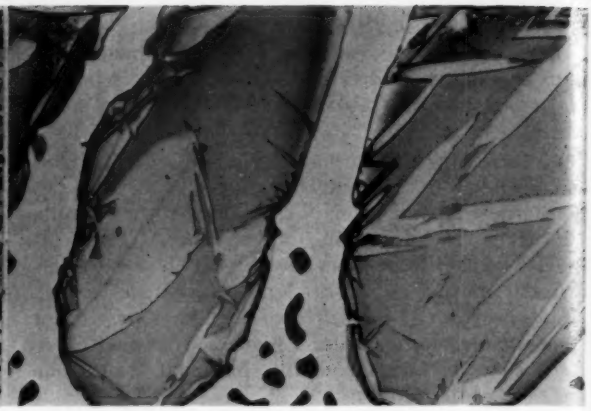
Picral and hot aqueous picric etch. 100 ×.

Picral and hot aqueous picric etch. 1000 ×.

Fig 2. — White iron containing 3.6 per cent carbon, 0.6 per cent silicon, 0.6 per cent manganese, 5.0 per cent chromium. Structure shows fine and coarse pearlite in a type C_1 carbide matrix, as developed near the center of a 2 in. diameter sand-cast ball, air cooled.



1 per cent picric, 5 per cent HCl in methanol etch. 100 ×.



Picral and hot aqueous picric etch. 1000 ×.

Fig. 3 — White iron containing 3.4 per cent carbon, 0.4 per cent silicon, 0.6 per cent manganese, 1.4 per cent chromium, 3.0 per cent nickel. Structure shows austenite-martensite in a type C_1 carbide matrix, as developed near the center of a 2 in. diameter sand-cast ball, air cooled.

30 per cent, and the carbon content is held below about 2.7 per cent, the use of other alloying elements to suppress pearlite may be unnecessary, except possibly in heavy-section castings.

A further important variable in the structure of these 12 to 30 per cent chromium irons must be considered, since it is easily possible to make them hypereutectic with respect to carbon. Hypereutectic low alloy white irons are quite rare and are difficult to produce commercially. When a 12 to 30 per cent chromium iron is hypereutectic, some of the type C_2 carbides solidify from the melt as relatively large needle-shaped crystals which tend to lower the toughness of the iron.

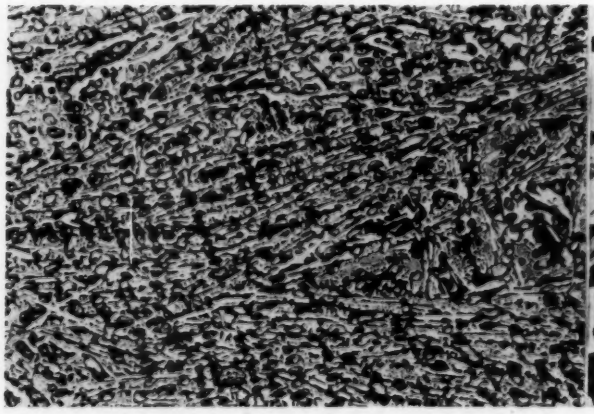
These needles also probably injure the abrasion resistance of the structure by crumbling or spalling microscopically under conditions where high abrasive pressures or some impact is involved. The structures of two such hypereutectic white irons, sand cast and chill cast, respectively, are shown in Figs. 8 and 9. The long carbide needles are particularly dominant in the sand-cast iron.

The eutectic composition range, with respect to carbon for a 12 per cent chromium iron, is between about 3.5 and 3.7 per cent carbon. For a 30 per cent chromium iron it is between about 2.4 and 2.8 per cent carbon. This is indicated by Kinzel and Franks,[2] and is in general agreement with the authors' own metallographic observations on such irons.

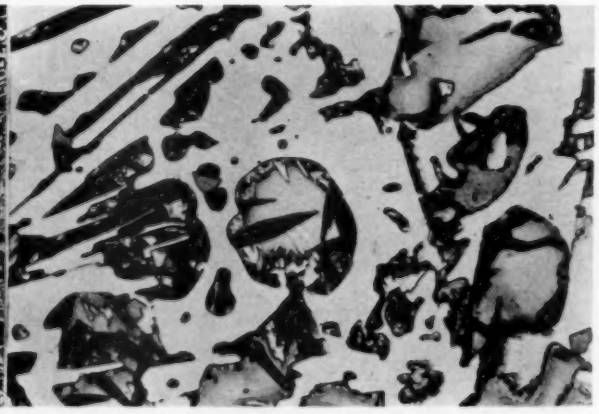
Response to Heat Treatment

It is possible to convert some white iron compositions from a pearlitic type, as-cast, to a martensitic type by reheating to suitable austenitizing temperatures, followed by a quench. Relatively mild heating and quenching rates must be used to avoid cracking or excessively high residual stresses in the casting. Still or moving air, molten salt baths and sometimes oil provide suitable quenching media.

The unalloyed pearlitic white irons have such shallow hardenability that it is practically impossible to convert them to a martensitic white iron, except in very thin sections, by reheating and quenching. With moderate additions of alloying elements such as manganese, nickel, chromium and molybdenum, the hardenability of these pearlitic white irons



1 per cent picric, 5 per cent HCl in methanol etch. 100 ×.

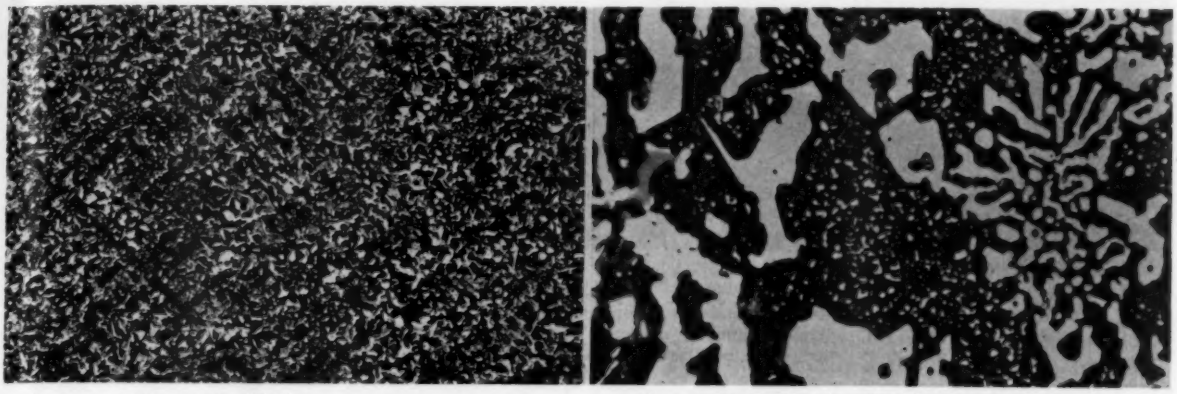


1 per cent picric, 5 per cent HCl in methanol etch. 1000 ×.

Fig. 4 — Chill-cast white iron containing 3.4 per cent carbon, 0.7 per cent silicon, 0.5 per cent manganese, 3.1 per cent chromium, 4.0 per cent nickel. Structure shows austenite-martensite in a type C_1 carbide matrix, as developed ¼-in. below the surface of a 2 in. chill-cast ball.

can be increased sufficiently to produce martensite in their structure when they are reheated and quenched. On abrasion-resistant castings with the matrix of type C_1 carbides, such heat treatments have seldom, if ever, been used commercially.

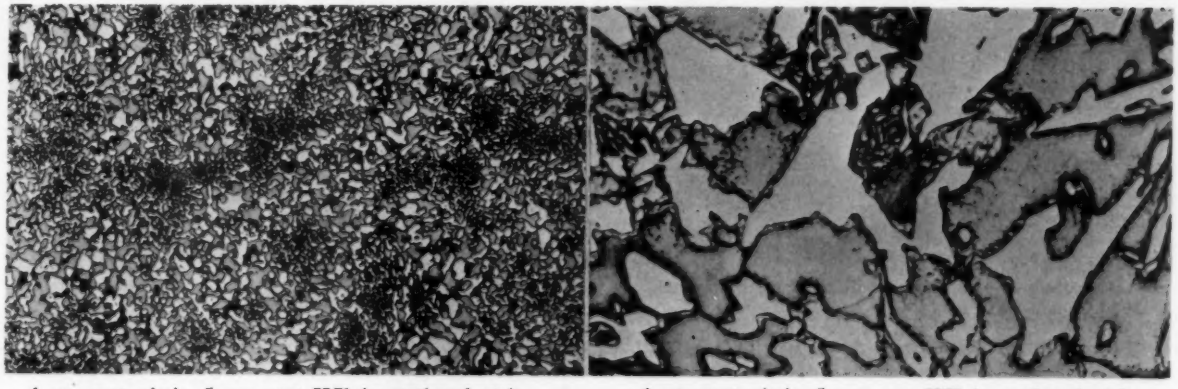
Probably there are a number of good reasons for this, which include the extra cost of such heat treatments and the tendency of such castings to crack or become brittle as a result of heat treatment. Limited investigations by the authors tend to confirm this.



4 per cent picral etch. 100 ×.

4 per cent picral etch. 1000 ×.

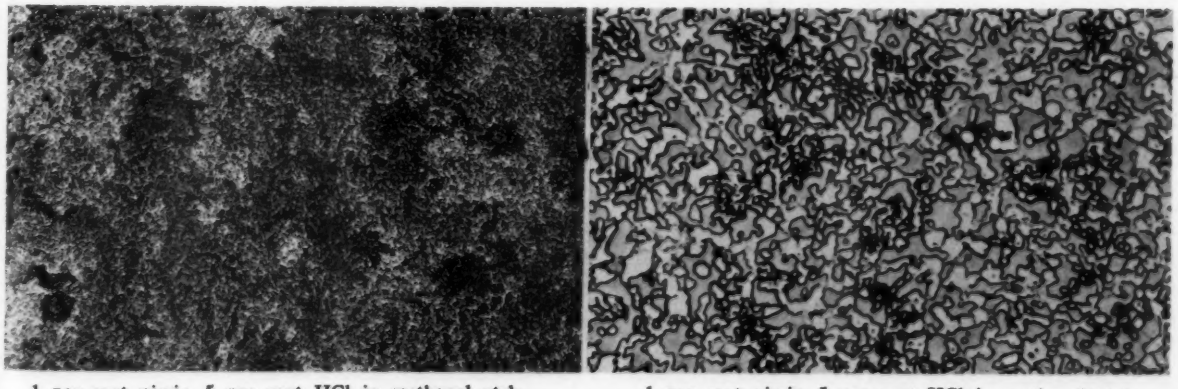
Fig. 5 — Sand-cast white iron containing 3.6 per cent carbon, 0.6 per cent silicon, 0.7 per cent manganese, 12.5 per cent chromium. Structure shows type C_2 carbides in a pearlitic matrix, as developed in a 1.2 in. diameter sand-cast test bar, transferred to a 1700 F holding furnace, then air cooled.



1 per cent picric, 5 per cent HCl in methanol etch. 100 ×.

1 per cent picric, 5 per cent HCl in methanol etch. 1000 ×.

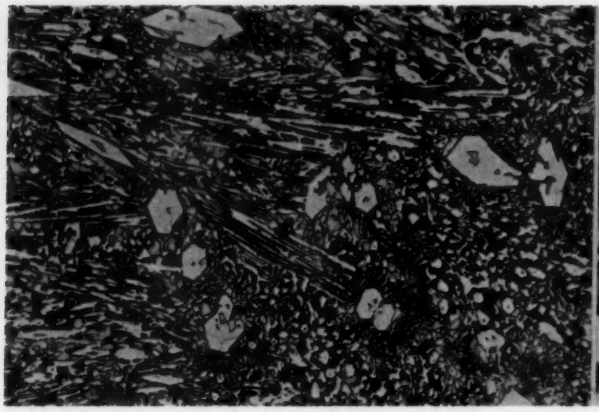
Fig. 6 — Sand-cast white iron containing 3.5 per cent carbon, 0.4 per cent silicon, 0.8 per cent manganese, 16.0 per cent chromium, 3.0 per cent molybdenum. Structure shows type C_2 carbides in matrix which is largely austenite with small amounts of pearlite and martensite, as developed in a 1.2 in. diameter sand-cast bar, air cooled.



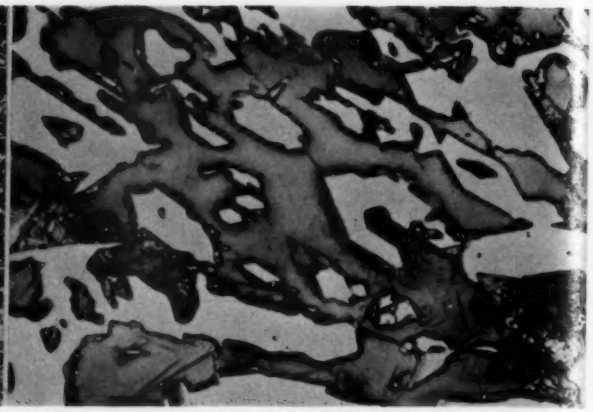
1 per cent picric, 5 per cent HCl in methanol etch. 100 ×.

1 per cent picric, 5 per cent HCl in methanol etch. 1000 ×.

Fig. 7 — Chill-cast white iron from the same composition as Fig. 6. Structure shows type C_2 carbides in an austenite matrix, as developed in a 1.2 in. diameter bar cast in a graphite mold and air cooled.

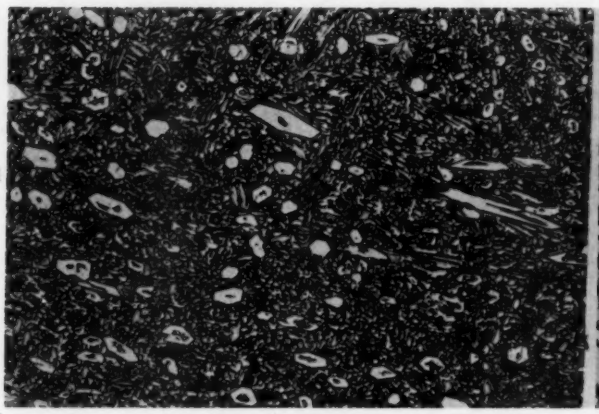


1 per cent picric, 5 per cent HCl in methanol etch. 100 ×.

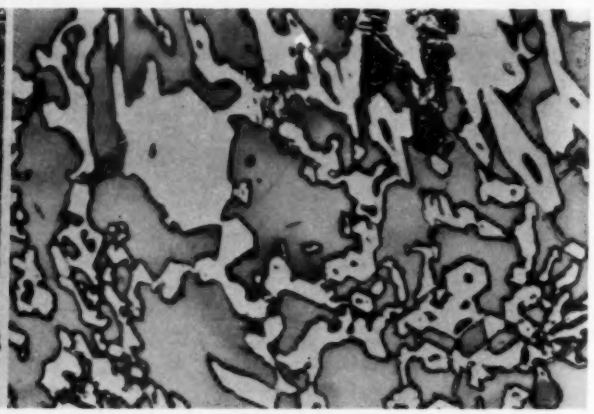


1 per cent picric, 5 per cent HCl in methanol etch. 1000 ×.

Fig. 8 — Sand-cast white iron containing 4.0 per cent carbon, 0.2 per cent silicon, 0.8 per cent manganese, 16.7 per cent chromium, 2.4 per cent molybdenum. Structure shows type C_2 carbides including (at 100 ×) the coarse needles typical of a hypereutectic iron. 1.2 in. diameter test bar.



1 per cent picric, 5 per cent HCl in methanol etch. 100 ×.



1 per cent picric, 5 per cent HCl in methanol etch. 1000 ×.

Fig. 9 — Chill-cast white iron containing 4.0 per cent carbon, 0.2 per cent silicon, 0.8 per cent manganese, 16.7 per cent chromium, 2.4 per cent molybdenum. Structure shows that chill casting results in much finer type C_2 carbides even in a hypereutectic iron. 1.2 in. diameter test bar.

The 12 to 30 per cent chromium white irons normally respond favorably to a reheat and quench type of heat treatment. Probably one of the principal reasons for this is that their type C_2 carbides are contained and well supported in a high carbon alloyed steel-type of matrix. Usually these heat treatments are applied to heavy section castings, which have such slow rates of cooling after casting that they tend to develop a pearlitic matrix in their as-cast condition.

Reheating and quenching (usually by an air quench) will convert this matrix to martensite, provided a sufficient concentration of alloying elements is present in this matrix to suppress pearlite formation. The effects of specific alloying elements on pearlite suppression in white irons will be discussed in later sections of this paper.

Figure 10 illustrates the structure obtained by heating a chute liner made from a 14 per cent chromium, 3 per cent molybdenum white iron to 1900-1950 F, followed by an air quench. The matrix in this structure consists of a dispersion of fine, spheroidized carbides in martensite. Relatively little retained austenite is present in this matrix. Hardness of the casting was 65 Rockwell C. In its as-cast condition (cooled in the mold), this chute liner has a matrix consisting of austenite and pearlite, and a hardness of 53 Rockwell C.

The primary type C_2 carbides in the high chromium irons are relatively unaffected by heat treatment. They appear to be extremely stable and retain their as-cast shape, quantity and high hardness after heat treatment.

Relative Abrasion Resistance

The principal reason for using the alloyed martensitic types of white iron in preference to unalloyed or low chromium pearlitic white irons lies in the fact that the martensitic white irons normally provide substantially better abrasion resistance. The degree of improvement varies, of course, with the application and type of abrasion.

In evaluating the relative abrasion resistance of any material it is advisable to classify the type of abrasion involved, since this usually has a pronounced effect

on the spread in wear rates obtained between a superior and inferior material. Avery[3] has classified abrasive wear into the following three types, which have provided a useful basis for the practical selection of suitable abrasion-resistant alloys.

1) Gouging abrasion, usually with impact.
2) High stress or grinding abrasion.
3) Low stress scratching abrasion or erosion.

Gouging Abrasion. White iron castings are not widely used in gouging abrasion because of the associated high impact or high structural stresses usually involved. There are, however, a number of uses which have provided comparisons between the pearlitic and martensitic types in Table 1. This table lists applications in which abrasive wear is believed to be largely of the gouging type.

The data in Table 1 indicate that the martensitic nickel-chromium white irons with type C_1 carbides are from two to four times as wear resistant as the unalloyed and low chromium pearlitic white irons when compared under conditions of gouging abrasion.

There is little available information as yet on the relative wearing properties of the high chromium martensitic irons and the unalloyed or low chromium pearlitic irons in gouging type of service. The high chromium irons have been extensively used as welded hard facings in this service, and they appear to have interesting possibilities as castings. Their potential strength and toughness should provide worthwhile advantages, which might allow the use of these irons under conditions where the low and medium alloy irons tend to fail by breakage or spalling.

TABLE 1 — RELATIVE LIFE OF PEARLITIC AND NICKEL-CHROMIUM (TYPE C_1 CARBIDE) MARTENSITIC WHITE IRONS IN GOUGING ABRASION[4]

Item	Application	Relative Service Life	
		Martensitic	Pearlitic
1	Crusher rolls for lead-zinc ore	3.8	1
2	Stamp shoes for copper ore	2.2	1
3	Grizzly disks for sizing coke	2.0-3.0	1
4	Hammers for pulverizing coal	2.5	1
5	Chute liners handling coarse ore	2.0-4.0	1
6	Cheek plates in a roll crusher	3.3	1
7	Scoop lips for ball mill feeder	4.0	1

TABLE 2 — LOW CHROMIUM PEARLITIC WHITE IRON AND HIGH CHROMIUM MARTENSITIC WHITE IRON COMPARISON ON A CHUTE LINER HANDLING MINUS 3 IN. SILICEOUS ORE

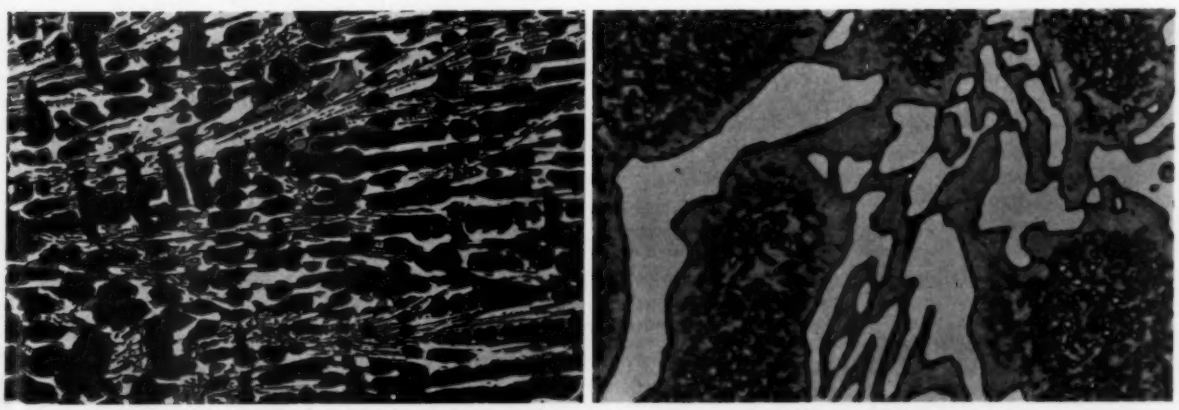
	Item and Description	
	1-Martensitic High Cr Iron	2-Pearlitic White Iron
Composition,%		
C	2.6	3.4
Si	1.5	0.5
Mn	1.1	0.5
Cr	14.3	1.0
Mo	3.0	—
Wear Rate, lb/1000 tons	0.22	0.43

The one known comparison between a martensitic high chromium iron and pearlitic white iron in gouging type of service is given in Table 2. Here a 14 per cent chromium iron had about twice the wear resistance of the pearlitic white iron. This performance could probably be further improved by increasing the carbon content of the 14 per cent chromium iron.

High Stress (Grinding) Abrasion. The use of both the pearlitic and martensitic white irons had been quite extensive in applications involving high stress abrasion. This type of abrasion is most commonly encountered in grinding of ores and industrial minerals. Principal wearing parts are grinding balls and rods, grinding mill liners and grinding rolls.

Table 3 gives the comparative wear rates of martensitic white iron balls in a number of ball mill grinding operations. These rates are compared to a relative wear rate of 100 for chill cast, unalloyed pearlitic white iron in the same service. In this table a rate of less than 100 indicates proportionately better abrasion resistance.

Table 3 indicates that hardness of the abrasive has a pronounced influence on relative wear rates. When hard minerals such as quartz are the principal abrasives, the spread in relative wear rates tends to be small, whereas when a softer mineral such as feldspar or hematite is the principal abrasive, the spread in relative wear rates is substantially greater. Possible



Hot aqueous picric etch. 100 ×. Hot aqueous picric etch. 1000 ×.

Fig. 10 — Structure developed by heat treatment of a high chromium iron. This is from a chute liner 2 in. thick, containing 2.6 per cent carbon, 1.5 per cent silicon, 1.1 per cent manganese, 14.3 per cent chromium, 3.0 per cent molybdenum, reheated to 1900-1950 F and air cooled.

TABLE 3 — RELATIVE WEAR RATES OF MARTENSITIC AND PEARLITIC GRINDING BALLS IN HIGH STRESS (WET GRINDING) ABRASION

Item	Reference*	Casting Method	Service	Principal Abrasives	Wear Rate Relative to Rate of 100 for Chill Cast Pearlitic White Iron
Ni-Cr martensitic irons with type C_1 carbides					
1	5	Chill	Mo ore, 6 ft dia. mill	Quartz and Feldspar	64-68
2	5	Sand	Mo ore, 6 ft dia. mill	Quartz and Feldspar	58-62
3	6	Sand(?)	Raw cement, 7 ft mill	?	71
4	5	Sand	Au ore, 5 ft mill	FeMg Silicates	68
5	7	Chill	Cu ore, 7 ft mill, Plant A	Quartz	68-70
6	7	Sand	Cu ore, 7 ft mill, Plant A	Quartz	65-66
7	8	Chill	Cu ore, 7 ft mill, Plant A	Quartz	65**
8	8	Chill	Cu ore, 8 ft mill, Plant B	Quartz and Feldspar	59**
9	5	Sand	Cu ore, 6.5 ft mill, Plant C	Feldspar and Quartz	45-50
10	7	Chill	Fe ore, 5 ft mill	Hematite	41
11	7	Sand	Fe ore, 5 ft mill	Hematite	39
12	5	Sand	Feldspar, 3 ft mill	Feldspar	29-34
Cr-Mo martensitic irons with type C_2 carbides					
13	7	Chill	Mo ore, 9 ft mill	Quartz and Feldspar	54-59
14	7	Chill	Cu ore, 7 ft mill, Plant A	Quartz	55-57
15	7	Chill	Fe ore, 5 ft mill	Hematite	34

*References listed here are given at the end of the paper.
**These wear rates are based on large-scale consumption tests.

reasons for this are indicated and discussed in a previous paper by one of the authors,[9] and by other investigators.[10,11,12]

Martensitic white iron grinding balls have probably shown their most outstanding superiority over other materials in the dry grinding of cement clinker, where their wear resistance is reported to be from four to seven times that of martensitic high carbon steel balls.[4,6] Data on the relative wear of martensitic and pearlitic white iron balls in this service have not been found, so comparisons in the grinding of cement clinker are not available for inclusion in Table 3.

Some typical relative wear rates of nickel-chromium and nickel-chromium-molybdenum martensitic white irons with type C_1 carbides are given in Table 4. These rates are, in most cases, based on averaged results from several sets of liners of each type, run under closely comparative operating conditions.

The data in Table 4, when compared to the results on balls in Table 3, indicate that somewhat greater improvement in comparative wear resistance is obtained by changing from a pearlitic to a martensitic iron in liners than is obtained in grinding balls. Quite possibly this is due to the effects of heavy section size. Low alloy and unalloyed pearlitic white irons normally contain rather coarse pearlite and considerable ferrite when made in heavy sections. These constituents have relatively poor wear resistance in high-stress abrasion.

Table 4 also shows the wider spread in relative wear rates when the softer minerals are ground, as was indicated in Table 3.

High chromium martensitic white irons with type C_2 carbides offer interesting possibilities under conditions of high-stress abrasion. Use of the high-chromium irons in such applications is relatively new, so little comparative data on their wearing properties are available at this time. Preliminary field reports indicate they have definitely better wear resistance than the martensitic irons with type C_1 carbides. This is further indicated by an evaluation test in the grinding mills at the authors' company. Relative wear rates obtained in this test are given in Table 5.

The procedures used in conducting the wear test for Table 5 are described in detail in a previous paper.[9] They involve the production and testing of large (5-in. diameter) marked balls with structures equivalent to those produced in heavy-section liners of the same compositions. All test balls were run in the same mill at the same time to ensure that each composition was exposed to identical service.

On the basis of the wear rates in Table 5, it appears that the high chromium-molybdenum irons may have interesting economic possibilities for ball-mill liners in some types of service. Their better wear resistance must, of course, be sufficient to justify their normally higher alloy cost, except possibly in those cases where shutdown time for relining a mill represents an important item of cost.

TABLE 4 — WEAR RATES OF MARTENSITIC WHITE IRON BALL MILL LINERS WITH TYPE C_1 CARBIDES IN COMPARISON TO PEARLITIC LOW CHROMIUM OR UNALLOYED WHITE IRON

Item	Reference	Service	Principal Abrasives	Wear Rate Relative to Rate of 100 for Pearlitic White Iron
1	13	End Liners in 9-ft ball mill. Mo ore.	Quartz and Feldspar	50-60
2	8	Shell Liners in 8-ft ball mill. Cu ore, Plant B.	Quartz and Feldspar	45-50
3	8	Shell Liners in 10.5-ft ball mill. Cu ore, Plant D.	Quartz and Feldspar	45-50
4	4	Shell Liners in 7-ft ball mill. Cu ore, Plant E.	Feldspar	30 (approx.)
5	4	Shell Liners in 7-ft ball mill. Wet raw cement.	Limestone(?)	37

TABLE 5 — COMPARISON OF MARTENSITIC WHITE IRON LINER ALLOYS IN GRINDING SILICEOUS ORE

Item	Description and Heat Treatment	Composition, % C	Mn	Si	Cr	Mo	Ni	Carbide Type	Hardness Rc*	Abrasion Factor**
1	Martensitic Cr-Mo white iron, heated to 1950 F, air cooled	2.8	1.0	0.6	15.0	3.0	—	C_2	66	89
2	Martensitic Cr white iron, heated to 2000 F, air cooled	2.7	1.0	0.6	26.0	—	—	C_2	64	98
3	Chill cast Ni-Cr-Mo white iron, cast, cooled in sand, tempered 425 F	3.2	0.7	0.5	2.0	1.0	3.0	C_1	59	107
4	Chill cast Ni-Cr white iron, cast, cooled in sand, tempered 425 F	3.0	0.5	0.4	2.1	—	4.5	C_1	55	116

*Average hardness on the worn surface after the wear test.
**Abrasion factor is the rate of wear relative to a rate of 100 for heat-treated high carbon Cr-Mo martensitic steel containing 1.0% carbon, 0.8% manganese, 6.0% chromium and 1.0% molybdenum. Note that the relative wear rates are on a different basis from those in Tables 3 and 4.

Low Stress Abrasion (Erosion). In erosive types of wear, where impact conditions are usually mild, the various types of white iron normally provide better abrasion resistance than any types of steel or other ferrous alloy. This is believed to be due to the presence of the large amount of carbides, which are normally present in white iron structures. The influence of these hard carbides in resisting erosive wear has been well demonstrated by Haworth[10] and by Avery and his associates.[3,14]

A comparison of relative wear rates of various white irons in erosive service is given in Table 6.

In Table 6 items 1 and 2 were most outstanding, and are the only two materials found to date which can compete economically with low chromium pearlitic white iron in this service. Probably the superiority of these two materials is due to the fact that they contain type C_2 carbides, which are substantially harder than the quartz in the ground-ore slurry. Similar results have been obtained by Haworth[10] and Avery[3,14] on their erosion tests with quartz abrasives, where the martensitic white irons with type C_2 carbides were found to be much superior to either the pearlitic or martensitic white irons with type C_1 carbides.

TABLE 6 — RELATIVE WEAR RATES OF PEARLITIC AND MARTENSITIC WHITE IRONS AND A STEEL FOR CLASSIFIER WEAR SHOES IN EROSIVE WEAR BY GROUND ORE SLURRIES

Item No.	Description	C, %	Hardness, Rc	Relative Wear Rate
1	Martensitic 15% Cr, 3% Mo iron, as-cast	3.3	58	28
2	Martensitic 27% Cr iron, as-cast	3.1	57	48
3	Martensitic 4% Ni, 2% Cr iron, as-cast	3.3	57	80
4	Pearlitic 1% Cr iron, as-cast	3.3	47	100
5	Martensitic Cr-Mo steel, heat treated	1.5	57	180

A comparison of items 3 and 4 in Table 6, which both contained type C_1 carbides, indicates that a small improvement was obtained by changing to a martensitic structure in these irons. The contribution of martensite to abrasion resistance in this wear shoe service appears to be of secondary importance in comparison to the effect of the carbides. This is further indicated by the relatively poor abrasion resistance of the martensitic steel (item 5) which was low in carbide content.

The relatively small contribution of martensite to abrasion resistance of white irons, as indicated in Table 6, is not in general agreement with the results from Haworth's and Avery's tests. For instance, in Haworth's laboratory tests, the 14 per cent chromium irons with a martensitic matrix (obtained by heat treatment) were much superior to the same irons (as-cast) with a pearlitic matrix. Also Avery's data indicate that nickel-chromium martensitic irons with type C_1 carbides had about three times the wear resistance of pearlitic low chromium iron in dry quartz sand erosion.

A further significant comparison is available from the service tests on blades in centrifugal abrading machines, as reported by Haworth.[10] His results for the white irons of approximately equal carbon content are summarized in Table 7.

In Table 7, a martensitic chromium molybdenum iron (item 6) with type C_2 carbides again shows best abrasion resistance, which is in line with the results in Table 6.

A comparison in Table 7 of the martensitic nickel chromium white iron (item 4) and the pearlitic un-

TABLE 7 — WEAR RATES OF PEARLITIC AND MARTENSITIC WHITE IRON AS-CAST BLADES IN A CENTRIFUGAL ABRADING MACHINE USING CHILLED IRON SHOT[10]

Item*	Material	% Composition C	Cr	Mo	Ni	Carbide Type	Hardness Rc	Wear Rate (grams/hr)
6	Martensitic Cr-Mo Iron	3.1	12.1	1.8	—	C_2	52-56	1.7
4	Martensitic Cr-Ni Iron	3.0	1.2	—	4.1	C_1	49-53	2.8
2	Pearlitic Cr Iron	3.2	9.6	—	—	$C_1 + C_2$	51-54	7.4
1	Pearlitic White Iron	2.8	0.2	—	—	C_1	44-48	19.0

*Item numbers are those given in Haworth's original table.

alloyed white iron (item 1), which both contained type C_1 carbides, indicates that in this case the martensitic white iron is much superior to the pearlitic white iron. It is also superior to the pearlitic 9.6 per cent chromium white iron which probably contained a mixture of C_1 and C_2 type carbides.

Since the martensite in white iron is somewhat softer than quartz, its principal contribution to abrasion resistance may be in the support which it provides to the hard carbides in the structure, thus preventing them from crumbling or spalling microscopically under conditions of severe erosive wear. This support may be more necessary under conditions involving high velocity erosion, as in Haworth's and Avery's tests, than it is in low velocity erosion, such as existed on the classifier wear shoes in Table 6.

The results in Tables 6 and 7, together with those given by Haworth,[10] Avery[3,14] and others,[11,15] all indicate that high carbon martensitic white irons with type C_2 carbides provide outstanding resistance to erosive wear. This is further indicated by many service results on these irons in such parts as sand pump impellers, brick mold liners and linings for pipes carrying abrasive fluids. These 12 to 30 per cent chromium white irons are similar in many respects to high carbon modifications of the high carbon, high chromium tool steels, which have also been found to be resistant to erosive wear.

A high vanadium modification of these compositions has also been developed,[16] and is reported to be giving outstanding service in both cast and wrought wearing parts for sand slingers and shot-blasting machines.

The contribution of molybdenum to the abrasion resistance of high chromium irons is significant. The results from items 1 in Tables 5 and 6 and item 6 in Table 7, together with a result reported by Haworth,[17] on brick mold liners, attest to this. While the primary function of molybdenum in these irons is to suppress pearlite formation, it obviously has a further favorable effect on the properties of the carbides as well as the matrix.

Mechanical Properties

Martensitic and pearlitic white iron castings are generally designed for use in applications where high tensile strengths and a substantial degree of toughness are unnecessary. Mechanical properties, other than hardness, are seldom specified for these castings. As a consequence, a common basis for evaluation of these mechanical properties has not yet been adopted. These castings are, however, used in certain applications where a limited amount of breakage in handling, or premature failure in service, is tolerated in the interest of using a material with high wear resistance.

There are other applications where white irons are not now specified, but would be used if a greater degree of toughness could be assured. It should be well worthwhile, therefore, to make greater use of mechanical property tests on white irons. These could lead to the development of improved casting techniques, compositions and heat treatments designed to provide greater strength and toughness in the castings.

Hardness. Pearlitic white irons free from graphite will normally have a hardness range from about 32 Brinell (35 Rc) for low carbon (2.40 to 2.60 per cent) unalloyed compositions, to as high as 600 Brinell (57 Rc) for high carbon (3.50 to 4.00 per cent) moderately alloyed compositions. The martensitic white irons, because of their variable retained austenite contents, show an even wider range in hardness, from a low of about 300 Brinell (32 Rc) for low carbon fully austenitic compositions, to a high of about 70 Rc (beyond the upper limit of the Brinell test) for high carbon, high chromium irons heat treated to develop a high martensite, low austenite matrix.

Although, in general, there is little correlation between the hardness of these irons and their wear resistance, this correlation is not entirely lacking for a given type composition and heat treatment and a specific condition of service. However, the effects of composition, solidification rates, microstructure and heat treatment normally have shown much more influence on wear resistance than the actual hardness of the iron.

Tensile Properties. Due to the difficulties involved in machining and testing white iron tensile specimens, particularly when they are of the martensitic type, there is relatively little good information available from actual tests on tensile test bars. An exception to this is to be found in the data published by Flinn and Chapin.[18] Their observations, on a series of carefully run tests, from specimens sand cast in Y blocks, indicated the following on a series of iron containing 3.4 per cent carbon:

1) Pearlitic white irons containing 1.5 per cent chromium had tensile strengths of 34,000 to 46,000 psi. Plastic elongation prior to breakage was less than 0.0003 per cent.
2) Martensitic nickel-chromium white irons with type C_1 carbides had tensile strengths of about 32,000 psi in their untempered condition, and about 50,000 psi when tempered at 500 F. These irons showed a measurable amount of plastic elongation ranging from 0.0080 per cent in their untempered condition to 0.0044 per cent when tempered.
3) Martensitic nickel-chromium white irons containing 0.33 per cent graphitic carbon in their structures had about the same strength, and appreciably lower plastic elongation, than similar irons free from graphite.
4) Martensitic nickel-chromium mottled irons containing 1.57 per cent graphitic carbon had tensile strengths of 43,000 psi (untempered) to 50,000 (tempered) and corresponding plastic elongations of 0.0505 per cent and 0.0213 per cent, respectively.
5) Charpy impact strength of both the pearlitic and graphite-free martensitic irons on 0.707-in. square unnotched test bars was between 4.8 and 7.5 ft-lb. This increased to a range of 10.3 to 11.0 ft-lb in the nickel-chromium irons containing 1.57 per cent graphitic carbon.

Flinn and Chapin's data indicate that the martensitic nickel-chromium irons, when tempered at

500 F, are somewhat stronger and have greater ductility than the low chromium pearlitic irons. The ability of the martensitic irons to yield plastically, even by small amounts, is probably quite significant in any evaluation of their relative toughness.

Although appreciable quantities of graphite in the structure of nickel-chromium martensitic irons have a favorable effect on toughness, the presence of this graphite seriously damages the abrasion resistance of these irons. It is generally desirable, therefore, to avoid the formation of graphite in these martensitic irons.

Published data[19] on the comparative tensile properties of a martensitic white iron with type C_2 carbides, and a nickel-chromium martensitic white iron with type C_1 carbides, are given in Table 8.

It is evident from the data in Table 8 that the high chromium iron had substantially superior strength and ductility. This is probably due to the more favorable distribution of carbides in this type of iron.

Tensile Data Summary

To summarize the available tensile data, it is indicated that the pearlitic white irons tend to have the lowest strength and elongation, the martensitic white irons with type C_1 carbides are intermediate in these properties and the martensitic white irons with type C_2 carbides have the highest strength and elongation of these three types of white iron. These observations are further confirmed by the authors' tests on the transverse properties of these irons, and by experience with each of these types of iron in actual service.

Transverse Properties. Standard A.S.T.M. transverse test bars, 1.2 in. in diameter, broken on an 18-in. or 12-in. span, have frequently been used to determine the relative strength and toughness of white iron. The authors' results with tests of this type indicate that a fairly wide range of values may be experienced, even with supposedly similar white iron compositions and structures. In spite of this, definite trends are observable. The ranges in results obtained by the authors on various low chromium pearlitic irons, and a series of 12-16 per cent chromium martensitic irons, are given in Table 9.

The results in Table 9 indicate a general improvement in properties when irons of approximately equivalent carbon content are converted from a pearlitic type with type C_1 carbides to a martensitic type with type C_2 carbides. The table further indicates that one of the most effective ways of increasing the strength and toughness of the 12-16 per cent chromium-molybdenum irons is to increase their solidification rate by chill casting. In this case the chills were graphite molds.

The data in Table 9 further indicate that as the carbon content of the 12 to 16 per cent chromium-molybdenum irons is increased, a drop in strength and toughness occurs. Lowest values are obtained in high-carbon, sand-cast irons.

Mechanical Properties as Indicated by Field Experience. In actual applications, the properties most desired in white iron castings, in addition to abrasion resistance, are freedom from breakage and spalling in service. Freedom from breakage is obtained by providing the castings with adequate strength and toughness, together with relatively low values of residual tensile stress. Freedom from spalling under repeated impact involves more complex and more obscure requirements, and appears to be largely a function of composition, microstructure and design of the casting.

In grinding mill liners, wear shoes, mixing paddles and other relatively flat shapes, the martensitic irons are reported to be definitely tougher and less susceptible to breakage than the pearlitic white irons of roughly equivalent carbon content. Experience with the 12 to 30 per cent chromium irons indicates they are the toughest of the martensitic types.

High Residual Stresses

Castings which tend to develop high residual stresses in their structure due to their shape, present a more complex problem insofar as resistance to breakage or cracking in concerned. Under these circumstances, it is difficult to predict the type of white iron which will be most resistant to cracking during production or use of the castings.

The toughness of grinding balls presents an interesting problem in the selection of a suitable material for service under conditions where appreciable impact occurs. Here the pearlitic white irons have been found to be more resistant to breakage or

TABLE 8 — TENSILE PROPERTIES OF TWO TYPES OF MARTENSITIC WHITE IRON[19]

Nominal Composition	Carbide Type C_1	Carbide Type C_2
Carbon, %	3.4	2.6
Chromium, %	2.0	26.0
Nickel, %	4.0	—
Condition	stress relieved	as-cast
Tensile strength, psi	30/60,000*	85,000
Yield strength, psi	—	80,000
Total elongation (elastic+plastic), %	0.10 to 0.35	0.20 to 0.40
Elastic modulus, psi $\times 10^6$	24 to 26	31.5
Brinell hardness	500 to 700	450 to 650

*The higher strengths were obtained with faster cooling rates.

TABLE 9 — TRANSVERSE STRENGTHS AND RELATIVE TOUGHNESS OF VARIOUS PEARLITIC AND MARTENSITIC WHITE IRONS*

Item	Type of Iron	Transverse Strength, lb	Deflection in.	Relative Toughness**
1	Pearlitic, 1 to 2% Cr, 3.2 to 3.5% C. Sand cast.	1400-1800	0.080-0.092	112-165
2	Martensitic 12-16% Cr, 2.0-4.0% Mo, 2.8-3.4% C. Sand cast.	2235-3015	0.125-0.143	278-432
3	Martensitic 12-16% Cr, 2.5-3.0% Mo, 3.5-4.1% C. Sand cast.	1760-2200	0.079-0.110	139-242
4	Martensitic 12-16% Cr, 1.5-3.0% Mo, 3.2-3.4% C. Chill cast.	4360-5060	0.202-0.256	880-1296
5	Martensitic 12-16% Cr, 2.5-3.0% Mo, 3.5-4.1% C. Chill cast.	2800-3470	0.140-0.148	392-486

*Obtained from 1.2-in. diameter bars on 18-in. span.
**Relative toughness here is expressed as the product of breaking load times deflection.

spalling than nickel-chromium martensitic white irons with type C_1 carbides and a high retained austenite content. Studies made on this problem by the authors and by others indicate that these martensitic iron balls are more resistant to breakage on a single blow impact test, but when these same balls are run in a ball mill the repeated impacts, and resultant work hardening of the entire surface of each ball, causes some of the retained austenite at and near the wearing surface to transform to martensite.

This in turn induces high compressive stresses in this spherical surface, since the tranformed austenite, with its greater specific volume, is restrained from expanding. To balance these high compressive stresses, high tri-axial tensile stresses develop in the interior of the ball, which may eventually cause rupture or severe spalling in service. A similar effect has also been experienced with martensitic steel balls when they contain a high proportion of metastable austenite.[9]

The unfavorable stress effects from metastable austenite in martensitic white iron grinding balls can be reduced by several means. These generally involve a reduction in the amount of retained austenite or in the austenite-stabilizing elements. These methods will be discussed in greater detail in later sections of this paper.

WHITE IRON MARTENSITE STRUCTURE DEVELOPMENT

Pearlite Suppression

Methods of Test. For the fabrication of martensitic iron castings, it is necessary to use balanced alloy compositions capable of suppressing pearlite in the structures of various size sections during continuous cooling after solidification. Of the many methods of determining the effects of various alloy additions on pearlite suppression in white irons, the Time-Temperature-Transformation diagrams, drawn from isothermal transformation data by Rote,[20] appear to most closely simulate actual casting practice, because, in this work, castings after shake-out were continuously air cooled to the transformation temperature. The T-T-T diagram, as proposed by Rote, appears to have been constructed to serve as a guide for the fabrication of large-section castings.

For a guide to the study of alloy additions to white iron for small size castings (those which will cool to 800 F in less than 30 min), a modified end-quench hardenability test was adopted. The end-quench hardenability test has many advantages. This test has been universally accepted for steels for its ease of specimen fabrication, economy of material and the simplicity and reproducibility of testing.

The end-quench test modified for white irons consists of casting a flanged specimen 1.2 in. diameter by 3½-in. long in a sand or chill-type mold. The specimen after shake-out at about 1800 F is transferred to a holding furnace at 1700-1800 F for temperature equalization, prior to end-quenching in a modified Jominy fixture. Although the transfer of the casting to a holding furnace for temperature equalization does not necessarily represent the continuous cooling encountered in actual foundry practice the test procedure is consistent, and the effect on transformation during end-quenching would be expected to be reproducible.

Hardness measurements are obtained 0.100 in. below the cast surface along the length of the specimen. Due to the large amount of carbide phase in a 3.5 per cent carbon white iron, the transformation of austenite to pearlite may be insensitive to hardness measurement. In addition to the influence of carbide on the hardness, the proportion of austenite in the martensite-austenite-carbide aggregate will also affect the hardness. Hardness values obtained in 3.5 per cent carbon martensitic irons ranged from 52 to 70 Rockwell C. A typical pearlitic white iron of the same carbon content had a hardness of about 56 Rockwell C.

To supplement hardness measurements, the distance from the quenched end to the point at which pearlite first appears is measured by metallographic examination of the surface on which hardness impressions were obtained. This location is called the pearlite point. The location of complete transformation to pearlite often occurs at cooling rates lower than obtainable with the end-quench bar.

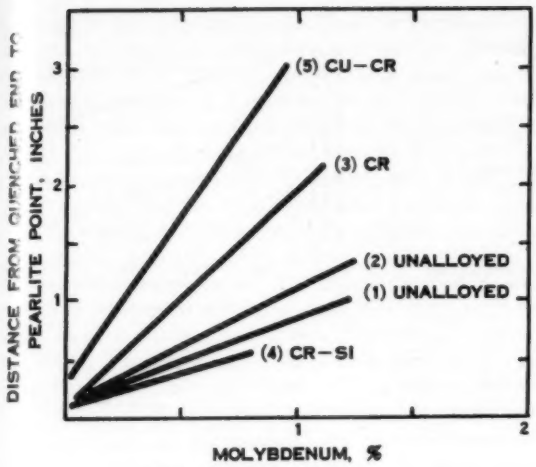
Carbon and Alloying Elements Effect. The effects of various alloy additions on the location of the pearlite point in a number of different investigations of white irons at the authors' laboratory are summarized in Figs. 11-14. In the various base compositions tested, nickel, molybdenum, manganese and copper additions effectively delayed the appearance of pearlite in the structure, as shown in Figs. 11 and 12. Chromium contents up to 13 per cent had relatively little effect on the appearance of pearlite at a constant carbon content, as shown in Fig. 13.

Silicon additions to the base compositions produced a negative effect and accelerated the appearance of pearlite, as shown in Fig. 14. Although the relative effectiveness of various alloying elements in suppressing pearlite is demonstrated in Figs. 11-14, the efficiency of each addition is strongly dependent on the base alloy combination, and to some extent on the carbon content of the austenite resulting from variations in solidification rate.

The effects of carbon and molybdenum additions, and the solidification rate on pearlite suppression of high chromium irons, are shown in Figs. 15-17. The data presented in Fig. 15 indicate that carbon additions affect the appearance of pearlite in a manner similar to the negative effect of silicon additions, shown in Fig. 14. Consideration of Fig. 16 suggests that the effectiveness of molybdenum additions is greater at low carbon levels. The data in Fig. 17 indicate that increases in solidification rate are more effective at low carbon levels.

Carbon and Solidification Rate Effect on Hardness

An example of the influence of carbon and solidification rate on the hardness of high chromium irons is given in Fig. 18. For the compositions shown in Fig. 18, it is quite probable that some correlation exists between hardness and abrasion resistance, since the hardness is primarily related to the quantity and



Base Composition, %

	C	Cr	Si	Cu	
(1)	3.5	—	0.32	—	sand cast
(2)	3.5	—	0.32	—	chill cast
(3)	3.6	5.0	0.50	—	sand cast
(4)	3.7	4.8	1.50	—	sand cast
(5)	3.4	0.9	0.19	2.1	chill cast

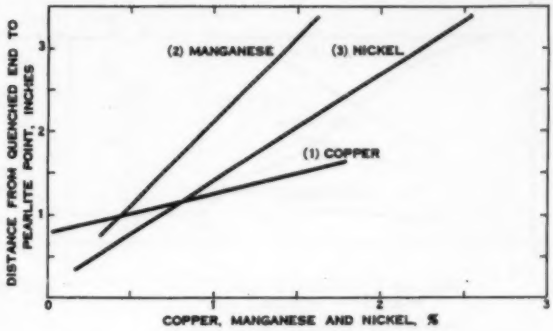
NOTE: 1.2 in. dia. specimens held at 1700-1800 F after casting prior to end quenching.

Fig. 11 — Molybdenum effect on hardenability (pearlite suppression) of low alloy white irons.

distribution of the carbide. However, it is apparent from the graph that an increase in hardness with increasing carbon content is nonlinear. The higher hardness values are related to an increase in the amount of martensite in the matrix, as well as an increase in the amount of carbide phase present.

It is apparent that hardness is also affected significantly by solidification rate. It is possible to achieve the same hardness at a lower carbon content by increasing solidification rate in these high chromium irons.

It has been observed by others that hardness of the nickel-chromium type martensitic white irons with

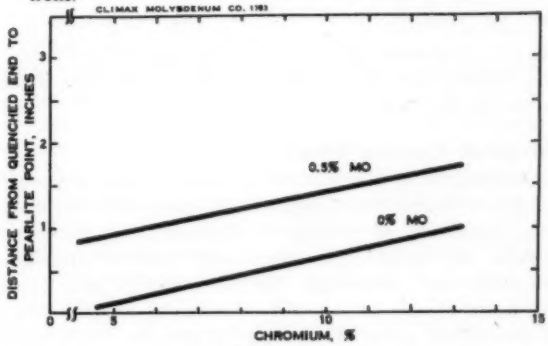


Base Composition, %

	C	Mn	Cr	Cu	Mo
(1)	3.5	0.6	—	—	1.0
(2)	3.4	0.6	—	1.5	1.0
(3)	3.4	0.7	1.5	—	—

NOTE: 1.2 in. dia. specimens held at 1700-1800 F after sand casting prior to end quenching.

Fig. 12 — Copper, manganese and nickel effect on hardenability (pearlite suppression) of low alloy white irons.

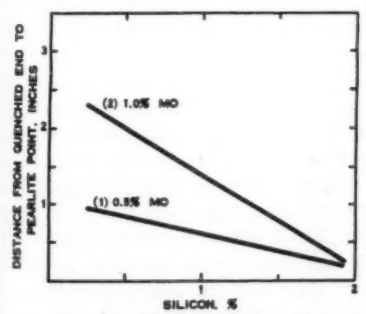


Base Composition, %

C	Mn	Mo
3.6	0.7	—
3.6	0.7	0.5

NOTE: 1.2 in. dia. specimens held at 1700-1800 F after sand casting prior to end quenching.

Fig. 13 — Chromium effect on hardenability (pearlite suppression) of alloyed white irons.

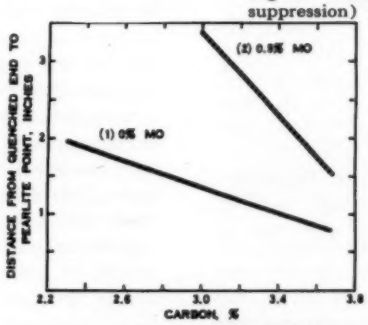


Base Composition, %

	C	Mn	Cr	Mo
(1)	3.6	0.7	5.0	0.5
(2)	3.6	0.7	5.0	1.0

NOTE: 1.2 in. dia. specimens held at 1700-1800 F after sand casting prior to end quenching.

Fig. 14 — Silicon effect on hardenability (pearlite suppression) of low alloy white irons.

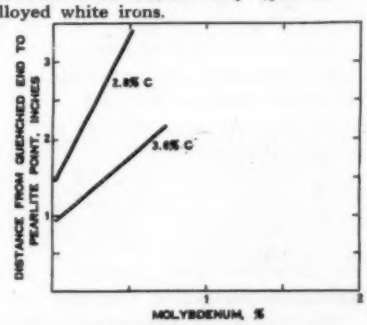


Base Composition, %

	Si	Mn	Cr	Mo
(1)	0.6	0.7	12.5	—
(2)	0.6	0.7	12.5	0.5

NOTE: 1.2 in. dia. specimens held at 1700-1800 F after sand casting prior to end quenching.

Fig. 15 — Carbon effect on hardenability (pearlite suppression) of high chromium white irons.

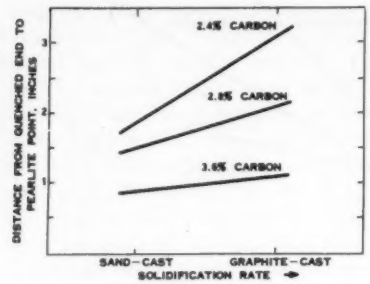


Base Composition, %

Mn	Si	Cr
0.74	0.6	12.4

NOTE: 1.2 in. dia. specimens held at 1700-1800 F after sand casting prior to end quenching.

Fig. 16 — Molybdenum effect on hardenability (pearlite suppression) of high chromium irons.

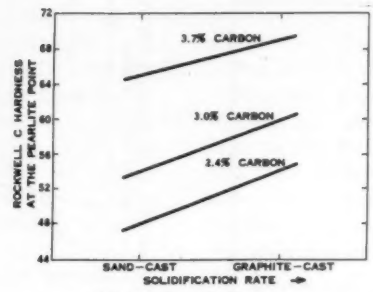


Base Composition, %

Mn	Si	Cr
0.74	0.6	12.4

NOTE: 1.2 in. dia. specimens held at 1700-1800 F after casting prior to end quenching.

Fig. 17 — Solidification rate effect on the hardenability (pearlite suppression) of high chromium irons at three carbon levels.

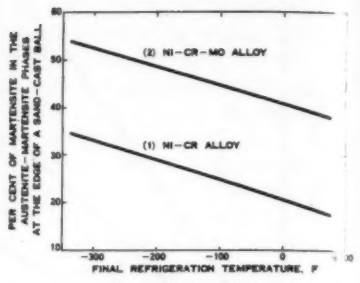


Base Composition, %

Mn	Si	Cr	Mo
0.74	0.6	12.4	0.5

NOTE: 1.2 in. dia. specimens held at 1700-1800 F after casting prior to end quenching.

Fig. 18 — Solidification rate effect on the hardness of high chromium irons at three carbon levels.



Base Composition, %

	C	Si	Mn	Ni	Cr	Mo
(1)	3.4	0.4	0.65	3.0	2.2	—
(2)	3.4	0.4	0.70	0.21	2.0	0.44

NOTE: 2 in. dia. balls were air cooled after sand casting and held at final refrigeration temperature for 30 min. Martensite determined by x-ray diffraction.

Fig. 19 — Final refrigeration temperature effect on amount of martensite in two low alloy white irons.

type C_1 carbides also generally increases with increasing carbon content, which is similar to the effect noted in Fig. 18.

Arrested Cooling Effect

Reduction of the carbon content of the austenite phase by arrested cooling, following solidification, offers a method of increasing the pearlite-suppressing power of nickel-chromium and nickel-chromium-molybdenum irons with type C_1 carbides. The work by Rote clearly shows the advantages of this treatment. This arrested cooling treatment shifted the pearlite transformation of the T-T-T diagram to the right, delaying the appearance of pearlite in the structure.

This may be accomplished in commercial practice by transferring the castings, after solidification, to a preheated furnace at a temperature of about 1700 F. Following this, the castings are cooled to room temperature. A somewhat similar effect can be achieved by burying the castings together after solidification.

Arrested cooling is quite effective in reducing retained austenite, and in suppressing pearlite in high chromium irons containing type C_2 carbides. In these irons, the arrested cooling tends to produce a dispersion of spheroidized carbides in the austenite, lowering the carbon content of the austenite. On subsequent cooling of the lower carbon austenite, pearlite transformation is delayed even in the presence of nucleating carbides.

Retained Austenite Control

Practically all martensitic white irons contain significant amounts of retained austenite, which has an influence both on abrasion resistance and mechanical properties. Its influence upon these properties is rather complex, and in some cases obscure.

Insofar as its effects on mechanical properties are concerned, retained austenite always tends to lower the hardness of martensitic white irons. At the same time it may increase their toughness as measured in a tensile or transverse test, or by single-blow impact. Where repeated impact occurs over a wearing surface, as in the case of grinding balls, the retained austenite may produce an apparent brittleness, due to buildup of residual stresses under this wearing surface. Under these circumstances it is desirable to reduce the amount of retained austenite.

When martensitic white iron castings must be cleaned or finished by grinding before service, the presence of high retained austenite tends to produce hair-line cracks or heat checks on the finished surfaces.

Retained austenite in martensitic white irons can be reduced by various means. One method is to refrigerate the castings at subzero temperatures. This should be followed by a stress-relieving treatment at about 400 F. Figure 19 shows the effect of refrigeration on the martensite content of a nickel-chromium and a nickel-chromium-molybdenum iron. A second method involves tempering at 500 to 550 F, which transforms some of the retained austenite, probably to low-temperature bainite. A third method is to arrest cooling after solidification, as described in the previous section.

A fourth method is to use selectively those alloying elements which suppress pearlite, but have little tendency to stabilize austenite or lower its M_s temperature. This fourth method may, if desired, be combined with one or more of the other three.

Retained austenite can also be reduced or controlled by reheating the castings to suitable austenitizing temperatures, followed by mild quenching. This procedure is well adapted to the 12 to 30 per cent chromium irons, since their type C_2 carbides are discontinuous and permit the use of heat treating procedures similar to those used on tool steels. This reheating and quenching procedure also provides a convenient means of producing a martensitic structure in those large or heavy section castings, which normally have a pearlitic structure after slow cooling in their sand molds.

The influence of retained austenite on abrasion re-

sistance, in the absence of spalling or breakage, appears to be largely dependent upon the composition of the austenite. This effect of composition on the abrasion resistance of a number of martensitic white irons and high carbon steels has been studied and reported in a previous paper by one of the authors.[9] These invesitgations have indicated that when austenite is retained by increasing its carbon or chromium content, an improvement in abrasion resistance is produced. On the other hand, when austenite is retained by increasing its manganese or nickel contents, a lowering in abrasion resistance occurs.

This effect for varying nickel contents in martensitic white iron grinding balls is indicated in Fig. 20. It appears desirable, therefore, in the interest of obtaining optimum abrasion resistance, to hold the manganese and nickel contents to the minimum required to suppress pearlite in their structure. This can be accomplished by the use of balanced alloy combinations, which will be discussed in the next section.

Refrigeration of nickel-chromium and nickel-chromium-molybdenum irons to reduce their austenite content has improved their wear resistance when they were tested as grinding balls in high stress abrasion. This is indicated by the nickel-chromium iron in Fig. 20, and has been further confirmed by similar investigations on nickel-chromium-molybdenum irons. Presumably this refrigeration treatment would be helpful in providing added abrasion resistance in other types of abrasive wear.

Tempering treatments at 500 to 550 F have been investigated in a few tests, which indicated that a small loss in abrasion resistance was produced by this treatment on grinding balls. This may be over-balanced by the increased resistance to spalling and breakage which the tempering treatment provides. In general, however, a refrigeration plus stress-relief treatment appears to be preferable to the tempering treatment, both from the standpoint of abrasion resistance and resistance to spalling.

The effects on abrasion resistance which are produced by arrested cooling treatments have not been investigated by the authors.

The reduction of retained austenite in nickel-chromium irons by the use of a lower nickel content, combined with a molybdenum addition for pearlite suppression, provides a means of improving the abrasion resistance of these irons. This is indicated by a comparison of items 3 and 4 in Table 5. A further improvement in the abrasion resistance of these two irons was obtained by refrigeration at minus 70 F.

BALANCED COMPOSITIONS FOR MARTENSITIC WHITE IRONS

On the basis of present knowledge, the formulation of balanced compositions for martensitic white iron castings should have the following major objectives:

1) A sufficient alloy content to suppress the formation of pearlite in a given casting section size and cooling rate.
2) A sufficient quantity of carbide-stabilizing elements to ensure freedom from graphite in the structure.
3) A high carbon content where optimum abrasion resistance is the controlling factor, or a low carbon content where maximum toughness is necessary.
4) A low silicon content to reduce the tendency for graphite formation and/or pearlite formation.
5) Adjustment of alloy combinations for control of retained austenite to suit the conditions under which the castings are produced and used in service.
6) Addition of alloying elements to increase the hardness of the carbide phase.
7) Addition of alloying elements to increase the toughness of carbides by modifying the composition, shape and distribution of the carbide phase.

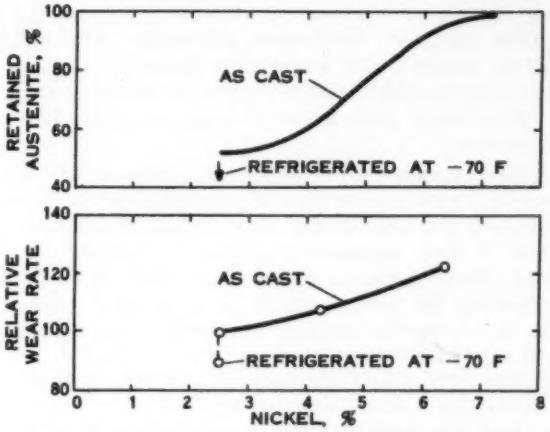


Fig. 20 — Wear rates and retained austenite contents (in martensite-austenite phases) for chill cast nickel-chromium martensitic white iron balls grinding river sand in a 3-ft ball mill. Wear rates are relative to 100 for 0.80 per cent carbon martensitic steel.

All of the foregoing objectives must, of course, be balanced against their cost in relation to the benefits obtained. For instance, a high-chromium molybdenum iron with type C_2 carbides may provide best abrasion resistance in a specific application, but a nickel-chromium-molybdenum iron with type C_1 carbides may provide better economy because of its normally lower first cost.

The foregoing objectives have resulted in the development of two major classifications of martensitic white iron. One depends largely upon the use of nickel, and, more recently, nickel plus molybdenum for pearlite suppression. Chromium, with a low silicon content, is used to suppress graphitization, and also assists in pearlite suppression. The type C_1 carbides are obtained in this type of iron.

The other major classification of martensitic white irons depends on the use of chromium contents in excess of 10 per cent to produce type C_2 carbides in the structure. The type C_2 carbides are harder than the type C_1. They are discontinuous and more favorably dispersed, and tend to have a fine particle size. Complete suppression of graphite in these high chromium irons is obtained due to the chromium content, while pearlite is suppressed by the use of

molybdenum in combination with the chromium. Retention of good mechanical properties with high carbon contents, and consequent optimum abrasion resistance, are obtainable when the chromium content of the irons is held within a range of about 12 to 16 per cent.

The recommendation of specific compositions for each of the two major types of martensitic white irons is beyond the scope of this paper. While the desirable limits of their composition ranges are believed to be fairly closely indicated by the data which have been presented, the most desirable or economical compositions for specific conditions of production and use will vary to some extent. A number of specific compositions for various applications have been indicated in the tables.

SUMMARY AND CONCLUSIONS

1) Two basic types of martensitic white irons are described and defined. One contains a structure of type C_1 carbides (Fe_3C) and austenite, or one of its low temperature transformation products. The other contains type C_2 carbides [$(Cr,Fe)_7C_3$)] and austenite, or one of its low temperature transformation products. Representative microstructures for each type of iron are given and compared to corresponding pearlitic white irons.
2) Both types of martensitic white irons have shown substantially better abrasion resistance than pearlitic white irons in comparative tests. Where comparisons were available, the martensitic white irons with type C_2 carbides were superior to martensitic white irons with type C_1 carbides.
3) The strength and toughness of martensitic white irons are generally superior to those of pearlitic white irons. The martensitic irons with type C_2 carbides showed the best values of strength and toughness.
4) The effects of the commonly used alloying elements, and of carbon and silicon, on the suppression of pearlite in white irons are presented and discussed. Molybdenum, manganese, nickel, copper and chromium were shown to be effective in retarding pearlite formation. Carbon and silicon were shown to accelerate pearlite formation.
5) The effects of solidification rates on mechanical properties and pearlite suppression are presented and discussed. Chill casting was shown to enhance both mechanical properties and pearlite suppression.
6) Heat treatment of martensitic white irons is discussed. Arrested cooling in the range 1400-2000 F is effective in suppressing pearlite. The white irons with type C_2 carbides respond well to reheat treatments, which promote martensite formation.
7) The influence of retained austenite, and methods used for its control, are presented and discussed. Both mechanical and abrasion resistance are influenced by the quantity and composition of the retained austenite.
8) Balanced compositions for the development of optimum properties in martensitic irons are discussed.

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REFERENCES

1. A. B. Kinzel and R. Franks, *The Alloys of Iron and Chromium, vol. 2,* Chapter 8, McGraw-Hill Book Co., 1940.
2. Reference (1), Fig. 62, p. 231.
3. H. S. Avery, *Surface Protection Against Wear and Corrosion,* A.S.M., Chapter 3, 1954.
4. Selected Data from Publications of the International Nickel Co.
5. T. E. Norman and C. M. Loeb, Jr., "Wear Tests on Grinding Balls," T.P. 2319, *Metals Technology* and *A.I.M.E. Transactions, vol. 176,* pp. 490-520, 1948.
6. A. L. Wesner, M. Pobereskin and J. E. Campbell, "A Study of Grinding Ball Wear Employing a Radioactive Tracer Technique," Presented at the annual meeting of the A.I.M.E., Feb. 1958.
7. Data obtained from wear tests on marked groups of balls at various milling operations.
8. Private communications from plants which have conducted large-scale comparative tests.
9. T. E. Norman, "Factors Influencing the Resistance of Steel Castings to High Stress Abrasion," AFS TRANSACTIONS, *vol. 66,* pp. 187-196, 1958.
10. R. D. Haworth, Jr., "The Abrasion Resistance of Metals," *A.S.M. Transactions, vol. 41,* pp. 819-869, 1949.
11. W. A. Stauffer, "Wear of Metals by Sand Erosion," *Metal Progress,* pp. 102-107, Jan. 1956.
12. M. M. Tenenbaum, "Laboratory Evaluation of the Wear Resistance of Steels on Abrasion with Sand Paper," *Vestnik Machinostroyeniya, vol. 36,* No. 8, pp. 24-30, 1956.
13. Data from comparative tests in grinding units at Climax Molybdenum Co.
14. H. S. Avery and H. J. Chapin, "Hard Facing Alloys of the Chromium Carbide Type," American Brake Shoe Co., reprinted from the *Welding Journal,* Oct. 1952.
15. A. H. Dierker and J. O. Everhart, "Wearing Properties of Some Metals in Clay Plant Operation," Ohio State University Engineering Experiment Station Bulletin No. 97, Nov. 1937.
16. D. J. Giles, U. S. Patent 2,575,218, Nov. 13, 1951.
17. R. D. Haworth, "Better Mold Liners for Dry Pressing Brick," *Iron Age,* pp. 79-83, Feb. 1950.
18. R. A. Flinn and H. J. Chapin, "Ductility and Elasticity of White and Gray Irons," AFS TRANSACTIONS, *vol. 54,* pp. 141-153, 1946.
19. Data from publications of the American Brake Shoe Co., (1954).
20. F. B. Rote, G. A. Conger, K. A. DeLonge, "Isothermal Transformation Characteristics on Direct Cooling of Alloyed White Iron," AFS TRANSACTIONS, *vol. 59,* 1951.

ULTRASONIC ATTENUATION IN CAST ALUMINUM

By H. Smolen and H. Rosenthal

ABSTRACT

In the past few years, techniques have been developed for producing high quality aluminum castings. Since radiography does not clearly distinguish the difference between this quality and the ordinarily acceptable grade, there is a need for a nondestructive test method for this purpose. Aluminum test plates (7 per cent Si — 0.3 per cent Mg) were cast with a chill on one edge and a riser at the opposite edge. Near the chill the properties are 38,000 psi ultimate tensile strength, 26,000 psi yield strength and 8 per cent elongation. Near the riser the properties fall off to 33,000 psi ultimate tensile strength, 26,000 psi yield strength and 2 per cent elongation.

Radiographic examination was not sensitive to this change. Ultrasonic attenuation measurements were made at various locations in the plates. There was a definite increase in attenuation in a direction away from the chill edge. Other experiments have been made on samples (5 per cent Si) with various gas contents. It is shown that the attenuation characteristics are markedly affected by the presence of gas and microshrinkage in the samples. Techniques for making ultrasonic attenuation measurements on commercial castings were not considered in this paper.

INTRODUCTION

High-strength, high-ductility aluminum castings have recently become a commercial reality, as shown by the issuance of a standard specification: MIL-C-21180 (Aer), Dec. 5, 1957. The technical developments leading to the adoption of the specification are reviewed by Flemings and Taylor.[1] It is now apparent that these high quality castings will find increasing application in aircraft, missiles and other areas where they will be critically stressed to take advantage of their strength-to-weight ratio.

A necessary consideration in the use of such castings is efficient inspection methods for detecting harmful flaws. Much stress is rightfully placed on the technique of sectioning sample castings to determine test bar properties at selected locations. In addition, 100 per cent x-ray inspection is necessary to eliminate castings with random flaws or excessive porosity.

The purpose of the work described here is to enlarge the scope of inspection which can be non-destructively applied to each casting. Harris and Crowther[2] have recently reviewed the use of ultrasonic testing on aluminum ingots as well as wrought products including plates, extrusions and forgings. Ultrasonic techniques have been used successfully to detect irregularities such as inclusions, porosity and shrinkage holes in large, ferrous castings,[3] in nonferrous cast billets[4] and in arc-cast zirconium and other light alloy ingots.[5]

Although these studies are not directly applicable to the problem of sand-cast aluminum, it is indicated that ultrasonics is potentially a more refined technique than radiography. This is especially important in view of the fact that radiography is of limited usefulness in distinguishing between ordinary levels of soundness, as compared to the soundness characteristic of a premium grade casting.

ULTRASONIC TECHNIQUES

Commercially available ultrasonic test equipment was used throughout the investigation. The equipment is a combined pulse generator and signal receiver. Pulses of electrical energy, generated at test frequencies of 0.5, 1.0, 2.25 or 5 megacycles per sec, are applied to an X-cut piezoelectric quartz crystal transducer and converted into mechanical vibrations.

The mechanical vibrations are coupled into the test piece through a thin layer of couplant (glycerine and aerosol), and continue to travel back and forth through the test piece until the total impinging ultrasonic energy pulse is exhausted. Each return to the transducer is converted to an electrical signal which is detected by the receiver, amplified and displayed on a cathode ray tube screen as a vertical "pip."

Porosity within a test piece tends to cause absorption and scatter of the impinging ultrasonic energy. In setting up an investigation, it is first necessary to find a test frequency capable of penetrating the material. As the frequency is decreased (i.e., the wave length is increased), less absorption and scatter occur and better penetration can be realized.

Initially, all samples of both alloys studied were examined at various test frequencies. It was found that the best definition of porosity content was obtained using a test frequency of 1.0 megacycle per sec on the 5 per cent silicon alloy sample, and of 2.25 megacycles per sec on the 7 per cent silicon—0.3 per cent magnesium alloy.

5 PER CENT SILICON SAMPLES

Castings were made, as shown in Figs. 1 and 2. Pouring temperature was 1288 F (698 C). One casting was poured from a heat degassed for 10 min with

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chlorine (Group I), one from a heat which was untreated (Group II) and one from a heat which was poled 2 min with green wood (Group III). Each casting consisted of five samples which were cast simultaneously from a central hub with different sized gates. Each sample was identified as to pre-pour treatment (Group I, II or III) and gate size (A, B, C, D or E).

The samples in Group I were designed to be least porous, Group II more porous and Group III most porous. In addition, the smaller the gate size the more porous a sample was likely to be. Therefore, sample IE was expected to be the least porous, and sample IIIA the most porous of all the castings studied. The samples were machined just enough to remove the rough skin typical of sand castings in order to provide an adequate surface finish for x-ray. Radiographs were taken of each casting to establish its porosity content. Standard 250 kv x-ray equipment was used.

Figure 3 shows where each specimen was ultrasonically examined at positions designated head, center and base.

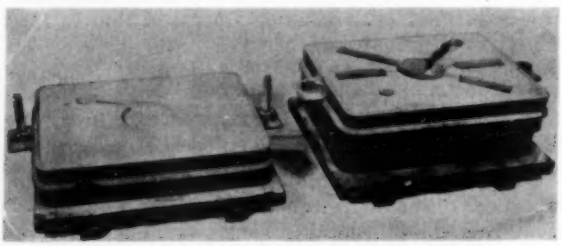


Fig. 1 — Sand mold in which samples were cast.

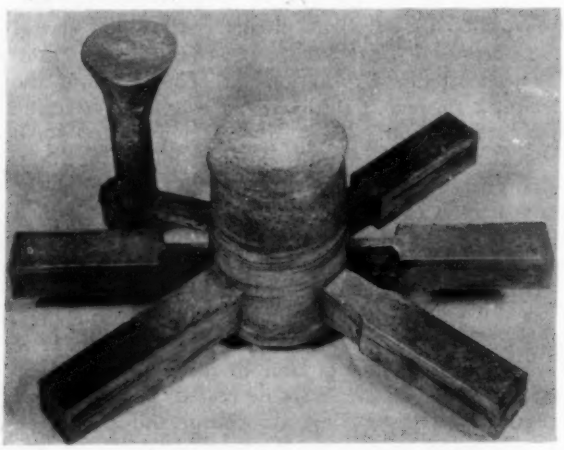
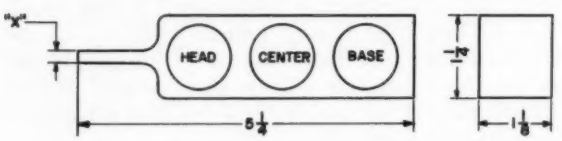


Fig. 2 — Samples as-cast showing riser and gating.



GATE	A	B	C	D	E
X	3/16	5/16	9/16	7/8	1 1/4

Fig. 3 — Line drawing of sample bar showing ultrasonic test positions.

Figure 4 contains the radiograms and reflectograms obtained on all samples tested. In those cases where the same ultrasonic pattern was obtained at the three positions examined on a given test bar, only one reflectogram was taken. Where variations in ultrasonic return on a given test bar were observed, reflectograms were taken of the maximum and minimum return.

The reflectograms showed a consistent change in the amplitude and/or number of back reflections observed. The change is evident in comparing the results obtained on the samples in Group I (least porous), as opposed to those of the samples in Group II (more porous) or Group III (most porous). For a given gate size, ultrasonic attenuation increased as porosity increased, i.e., the ultrasonic return observed on samples in Group I was greater than that observed on samples in Group II, which in turn was greater than that observed on samples in Group III.

Variations in the porosity content within any one sample were readily apparent. The head sections of samples with the smaller gate sizes tended to be more porous than the base sections, as verified by the radiograms. The reflectograms showed that ultrasonic attenuation tended to increase at the head sections.

The radiograms did not show any marked difference in the porosity content of samples in Group II, compared with those in Group III. However, radiograms of the samples in Group I were consistently darker than the others obtained, verifying that these were the least porous studied.

Figure 5a is a reflectogram showing the ultrasonic return observed when a wrought aluminum standard test block (24ST alloy) was examined in a flaw-free area at a test frequency of 1 megacycle per sec. Note that the somewhat bulbous return characteristic of the castings is replaced by sharp "pips," each "pip" representing an echo (back reflection) returned from the bottom surface of the wrought bar.

Figure 5b is a reflectogram taken at the same area of the standard block with the test frequency increased to 2.25 megacycles per sec.

7 PER CENT SILICON — 0.3 PER CENT MAGNESIUM SAMPLES

Four test plates were specially prepared to have a gradient of mechanical properties ranging from premium quality to ordinary sand-cast quality. As shown in Fig. 6, the plates were cast horizontally in green sand. The metal was poured down a 1 in. diameter sprue connected centrally with a runner lying parallel to the 8-in. side of the plate. Two cross runners, attached to the top of the main runner, led the metal directly into the plate. Along this edge of the plate there was a riser running the full 8-in. length, its height was 3 in. and it tapered slightly to a base 2 in. wide.

On the opposite 8-in. side of the plate there were two evaporative water-type chills. These chills consisted of water-filled tubes closed on the bottom. The open end extended above the surface of the sand mold. As illustrated, there were integral lugs on the plate which surrounded the base of these tubes. In practice, the molten metal ran into these lugs, sur-

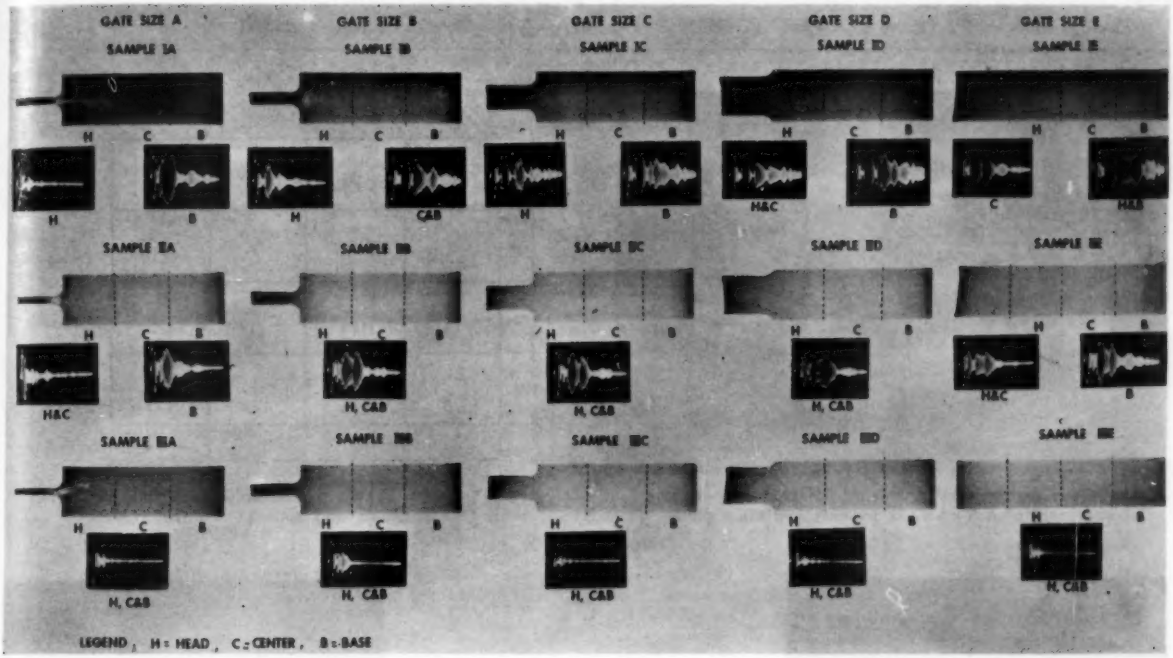


Fig. 4 — Radiograms and reflectograms of samples studied.

rounded the tube and caused the water to boil. So long as water remained available, the temperature in the inner part of the tube could not rise above 212 F (100 C). The result was an efficient chill which occupied a small mold volume.

Two heats of aluminum alloy 356 were induction melted. Two plates were poured from each heat. The metal in plates 1 and 2 (heat 1) was melted, brought to 1350 F (732 C) and poured. Metal in plates 3 and 4 (heat 2) was melted, degassed 10 min at 1300 F (704 C) by chlorine flushing and then poured at 1350 F (732 C). It was anticipated that there would be little or no porosity at the chill edge, and a gradual increase in porosity content as the distance from the chill edge increased.

The plates were machined to a surface finish of approximately 250 micro-in. (rms) and x-rayed. Each was marked off into 16 test positions measuring approximately 1.5 by 2 in. and numbered. Test positions 1 through 4 located at the chill edge were designated Zone A, positions 5 through 8 were designated Zone B, 9 through 12, Zone C and 13 through 16, the area under the riser, Zone D (Fig. 7). Ultrasonic measurements were made of each test position on each plate.

The data are listed in Table 1. The number of back reflections observed at each test position with an amplitude greater than three instrument scale divisions (half-peak) is recorded in column "E" (echoes above three instrument scale divisions). The amplitude of the first back reflection which measured less than or equal to three divisions (half peak) is recorded in column "H" (next echo height).

Reflectograms (Fig. 8) were taken of the ultrasonic return observed at positions 2 (Zone A), 6 (Zone B), 10 (Zone C) and 14 (Zone D) on plate 2 (not degassed) and plate 3 (degassed 10 min by chlorine). The reflectograms clearly show the tendency of ultrasonic attenuation to increase with an increase in distance from the chill edge, and are representative of the patterns observed throughout the the tests.

Plates 2 and 3 were then heat treated to the T-6 condition, and the ultrasonic measurements were repeated. The data obtained are also recorded in Table 1. Plates 1 and 4 were cut into 16 blocks, each block representing an ultrasonic test position (Fig.7), and the density of each was measured by weighing in air

Fig. 5a
A — Initial pulse.
B — First back reflection.
C — Second back reflection.
D — Third back reflection.
E — Fourth back reflection.
F — Fifth back reflection.

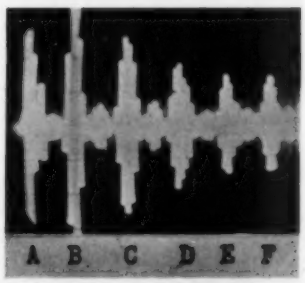


Fig. 5b
A — Initial pulse.
B — First back reflection.
C — Second back reflection.
D — Third back reflection.
E — Fourth back reflection.
F — Fifth back reflection.

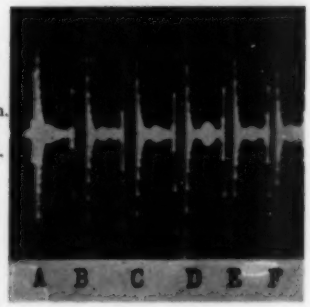


Fig. 5 — Reflectograms taken on flaw-free wrought aluminum bar at test frequencies of 1 megacycle/sec (5a) and 2.25 megacycles/sec (5b).

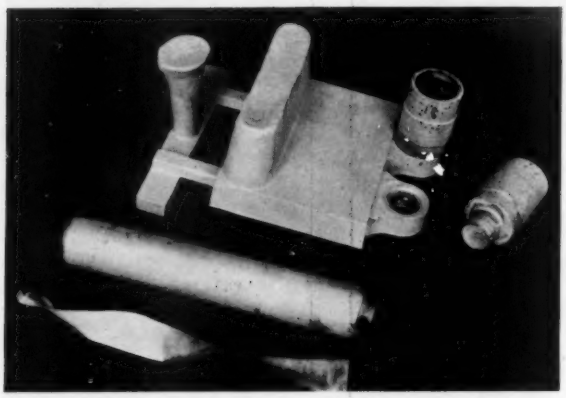


Fig. 6 — Casting showing gating, risering and chilling arrangement of a 8 x 6 x 1¼ in. cast 356 alloy plate.

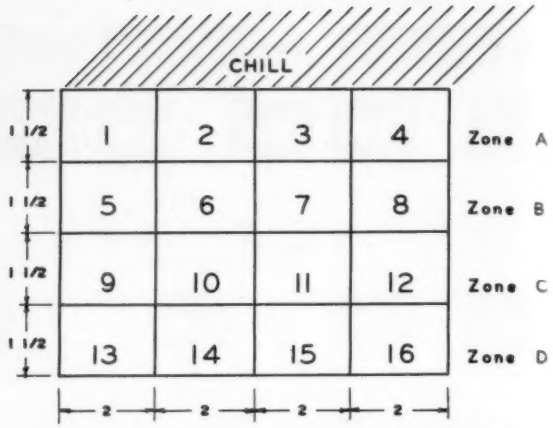


Fig. 7 — Line drawing of sample plate showing ultrasonic test positions and zones.

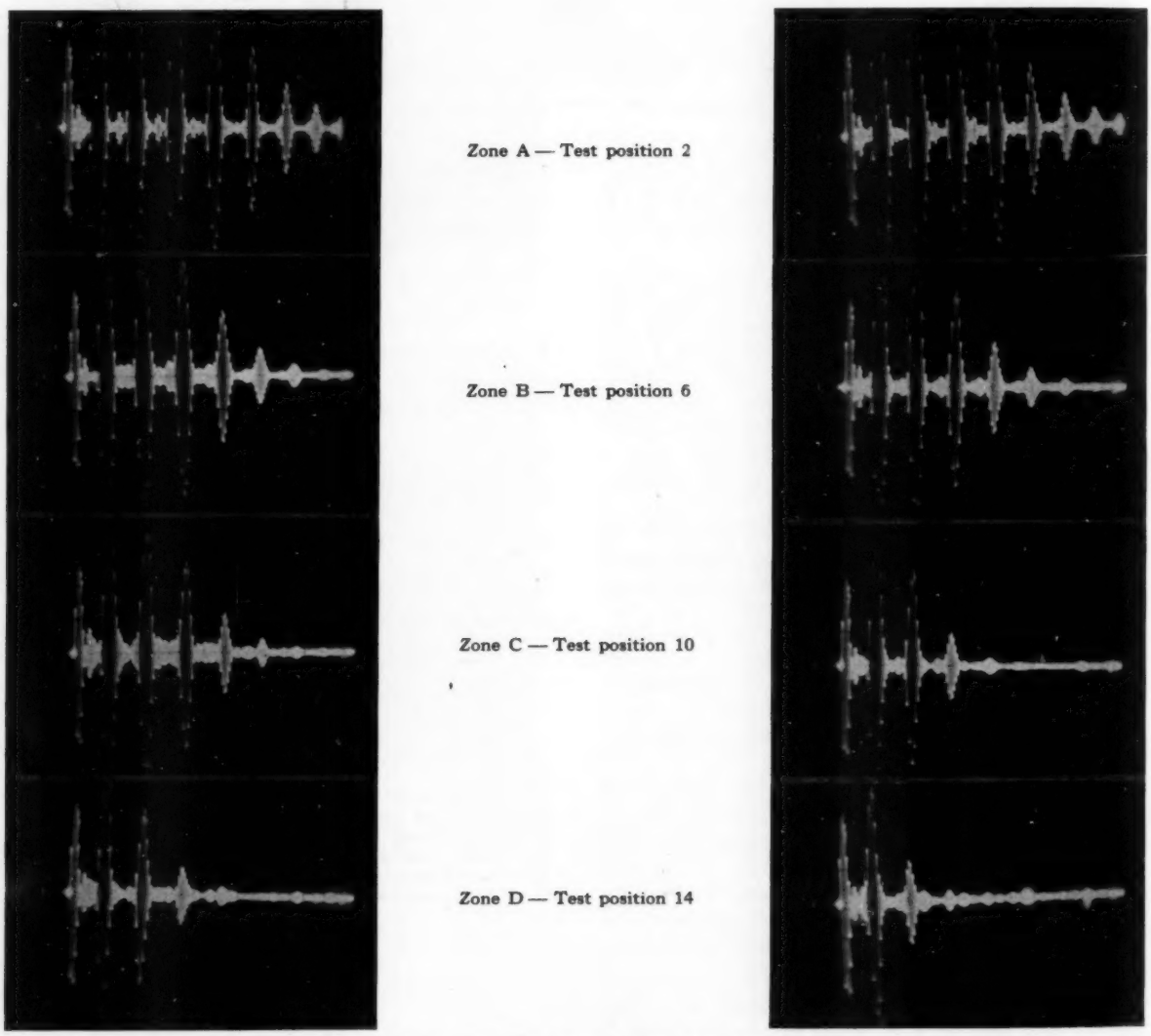


Fig. 8 — Typical ultrasonic test patterns observed. Plate 2, left (not degassed). Plate 3, right (degassed).

TABLE 1 — ULTRASONIC TEST DATA ON CHILLED PLATES (2.25 MCS)

Position Number	Plate 1 (AC)			Plate 2 (AC)			Plate 2 (HT)			Plate 3 (AC)			Plate 3 (HT)			Plate 4 (AC)		
	E	$\bar{X}E$	H	E	$\bar{X}E$	H	E	$\bar{X}E$	H	E	$\bar{X}E$	H	E	$\bar{X}E$	H	E	$\bar{X}E$	H
Zone A																		
1	4		2.3	4		2.0	3		2.9	4		2.0	4		2.4	3		2.0
2	4	4.0	2.7	4	4.0	2.5	4	3.75	2.1	4	4.0	2.0	4	4.0	2.3	4	3.5	2.0
3	4		2.5	4		2.5	4		2.4	4		2.2	4		2.7	4		2.3
4	4		1.6	4		2.1	4		2.0	4		2.2	4		2.7	3		2.8
Zone B																		
5	3		2.2	3		2.8	3		2.7	3		2.2	3		2.7	2		3.0
6	3	3.0	2.3	3	3.0	2.0	3	3.0	1.8	2	2.5	3.0	3	2.75	1.6	2	2.0	2.2
7	3		1.4	3		2.1	3		2.2	2		2.5	2		3.0	2		2.7
8	3		2.1	3		2.6	3		2.2	3		1.7	3		2.2	2		2.8
Zone C																		
9	2		2.8	2		3.0	2		2.9	2		2.0	2		2.3	2		1.3
10	2	2.0	2.1	2	2.0	2.7	2	2.0	2.7	1	1.5	2.8	1	1.75	2.9	2	1.5	1.4
11	2		2.0	2		2.5	2		2.1	1		1.9	2		1.5	1		2.5
12	2		2.1	2		3.0	2		3.0	2		2.3	2		2.8	1		2.9
Zone D																		
13	2		2.5	2		2.9	2		3.0	1		3.0	2		1.8	1		2.5
14	1	1.5	2.7	1	1.5	2.5	1	1.5	2.3	1	1.25	1.1	1	1.5	1.5	1	1.0	1.6
15	1		2.2	1		1.5	1		2.5	1		0.6	1		1.0	1		1.6
16	2		1.6	2		2.5	2		2.8	2		1.6	2		2.1	1		2.7

E = Echoes above 3 instrument scale divisions.
$\bar{X}E$ = Average number of echoes above 3 instrument scale divisions observed at each zone.
H = Next echo height (in divisions).
AC = As cast.
HT = Heat treated.

and alcohol. These measurements are recorded in Table 2.

At the time the plates used in this study were fabricated, two additional plates were cast under the identical conditions used for plates 3 and 4. They were heat treated to the T-6 conditions and cut into 0.505 in. diameter tensile test bars. Ultimate tensile strength, yield strength and per cent elongation were determined. The mechanical property data obtained are graphed in Fig. 9.

DISCUSSION

There were not enough plates tested to draw any firm conclusions on the difference between the chlorine treated and untreated, or the as-cast vs. heat-treated specimens. However, in general, Table 1 indicates that there was little difference between the ultrasonic attenuation characteristic of plates which were chlorine treated and those which were not, or between plates which were heat treated and those which were not.

It is not surprising that chlorine degassing has no apparent effect under the experimental conditions employed. With the use of clean ingot, induction melting and avoidance of overheating, it is well known that gas pickup by the melt is negligible. A degassing treatment is a worthwhile precautionary measure which, in this case, proved unnecessary.

The testing of plates in the as-cast condition, as well as the heat-treated condition, was done to confirm what was theoretically expected. Although the form and distribution of the precipitates in the microstructure are altered by the heat treatment, it is improbable that these fine scale changes would be detectable except at a much higher frequency than that used in these tests.

A study of Table 1 shows that the zone-to-zone variation in the ultrasonic response observed on plates 1 and 2 in the as-cast state was consistent, as was that observed on plates 3 and 4 as-cast, indicating the good reproducibility of ultrasonic measurements for any one pre-cast treatment. It can be seen that the spread of ultrasonic attenuation values for any given zone is quite narrow.

Table 2 shows that the density measurements of plates 1 and 4 were essentially the same, so that a combined average density is also shown for each zone.

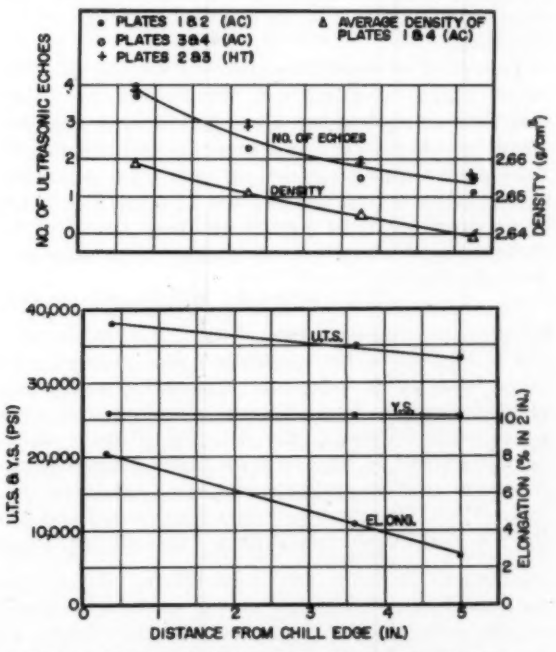


Fig. 9 — Graph showing variation in number of ultrasonic echoes, density and tensile properties with increased distance from chill edge.

TABLE 2 — DENSITY MEASUREMENTS OF PLATE 1 AND PLATE 4 AT EACH TEST POSITION

Position Number	Density (g/cm³) Plate 1	Plate 4	$\bar{x}$
	Zone A		
1	2.65	2.66	
2	2.66	2.66	2.659
3	2.66	2.66	
4	2.66	2.66	
	Zone B		
5	2.65	2.65	
6	2.66	2.65	2.651
7	2.65	2.65	
8	2.65	2.65	
	Zone C		
9	2.64	2.65	
10	2.65	2.65	2.645
11	2.65	2.64	
12	2.64	2.64	
	Zone D		
13	2.64	2.64	
14	2.64	2.64	2.639
15	2.64	2.63	
16	2.64	2.64	

$\bar{x}$ = Combined average density of each zone in plate 1 and plate 4.

The density change amounts to 0.8 per cent over the 5-in. range, i.e., from Zone A (closest to the chill), to Zone D (farthest from the chill). Detection of a 2 per cent density change by radiography is usual by normal techniques, but careful control of radiographic variables is required to detect voids 0.5 to 1 per cent of the thickness.[6] Figure 10, a radiograph of test plate 3, shows the relative insensitivity of the x-ray method used to the detection of small density changes in the chilled plate.

Figure 9 illustrates the relationship between proximity to chill and a number of variables dependent on this distance.

The yield strength is unaffected by the chill distance. This is consistent with the general observation that yield strength in cast alloys is relatively independent of porosity. There is a moderate decline in ultimate tensile strength of 13 per cent from a position 0.5 in. from the chill to a position 5 in. away. As might be expected, the most marked decline is shown by the elongation, which falls to one quarter of its value over the same distance.

Fig. 9 also shows the excellent correlation of ultrasonic attenuation with both density and tensile properties. The ultrasonic attenuation data graphed were taken from Table 1. In order to simplify comparison of plates which were chlorine degassed and those which were untreated, the data for plates 1 and 2 were combined, as were data for plates 3 and 4. Again it is seen that the chlorine degassing treatment had little effect. The combined data for plates 2 and 3 in the T-6 heat-treated condition are also plotted. It can be seen that heat treatment does not affect ultrasonic attenuation at the test frequency used.

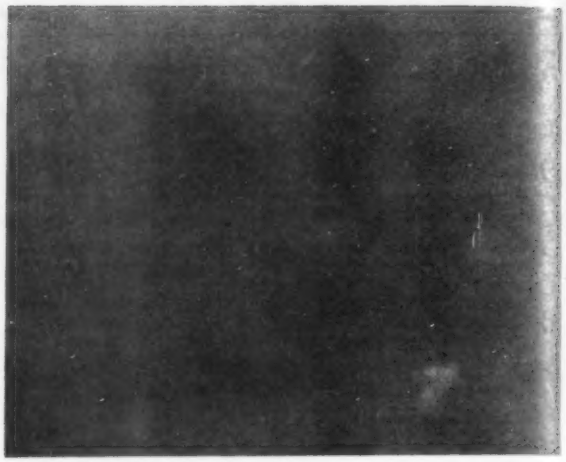


Fig. 10 — Radiogram of plate 3.

CONCLUSIONS

1) Ultrasonic attenuation is sensitive to both small and large changes in the porosity content of aluminum castings.
2) Ultrasonic attenuation shows a clear correlation with the measured mechanical property gradients in a chilled plate casting.
3) Ultrasonic attenuation appears to be more sensitive than radiography in the evaluation of porosity content.
4) In principle, ultrasonic attenuation can be used to differentiate between aluminum castings of premium quality vs. ordinary quality.

ACKNOWLEDGMENTS

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REFERENCES

1. M. C. Flemings and H. F. Taylor, "Adapting Theory to Practice in the Manufacture of Light Metal Castings," 25th International Foundry Congress, Paper No. 9 (1958).
2. J. G. Harris and J. Crowther, "The Use of Ultrasonic Testing for the Control of Quality in the Manufacture of Aluminum Alloys and Other Non-ferrous Materials," *Journal of the Institute of Metals, vol. 86* (5), p. 193-207 (1958).
3. Sperry Products, Inc., Technical Data No. 50-713.
4. S. L. Fry, "Non-Destructive Testing in the Control of Quality of Non-ferrous Castings," *Journal of the Institute of Metals, vol. 85* (8), p. 361-366, 1957.
5. F. W. Wood and J. O. Borg, *Ultrasonic Inspection of Arc-Cast Zirconium and Light Alloys*, U.S. Bureau of Mines Report No. 5126, (1955).
6. K. R. Van Horn, "Radiography of Metals," *Metals Handbook*, American Society for Metals, Cleveland, Ohio, p. 143, 1948.

CARBON FLOTATION IN DUCTILE IRON

By A. H. Rauch, J. B. Peck and G. F. Thomas

ABSTRACT

In hypereutectic ductile iron castings, carbon in excess of that necessary to maintain a carbon equivalent (C.E. = % T.C. + ⅓% Si) of 4.55 per cent, was found to be precipitated during the solidification period. The precipitated carbon is floated to the top surface of the casting, where it is trapped, resulting in an area of high graphite nodule concentration. The flotation of carbon also occurs in the ladle in the time interval between silicon inoculation and pouring of the castings.

INTRODUCTION

As part of the control procedure in the ductile iron operation at the authors' company, pins are cast in a permanent mold from a 1500 lb transfer ladle immediately after inoculation with ferrosilicon at the cupola. The carbon content of these pins was found to be significantly greater in some instances than the amount found in pins poured from a 400 lb pouring ladle on the foundry floor with iron from the same transfer ladle. A further discrepancy was noted between the carbon content of the cast pins and the carbon content of a Y block test bar poured from the same ladle.

The discrepancy between the carbon content obtained by the analysis of chilled pins, and the analysis of wafers cut from test bars, was at first ascribed to the lack of accuracy of the latter procedure. However, subsequent work showed that the analysis of wafers is an acceptable technique for carbon determination, and that the discrepancies between ladle analyses and between test bar and ladle analyses could be explained on the basis of silicon and carbon contents (carbon equivalent) and temperature.

This led to an investigation of the flotation of graphite that is occasionally observed in nodular iron castings. Graphite flotation was noted by Kuniansky* in sand casting where hypereutecticity was pronounced. The graphite flotation phenomenon was differentiated from dross flotation or cope defect.

PROCEDURE

Nodular base iron is melted in a water cooled cupola using basic slag practice, and tapped into a forehearth containing 3000-4500 lb of metal. From the forehearth, 1500 lb of metal is weighed into a treating ladle containing the magnesium alloys, and subsequently poured into a transfer ladle where it is inoculated with 85 per cent ferrosilicon. The iron is then taken to the foundry floor where it is transferred to 400 lb pouring ladles for filling the molds.

Chilled pins, 0.135 in. in diameter, are poured immediately after the silicon inoculation (called cupola pins), and after pouring a 1-in. Y block casting on the foundry floor (called foundry pins). Carbon loss from the time of the silicon inoculation until pouring of the Y block casting was determined by comparing the carbon analysis of the cupola pin with the carbon analysis of the foundry pin. Eighty-three observations were made.

The carbon loss of the metal during solidification was determined by comparison of the carbon analysis of the foundry pins with that of the bottom inch of the Y block. For this portion of the investigation, 128 observations were made.

Carbon analysis of the chilled pins was made on pulverized samples. Carbon analysis of the Y block castings was made on a wafer approximately $\frac{1}{16}$-in. thick, that was further cut into approximately $\frac{1}{8}$-in. squares. This wafer technique was used on all castings examined. Carbon equivalent was determined as per cent carbon plus $\frac{1}{3}$ per cent silicon.

Regression analyses were performed to determine the relationship of carbon loss to carbon equivalent.

To determine the effect of carbon flotation on carbon distribution in the solidified iron, carbon analyses were also made at various locations in the same casting.

METAL TRANSFER CARBON LOSS

A loss in carbon content of the molten iron occurs in the time interval between silicon inoculation and pouring of the castings, whenever the carbon equivalent immediately after inoculation exceeds 4.77 per cent. Figure 1 shows the relationship of carbon loss during this time interval to carbon equivalent of the iron immediately after inoculation.

When the carbon equivalent of the molten iron immediately after inoculation is above 4.77 per cent, kishing, the precipitation and flotation of graphite, occurs on cooling during the time interval necessary for transferring and pouring. Unfortunately, temperatures were not determined and the 4.77 per cent carbon equivalent is an average representation of the pouring temperature only.

*Max Kuniansky, "Problems in Producing Ductile Iron," AFS TRANSACTIONS, *vol. 58*, p. 707 (1950).

A. H. RAUCH is Mgr., Met. Div., and J. B. PECK and G. F. THOMAS are Met., Materials Engrg. Dept., Deere & Company, Moline, Ill.

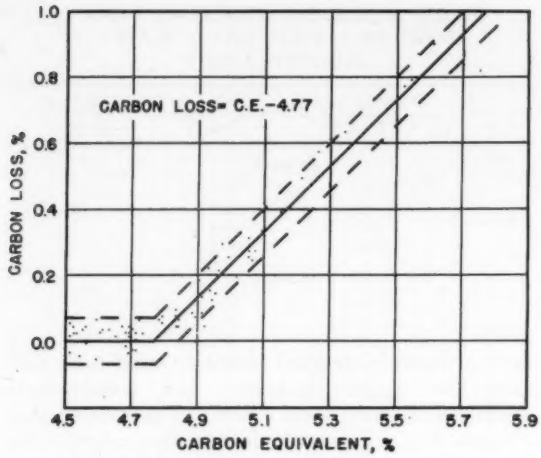


Fig. 1 — Relationship of carbon loss in time interval between inoculation and pouring to carbon equivalent immediately after inoculation with observed data plotted.

CASTING AND SOLIDIFICATION CARBON LOSS

When the carbon equivalent of the molten iron was less than 4.55 per cent, the carbon content of the wafer sample from the bottom inch of the Y block, and the carbon content of the chilled pin, were essentially the same. When the carbon equivalent of the molten iron was above 4.55 per cent, the carbon content of the Y block sample was less than that of the pin. This relationship is shown in Fig. 2.

When the carbon equivalent exceeded 4.55 per cent in the molten iron, flotation or migration of carbon to the cope surface occurred. This was demonstrated by carbon analyses at 1 in. increments from the bottom of the Y block to the top surface, as shown in Table 1.

Carbon flotation manifests itself as a segregation of graphite nodules in a layer at the cope surface in castings of moderate to heavy section size. Figure 3 shows a fractured section of a casting exhibiting carbon flotation, and Fig. 4 is a photomicrograph through the graphite-rich layer. Analyses for the carbon content

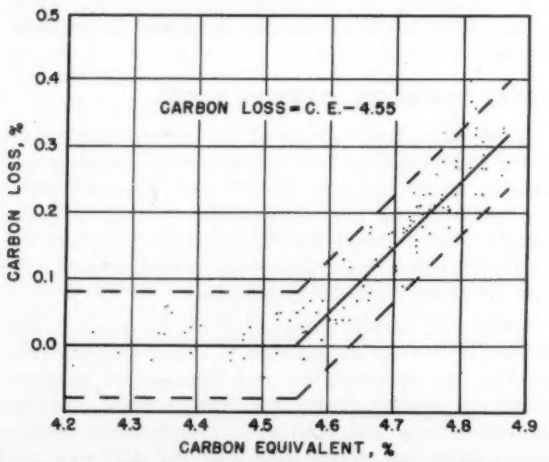


Fig. 2 — Relationship of carbon loss during solidification to carbon equivalent at time of pouring with observed data plotted.

TABLE 1 — CUPOLA PIN, FOUNDRY PIN AND Y-BLOCK CASTINGS CARBON EQUIVALENT COMPARISON

	Sample number							
	1		2		3		4	
Location	T.C., %	C.E., %	T.C., %	C.E., %	T.C., %	C.E., %	T.C., %	C.E., %
Cupola pin	3.76	4.70	3.89	4.80	4.01	4.91	3.41	4.24
Foundry pin	3.78	4.72	3.78	4.69	3.86	4.76	3.48	4.31
Y Block								
Bottom in.	3.59	4.53	3.61	4.52	3.65	4.55	3.39	4.22
2 in. from bottom	3.56		3.60		3.71		3.40	
3 in. from bottom	3.68		3.64		3.55		3.34	
4 in. from bottom	3.63		3.59		3.56		3.37	
5 in. from bottom	4.48		4.85		4.99		3.36	
% Silicon	2.83		2.74		2.70		2.50	

at the cope surface, and below the cope surface, showed that when carbon flotation occurs, the carbon content below the area of flotation yielded a carbon equivalent of 4.55 per cent. Where no flotation occurred, the carbon equivalent was at or below 4.55 per cent throughout the casting. Table 2 shows the analyses of representative castings.

DISCUSSION

The flotation of carbon during the time interval between inoculation and solidification occurs only when the composition is hypereutectic, and can be explained on the basis of the conventional iron-carbon-silicon equilibrium diagram. Figure 5 is a section of the iron-carbon-silicon diagram at 2.0 per cent silicon and 2.0 to 5.0 per cent carbon.* This diagram was originally drawn to illustrate the metastable system. Since the position of the lines are not known with great precision, little inaccuracy will result in adapting this diagram to explain the cooling and solidification of a hypereutectic iron in the stable system.

If the temperature at the time of inoculation is 2500 F, and the carbon content is 4.5 per cent (C.E. = 5.2%), all of the carbon will be in solution, as shown at point A. If, during the time interval necessary to transfer and pour one-half of the ladle, the temperature drops to 2300 F, the composition of the iron will be at point B. The carbon loss will be the horizontal distance between the projection of point A and point B on the abscissa. The equilibrium carbon content at

*Alfred Boyles, *The Structure of Cast Iron*, American Society for Metals, p. 2, (1947). Cleveland.

TABLE 2 — CARBON EQUIVALENT EFFECT ON CARBON FLOTATION OF DUCTILE IRON CASTINGS

	Carbon, %		Carbon Equivalent, %		
Sample	Cope	Drag	Cope	Drag	Silicon, %
1	3.47	3.45	4.47	4.45	2.99
2	3.29	3.29	4.17	4.17	2.65
3	3.47	3.48	4.37	4.38	2.69
4	3.58	3.58	4.49	4.49	2.74
5	4.89	3.51	5.89	4.51	2.99
6	4.39	3.63	5.33	4.57	2.81
7	5.25	3.53	6.22	4.50	2.90
8	8.45	3.53	9.36	4.44	2.74

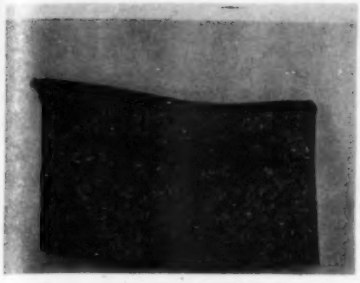


Fig. 3 — Fracture face showing sooty fracture layer resulting from carbon flotation. Dark area at bottom is shear area at fracture origin and not a defect. 1 ×.

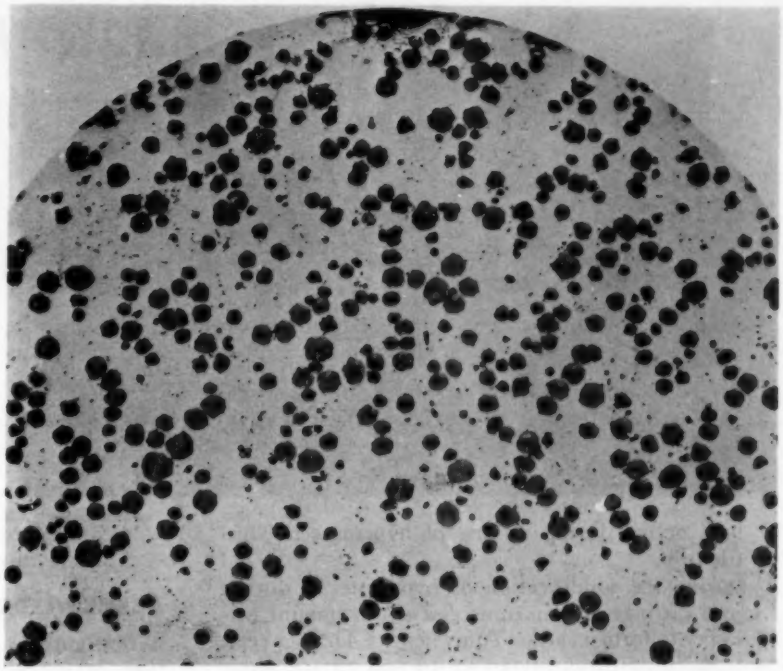


Fig. 4 — Graphite flotation along cope surface of casting (top of photo) shown in Fig. 3. Unetched. 50 ×.

point B is 4.3 per cent (C.E. = 5.0%), therefore the molten iron will have lost 0.2 per cent carbon by kishing. Further carbon precipitation occurs as the temperature of the iron drops to the point where solidification begins, the eutectic temperature shown at point C, where the carbon content is 3.6 per cent (C.E. = 4.3%).

The difference in the carbon content at point B where the mold is poured, and at point C, where solidification begins, is 0.7 per cent. With the density ratio of 3.5 to 1 for iron to graphite, the precipitated graphite rises to the surface of the metal.

The foregoing discussion describes solidification of an iron-carbon-silicon alloy under equilibrium conditions. Ductile iron contains other elements, and does not generally solidify under equilibrium conditions in normal foundry operations. This results in a shift of the phase lines, but should not affect the precipitation and flotation of graphite per se.

The indicated eutectic at a carbon equivalent of 4.55 per cent found in this investigation is probably not the true eutectic point for ductile iron. It may be high due to one of the following conditions:

1) The cooling rate is too rapid for complete carbon precipitation to occur resulting in a supersaturated solution at the eutectic point.
2) As the eutectic starts to form, carbon nodules may be trapped and prevented from floating to the top.

However, the carbon equivalent of 4.55 per cent is a good practical maximum value for moderate section casting if flotation is to be avoided.

This work also indicates that with hypereutectic irons, graphite nodules form in the melt, which is analogous to the formation of kish in hypereutectic flake graphite irons. In the controversy between those who maintain that nodules form in the melt, and those who say nodules form after solidification, frequently no distinction is made between hypereutectic, eutectic and hypoeutectic irons.

Carbon flotation is often confused with another surface defect occasionally found in castings. This condition is commonly called cope defect, and results from a flotation of dross consisting of magnesium sulfides and oxides which have not been cleansed from the metal. Figure 6 is a fractured section through a casting with cope defect, and Fig. 7 is a photomicrograph of the defective area. Dross flotation is dependent upon metal cleanliness, pouring temperature, skimming practice and gating system. Graphite flotation is dependent upon the carbon equivalent in the ladle, pouring temperature and solidification rate.

CONCLUSIONS

1) Kishing, or the precipitation and flotation of car-

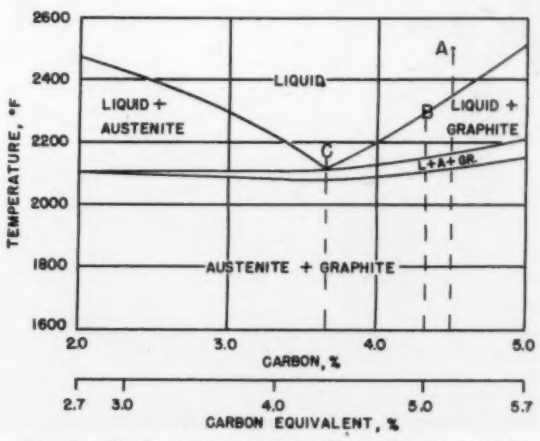


Fig. 5 — Section of iron-carbon-silicon diagram at 2.0 per cent silicon.

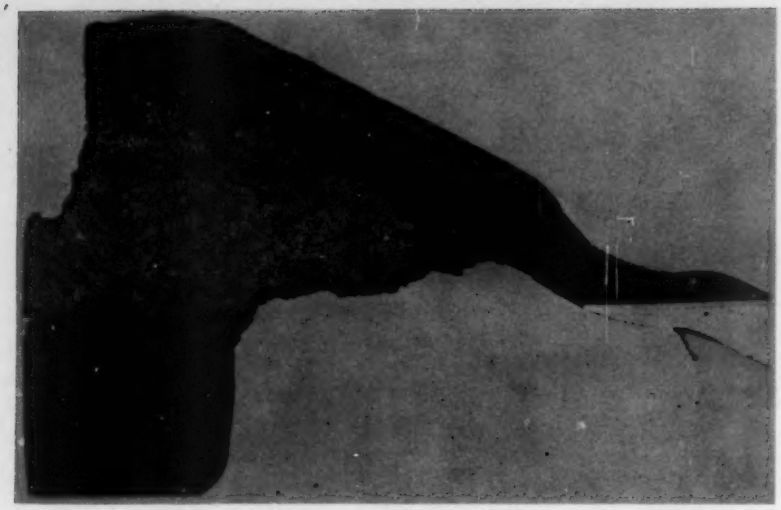


Fig. 6 — Fracture face showing sooty fracture layer resulting from dross and slag flotation. Dark area at bottom of fracture face is shear area at fracture origin and not a defect. 1 ×.

bon, occurs during cooling of hypereutectic ductile iron.

2) During the solidification of hypereutectic ductile iron castings carbon, in excesss of the amount necessary to form a carbon equivalent of 4.55 per cent, is precipitated and floats to the top surface where it is trapped in the solidifying metal.
3) Carbon wafers are a reliable method of sampling ductile iron castings for carbon analysis. The discrepancies in carbon analysis between wafer samples and chilled samples in hypereutectic ductile iron are due to carbon flotation.

ACKNOWLEDGMENT

The authors gratefully acknowledge the cooperation and encouragement of the John Deere Malleable Works, which supplied the castings used in this investigation, and the personnel of the Materials Engrg. Dept., who conducted the chemical analyses and metallurgical examinations.

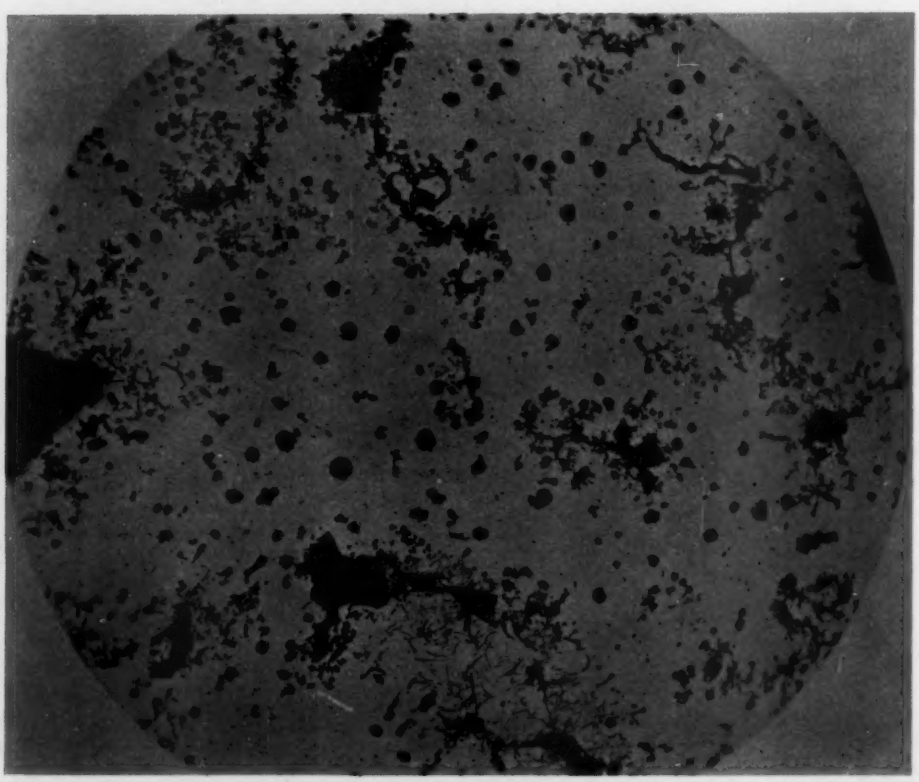


Fig. 7 — Dross and slag flotation in cope defect area in casting shown in Fig. 6. Unetched. 50 ×.

TIMING OF EXPANSION SCAB FORMATION

By J. E. Haller

ABSTRACT

Mold surface penetration was studied to find when expansion scabs occurred. A step casting was designed for making a buckled or expansion scabbed casting. The tests showed that one type of scabs and/or buckles occurred because of sand expansion, and occurred at the time the mold was filling, or while the metal was still in a liquid state.

INTRODUCTION

This investigation was initiated to study mold surface penetration using the theory of exudation of metals during solidification as its basis.* The reference author hypothesizes that the occurrence time of expansion-type scabs is most important, and suggests that the name of this defect be changed to "exudate scab" so it will be differentiated from the errosion-type scabs.

On the basis of this investigation this author disagrees with the reference author. It is stated in the article:

> "From several molds where it is known that the mold cavity has been filled with metal and, subsequently, a run-out has occurred leaving a shell, an expansion scab has never been observed on a casting. This observation leads one to question the commonly-accepted thinking that these scabs form as soon as the mold is filled. Further, many bleeding tests have been conducted and reported in the literature and no mention has been found concerning scabs, penetration or veining being present on such castings."

PROCEDURE

A step casting, as shown in Fig. 1, was designed for these tests to promote a buckled or expansion scabbed casting. The gating used had a basin under the sprue with a single, short ingate into the thickest part of the casting. The gate was $\frac{3}{16}$-in. x $\frac{3}{4}$-in., or 0.14 in.^{2} in area. The actual pouring time for this casting was 18 sec, although normal pouring time would have been 13.8 sec for a gray iron casting weighing 23 lb with an average thickness of $1\frac{1}{4}$ in. A choke ingate was used with runner and ingate in one, as shown in Fig. 2.

Misruns, rattails and expansion scabs and/or buckles may occur from pouring a casting too slowly, according to the author's experience. The theory is that the slower the pour, the more radiant and convection heating can take place on the surfaces of the mold not yet covered by metal. These uncovered surfaces are unlike those immediately covered by metal in that they are somewhat dried.

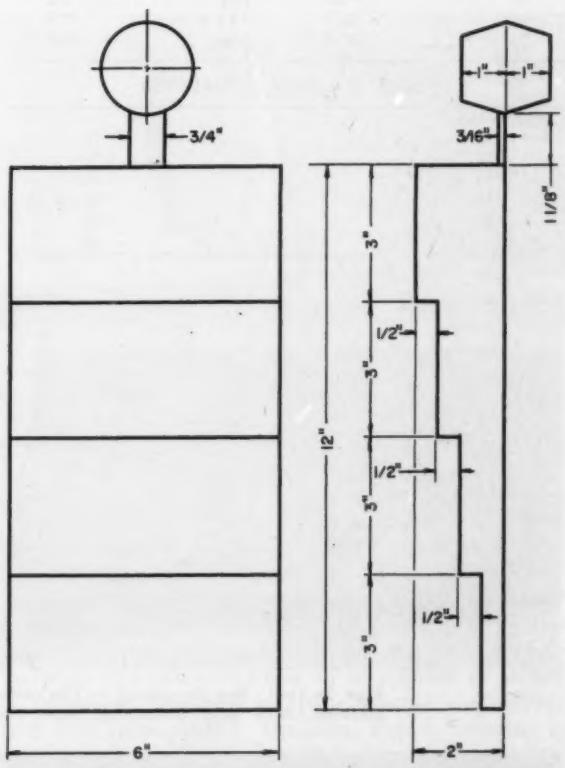


Fig. 1 — Drawing of step casting designed to make a buckled or expansion scabbed casting easily.

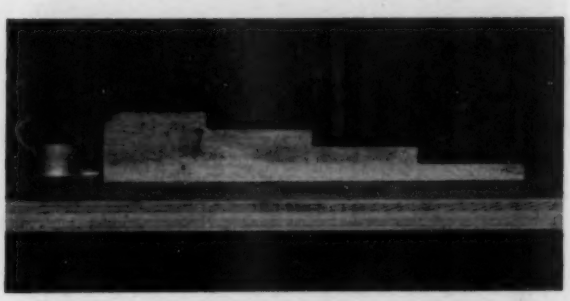


Fig. 2 — A choke ingate was used, the runner and ingate all in one.

*D. C. Williams, "Another Look at Expansion-Type Scabs," *Foundry Trade Journal*, Feb. 17, 1955.

J. E. HALLER is Asst. Supt., James B. Clow & Sons, Inc., Coshocton, Ohio.

TABLE 1 — SAND ANALYSIS

Property	Amount
Moisture,%	7.3
Green Permeability	38
Green Compression	10.2
Green Shear	2.8
Dry Compression	158
Deformation	0.033
Toughness No.	337
Flowability	68
Sample Hardness (2 x 2)	80
Sample Hardness (2 x 1⅛)	89
Hot Strength	212
Combustible,%	6.40
Density (2 x 1⅛) g.	56.5
Density (2 x 2) g.	167.0

Retained on Screen,%

Screen	%	Screen	%
6	0	100	27.0
12	0	140	13.6
20	T	200	6.4
30	0.4	270	2.2
40	2.0	Pan	2.2
50	11.2	AFS Clay	10.0
70	27.2	Gfn	70.9

TABLE 2 — METAL ANALYSIS

Composition,%	
Si	1.79
C	3.28
S	0.055
Mn	0.39
P	0.59
Eutectic,%	—0.20

The gate used for this casting gave a squirting action to the metal. This action, and the way the mold filled when all of the casting was produced in the drag, caused buckles and scabs on the drag surfaces of the casting. No scabs or buckles appeared on the ½-in. step farthest from the gate at an elevation that filled last. Perhaps this was due to using sand with higher moisture content than optimum.

Pouring a casting too slowly, especially with a single ingate, subjects the path of the metal to a higher temperature than is in the remainder of the mold, because it is in contact with entrance metal during the full time the mold is being poured.

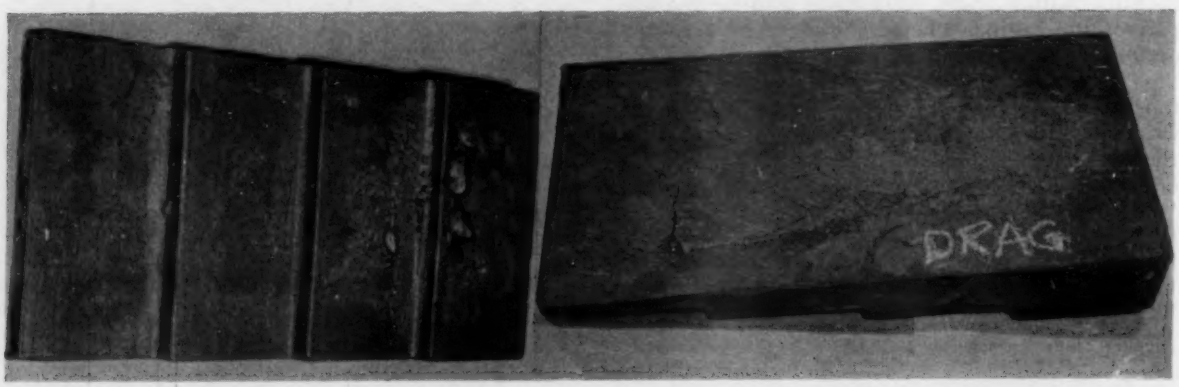
For this test, the first casting was made from a matchplate using a jolt squeeze machine. The sand used contained more than the normal amount of moisture. A typical analysis of the sand and metal is shown in Tables 1 and 2. There were no buckles or scabs on the casting, as shown in Figs. 3a and 3b.

Since this casting did not show the expected defects, it was decided to over-ram the mold with an air rammer. This resulted in the iron blowing out of the mold, as the mold was so hard and wet.

Casting 4 was made entirely in the cope of the mold. The average mold hardness was 90. In Fig. 4a, a shrink or draw can be seen, caused by low ferrostatic pressure with some exudation at the shrink cavity. The drag section of this casting is shown in Fig. 4b. Two buckles and a rattail are on this section.



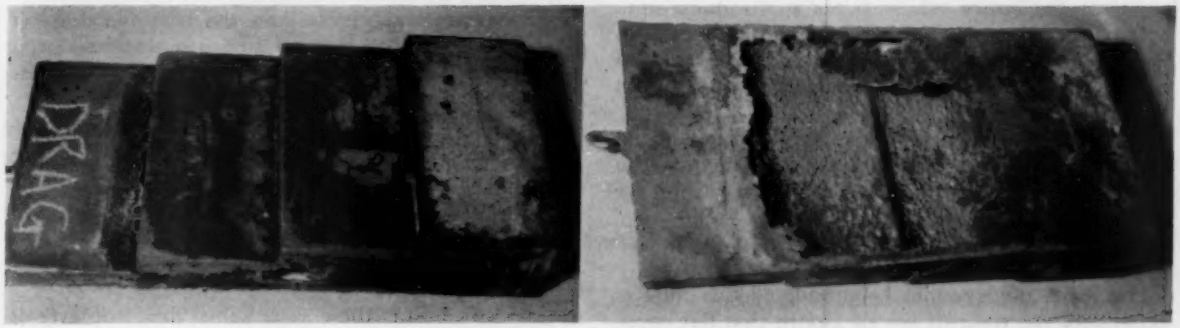
Figs. 3a (left) and 3b (right) — The casting made from a matchplate using a jolt squeeze machine, with sand containing more than normal moisture did not buckle or scab.



Figs. 4a (left) and 4b (right) — There was a shrink or draw because of low ferrostatic pressure with some exudation at the shrink cavity and adjacent to it in casting no. 4. Fig. 4b—Drag section of casting in Fig. 4a.



Figs. 5a (left) and 5b (right) — Casting no. 5 was made in the same manner as casting no. 4, except that it was made in the drag. Scabs occurred on every step except the ½-in. thick step farthest away from the gate.



Figs. 6a (left) and 6b (right) — Casting no. 6 was poured out the gate 12 sec after the mold was filled. The casting is without expansion defects except on the cope molded section. There is a well-defined buckle entirely across the casting at the ½-in. section.

Casting 5 was made in the same manner as casting 4, except that it was made entirely in the drag section of the mold. Iron from the same ladle was used to pour this casting. Scabs occurred on every step except the ½-in. thick step farthest from the gate, as shown in Figs. 5a and 5b. The flat cope in Fig. 5a shows the surface defects caused by the buckled sand in the drag that floated to the cope before solidification took place. From this it can be seen that the buckles in the drag had to take place before the mold was completely filled, or at the time of filling, and when the iron in the second, third and fourth steps was liquid.

Pour-Out Castings

A series of pour-out casting was made from molds following the procedure previously described. Casting 6 (Figs. 6a and 6b) was poured out of the gate 12 sec after the mold was filled. The casting had no expansion defects, except in the cope. At the ½-in. thick section there was a well-defined buckle entirely across the casting.

None of the pour-out castings was strong enough to withstand shot blasting. Discolored sections seen in Figs. 6a and 6b are adhering sand.

Another of the pour-out castings is shown in Fig. 7. This casting was poured out the gate in 15 sec. Two buckles or scabs can be seen on the third step, and one small buckle on the side at the beginning of the top step.

These buckles (Fig. 7) could not have formed from solidification pressure of the iron as it goes through its graphitization period, or from atmospheric pressure at a hot spot, because only a thin shell of iron formed in the specified time.

Additional pour-out castings were made using the same ramming techniques, varying the pouring temperature and pour-out times, in an attempt to obtain a minimum thickness complete casting shell. This was not accomplished. However, Fig. 8, showing a



Fig. 7 — The metal was poured out of the gate in 15 sec. Note two buckles or scabs on the third step, and one small buckle on the side of the top step.

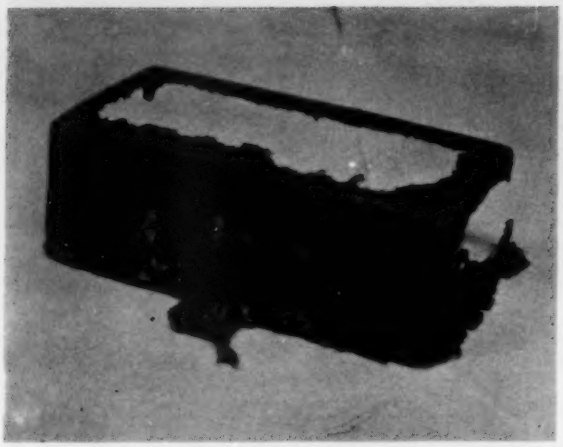


Fig. 8 — A buckle can be seen on the back next to the gate entrance. This casting was poured out so soon that only an outline of the shape is left.



Fig. 9 — This pour-out casting is one where the mold was broken in lifting and pouring out. This caused a fin at the joint of the cope and drag. A small buckle is on the third step.

part of a casting, has a buckle on the back next to the gate entrance. This casting was poured out in such a short length of time that only an outline of the shape formed.

Figure 9 shows the pour-out casting where the mold was broken in lifting and pouring out. This caused a fin at the joint of the cope and drag. Note the small buckle on the third step.

CONCLUSIONS

Hot spots alone cannot be considered the cause of expansion-type scabs. It is well known that flat copes, last to contact the molten iron, are more prone to expansion-type scabs than other parts of the mold. If hot spots or hot areas in the mold and casting were the sole cause, every casting would have these defects at the mold gate.

It is possible that the original hypothesis of this investigation is more applicable to penetration than to buckling or scabbing. The fact that a scab in a sand mold, or an exudation from the top of the downgate on a permanent mold, has different percentages of element composition than the base metal is recognized. Literature has pointed out that at various solidification temperatures varying amounts of the elements are combined.

It is well known that to obtain representative samples of any complex material such as cast iron, drillings must be taken through a full section to overcome errors incident to a partial section sample. This is because of the elements and percentages that combine at various temperatures and times under certain conditions.

In any event, these tests show that one type of scabs and/or buckles is a function of sand expansion, and the occurrence time is as the mold is filling or while the metal is still liquid.

There is nothing a foundryman can do to alter the solidification combination of elements. However, he can control the solidification rate toward the desired characteristics of the metal and casting by ramming, chills, grain fineness, etc.

RADIOISOTOPES UTILIZATION IN THE FOUNDRY INDUSTRY

By M. Pobereskin and D. N. Sunderman

ABSTRACT

The application of radioisotopes is saving American industry 500 millions of dollars each year at the present time. This saving is made with only 0.3 per cent of our industrial concerns using radioisotopes. On this basis a reasonable potential savings would be on the order of 50 billions of dollars per year.

Applications of radioisotopes in the metals industry, while relatively few in number, have contributed measurably to an improved product and a better understanding of the processes involved. These applications include measurement of wear of cutting tools, bearings and refractory surfaces, tracing of chemical and physical species through complex processes, measurement of thickness of solid and fluid metal and surface coatings, corrosion, location of alloying elements in microstructure and the diffusion of metals in alloys.

While the Atomic Energy Commission Office of Isotopes Development is engaged in an extensive program of encouraging further applications, the major barrier to such work is the lack of communication between the scientists in the metal industry who have the problems, and the scientists in the radioisotope laboratories who have the techniques with which to solve them. Widespread uses of radioisotopes in the metals industry do not require new techniques, but only the application of already developed techniques to new problems.

INTRODUCTION

A recent survey by the United States Atomic Energy Commission indicated that industry in this country saves about 500 millions of dollars annually through the application of radioisotopes and their radiations. This remarkable statistic is made even more remarkable when one considers that only about 0.3 per cent of our industrial concerns are using radioisotopes in some fashion. Consider further that a major part of the above savings are being made in the petroleum industry, and in applications involving thickness gaging and control of sheet materials.

Since potential production savings through application of radioisotopes may be estimated at 100 times the above figure, the question naturally arises, "Why are not the applications more widespread than indicated?" We shall consider in part the answering of the above question, and further to providing examples of radioisotope applications that have been made and that may be made in the metals industries.

The factors which have affected the rate of application of radioisotopes in industry are undoubtedly fairly complex, and beyond the scope of this paper to analyze in detail. Lack of familiarity with properties of radioisotopes, exaggerated hazards and costs, restrictive regulations governing use, etc., have all been considered as retarding factors in the application of radioisotopes. The authors feel that another factor, usually ignored, is that present-day radioisotope applications are generally considered completely divorced from their historic background and development.

One could easily form the impression from reading current literature that radioisotope applications were born full grown from the release of atomic energy. While it is true that the broad scope of modern radioisotope applications stem from the atomic energy development, in actuality the release of atomic energy resulted from the study of the nucleus which was sparked by the discovery of radioactivity, and the natural radioisotopes. These studies originated at the end of 19th century with Becquerel's discovery of the radioactivity of uranium, and the Curies' discovery of radium and many of the other natural radioisotopes.

In the first third of this century, many scientists, working with natural radioisotopes carried out fundamental studies in such fields as self-diffusion, crystal growth, solubility, tracer reactions and determination of the age of the earth. Many of these investigations still stand as models in their field, and they provide the fundamental basis for continuing studies to this date. Soon after natural radioisotopes were discovered and some of their properties understood, applications of industrial importance were made.

Early Radioisotope Uses

Everyone was familiar, long before the discovery of nuclear fission, with the use of radioisotopes in producing self-luminescent phosphorus, static electricity eliminators and radiation therapy sources. Such applications, albeit with improvements based upon broader knowledge, are still being made today. It is in this context that present-day radioisotope applications must be considered as being directly derived from, and owing much to the studies of the early workers with radioisotopes.

Based on this background, the aura of awe surrounding radioisotope applications and based on its relationship with nuclear fission can be eliminated. From this viewpoint this relationship falls into its proper perspective; namely, that we are able to produce in nuclear reactors large quantities of radioiso-

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topes of all elements in the periodic table, including many which had never been found in nature. It is this capability which has added a new dimension to the use of radioisotopes.

RADIOISOTOPE APPLICATIONS

Before considering some of the applications of radioisotopes which are of particular interest to the metals industries, let us review the characteristics which make radioisotopes such powerful tools. From the viewpoint of their use as tracers, the basic factor is that radioisotopes exhibit the chemical and physical properties of the elemental species to which they belong. They may be identified, however, in the presence of large amounts of the nonradioactive isotopes, by virtue of the radiations which they emit.

Moreover, each radioisotope may be specifically identified by virtue of the specific type of radiation which it emits, by the energy of this radiation and by its half life (the period during which the radiation intensity is reduced to half its value). All these properties relating to energy may be readily and precisely measured.

Many of the rewarding industrial applications of radioisotopes to date, are based upon the effects matter has on radiation, namely, matter absorbs, scatters and moderates radiation in a manner which depends on the nature of the material and the radiation. These effects have become the basis for a large number of devices for measuring properties of and controlling the production of a wide variety of materials.*

Wear

The problem of metals wear is important to our economy generally, and to the foundry industry in particular. On the one hand the foundry must produce tool steels which wear as little as possible, and on the other hand it must produce metals for fabrication which are readily machinable. From the viewpoint of end use it usually is desirable for the final components to be resistant to wear, also. Application of radioisotopes to study of the metallurgical factors which affect wear is, therefore, of considerable interest.

In measuring wear with radioactive tracers, it is first necessary to incorporate radioactive material in the part whose wear is to be measured. This may be accomplished most simply by placing the part in the neutron field of a nuclear reactor. For an iron part, the radioactivity induced by such irradiation is due to the iron atoms themselves, so that the part has undergone no physical change during irradiation. If it is not possible to irradiate the entire part, radioactive material may be placed on its surface by electrodeposition or vapor deposition. Radioactive metal inserts are sometimes placed at the points where wear measurement is desired.

Wear Test Advantages

The primary advantages of radioactive wear testing methods are high sensitivity, rapidity and ease of operation. The sensitivity is sufficiently high that 10^{-8} grams of wear material are frequently measurable. Other methods based upon weighing or chemical analyses cannot compete either in sensitivity or with regard to rapidity and ease of operation. Moreover, radioactive wear measurements can generally be performed without disassembly of the machine, and may frequently be made automatic with wear results recorded continuously.

Several examples will illustrate the versatility of the technique. Tool bit wear, or machinability tests, can be made by irradiating the tool bit in a reactor for several weeks (Fig. 1). Following irradiation, the "specific activity" of the bit is determined to obtain a precise correlation between radioactivity and weight of tool bit. The radioactive bit is then placed in the lathe or other cutting tool, and chips taken from the stock during machining are radioassayed. The radioactivity of the chips is directly proportional to weight of tool bit worn from the cutting edge.

Bearing wear is similarly easy to measure. The wear part is made radioactive by one of the previous techniques. It is run in a system where the lubricant can be collected and radioassayed either continuously or periodically. The radioactivity of the lubricant is a measure of the wear of the bearing.

Similar procedures are used routinely to determine the quality of lubricating oil in internal combustion engines (Fig. 2). Piston rings are irradiated, and the radioactivity of the oil measured as a function of type of lubricant, type of fuel, water temperature, load and other engine variables.

The application of radioactivity to wear measurement from a somewhat different angle has been reported by Russian and English investigators who determined the rate of wear of a refractory liner in a

*M. T. Simnad, "Review of Metallurgical Applications," *International Journal of Applied Radiation and Isotopes, vol. 1,* no. 3, Nov., 1956.

*"Applications of Radioactive Isotopes and Fission Products in Research and Industry," *Proceedings of the Geneva Conference on Peaceful Uses of Atomic Energy, vol. 15,* United Nations, New York, 1955.



Fig. 1 — Measurement of cutting tool wear and life by radioactivity tests.

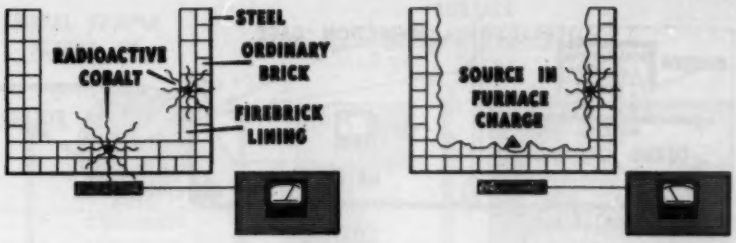


Fig. 3 — Measuring wear of firebrick lining using radioactive cobalt.

blast furnace (Fig. 3). Capsules containing the radioisotope cobalt-60 are placed at different distances from the internal surface of the refractory. Their presence is noted by measuring the radiation outside the furnace opposite the radioactive source. When the refractory liner is worn away to the extent that the capsule is exposed to the molten metal, the source dissolves in the metal.

The disappearance of the radiation source is detected by the measuring the decrease in radiation level outside the furnace. By placing several such sources at different distances from the inside surface, the rate of wear can be determined, and measures taken to replace the liner before damage is done to the shell. The cobalt-60 dissolved in the melt becomes so dilute that there is no hazard in use of the metal. This type of radioisotope application may serve as a form of process control, helping to prevent accidents and reducing the idling of equipment and pig iron loss.

Chemical and Physical Tracing

In the application of radioactive materials to chemical and physical tracing, the techniques are chosen to suit the particular problem. A description of several applications, however, will serve to elucidate the more generally used techniques.

The vapor pressure of metals has been studied extensively with the help of radioactive tracers. In general, the vapor pressure of a metal system, because it is so low, is difficult to measure by other techniques, but by the use of radioactive tracers detection is simple and sensitivity of measurement is increased. Tracers have been used in pure metal and alloy systems up to 2000 C. The common technique is to heat the radioactive metal in a vessel with a small orifice above the liquid level (Fig. 4).

A cold surface is placed above this orifice and the condensed metal atoms measured by determining their radioactivity. Vapor pressure measurements of this type furnish information on metal activity versus concentration in alloys, and the extent of formation of compounds of metal and nonmetals such as silicon, sulfur and phosphorus.

Radioactive tracers can also furnish information on the rate of desulfurization and dephosphorization of metal in contact with slag. Tracers of sulfur or phosphorus added to the metal before slagging will distribute themselves between the metal and slag phases in the same ratio as the nonradioactive sulfur or phosphorus. By taking samples of slag at various times the rate of removal of these elements from the metal can be determined. The effect on this rate of the presence of other metal salts in the slag can be evaluated and the optimum system chosen.

The rate of mixing and melting in furnaces can be studied with radioactive tracers. It was found, for example, that ferro-chrome added to a 190-ton open hearth furnace, melted in 10 to 20 min and was by this time distributed uniformly through the melt. In these studies radioactive ferro-chrome was added to the furnace and small samples taken periodically. Measurement of the radioactivity of these samples indicated when the entire addition had melted and when it became uniformly distributed in the melt.

Mixing Rate Determination

Mixing rates in a steel bath during decarburization were also determined in this manner. When such mixing rates are accurately known it is frequently possible to reduce the residence time in the furnace and increase utilization of equipment.

The source of sulfur in open hearth furnaces was determined by use of radioactive tracers. Radioactive sulfur added to the fuel gas was traced by its radioactivity into the melt and its concentration measured as a function of time. It was found that 10 to 20 per cent of the sulfur in the fuel gas was transferred to the melt during the melting period, while

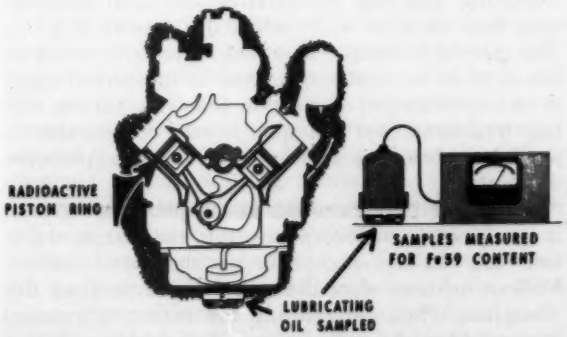


Fig. 2 — Radioactive iron for friction and lubrication studies.

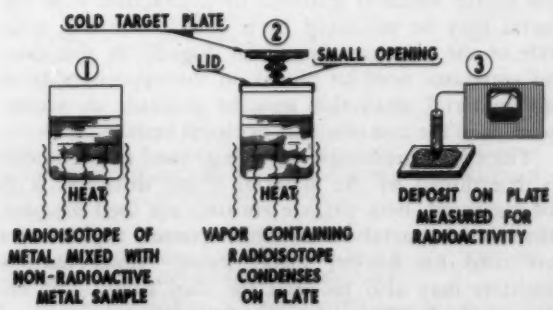


Fig. 4 — Radioactive isotopes for measuring vapor pressure of metals.

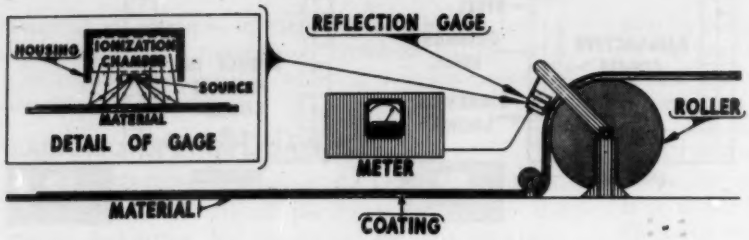


Fig. 5 — Radioactive source of backscattering thickness gage.

only 1 per cent was transferred during the subsequent refining period.

Sources of nonmetallic inclusions in steel may be identified by use of tracers. Radioactivity added to slag, ladle brick and refractory lining was traced to the finished metal part and the source identified. It was found in one case that modifications in the methods of deoxidation and pouring reduced the quantity of inclusions and hence the amount of waste.

The volume of metal in a furnace can also be determined by tracer techniques. A known quantity of radioactive material is added to an unknown volume of metal. When thoroughly mixed, the metal volume may be calculated from the degree of dilution of the tracer.

The kinetics of crystallization of steel has been studied with radioactive tracers. Previously used methods based upon pouring the uncrystallized metal from the mold after a known period of cooling were found inaccurate, since pouring tended to break off dendrites that extended into the melt. In the radiotracer method, a small quantity of radioactive metal was added to the mold at various times during solidification. Sectioning the solid metal followed by autoradiography on a cross-section showed the extent of crystallization at the time of each addition of tracer.

Thickness Gages

The application of radioisotopes to the measurement of thickness is probably the most extensively applied of any radioisotope technique. The general technique is to place the radioisotope source adjacent to the metal material whose thickness is to be measured. The radiation from the radioisotope source is attenuated by its passage through the metal. The intensity of this transmitted radiation is a function of the thickness of the metal and its density. The output signal from a radiation detector on the opposite side of the metal from the source can be calibrated to read directly in thickness.

As an alternative, the fraction of the radiation from the source which is scattered by interaction with the metal may be measured by a detector on the same side of the metal as the source (Fig. 5). In this case, no apparatus need be placed on the opposite side of the material since this may be difficult to accomplish as in the case of a pipe or closed vessel.

The type of radioactive material used in the source is determined by the material whose thickness is to be measured. Beta particle emitters are used for plastics or thin metal films, while gamma ray emitters are used for heavier metal forms. Alpha particle emitters may also be used for thin films. The energy of the beta particle or gamma ray emission must also be selected to obtain the highest sensitivity in the thickness range of interest. As a rough rule of thumb, alpha particle emitters are useful up to about 1 mil of aluminum, beta particle emitters up to about 0.2 in. of aluminum and gamma ray emitters up to several inches of steel.

As examples of the utility of this technique, hot and cold rolled metal strips of 2 in. of iron and 3 in. of aluminum may be measured to 2 per cent accuracy. Wall thickness of tubes or enclosed tanks of up to 1 in. of steel have been measured to 4 per cent accuracy using a backscatter technique. Coatings of metal on metal can be measured by the backscatter technique, if the coating differs in atomic weight from the base. Examples are tin and cadmium on iron which are readily measured in the 0.75 to 0.05 micron range.

The radiation detector output signal can be calibrated in terms of thickness, and used to control the metal-forming process. In such situations, it has been observed that the control is much better than with conventional methods and waste is considerably reduced. Results of radioisotope gaging control of a tandem cold roll mill are shown in terms of the precision with which the thickness of the stock was controlled on the 95 per cent confidence level:

Conventional control	± 10%
Isotope gage with manual control	± 7%
Isotope gage with automatic control	±1.5%

While the uses of radioisotope gaging continue to be increased, the ultimate applicability of this technique has not yet been tapped.

Miscellaneous Tracer Applications

A wide variety of tracer techniques have some applicability to foundry practice, and these will be reviewed briefly here.

Surface reactions, which are of importance in electroplating, pickling, phosphating and sand blasting, have been studied with radioactive tracers (Fig. 6). The general technique is to add radioactive tracer to the solution or solid surface and to measure changes in its concentration or location following various surface treatments. For example, it may be desirable to remove a phosphate coating prior to the application of paint.

By adding phosphorus-32 to the phosphating bath it was possible to incorporate this radioactive tracer into the residual phosphate on the metal surface. Various methods were then tried for removal of this phosphate while determining the extent of removal by measuring the radioactivity of the surface. It was found that an alkaline solution treatment was effi-

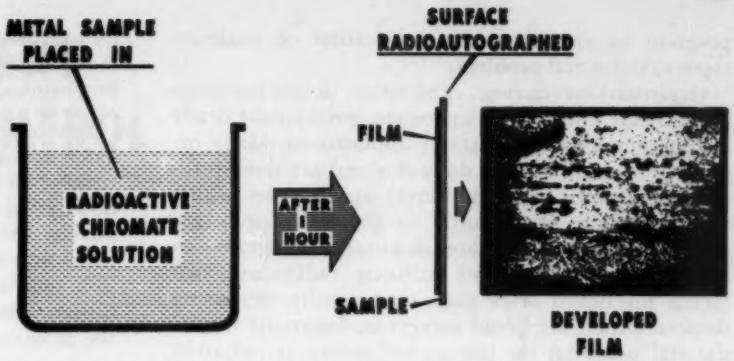


Fig. 6 — Study of mechanism of chromate induced passivity.

cient in removing this phosphate, while sand blasting allowed much of it to remain.

The study of low rates of corrosion has been aided by use of tracers. If the metal is made radioactive and contacted with the corroding solution, the rate of increase of radioactivity in the solution is a sensitive measure of corrosion. In many cases the identity of the corroded specie can be determined simply by instrumental counting techniques without any chemical separation or treatment.

The location of alloying elements in the microstructure of metals has received extensive study with radioisotopes. In this case the alloying element is made radioactive before addition to the melt. The alloy may be cast and forming operations carried out, if desired. Metallographic specimens can be autoradiographed, and the autoradiograph superimposed on the photomicrograph. The first will show only the location of the radioactive element, and the photomicrograph will show its relation to the microstructure. Examples of these techniques include the study of rare earth additions to stainless steel.

In this case, it was found that the rare earths segregated at the grain boundaries, which was the reason for increased machinability. Segregation in lead-bismuth alloys was also studied, and found to be either of the crystal boundary or surface type depending upon the speed of solidification under conditions in linear growth. Tin in bronze, silver in tin, bismuth in copper and aluminum in copper are other systems studied in this manner. Diffusion of metals may be studied in a similar manner by the use of radioactive tracers to determine atomic placement and mobility (Fig. 7).

RADIOISOTOPES FUTURE APPLICATIONS

The expansion of radioisotope uses since 1945, and also future applications, will be based upon the wide variety of isotopic tracers which are available. If one lists all the elements for which radioactive isotopes are available for tracers they may be grouped roughly into three categories. The first group consists of those elements having radioactive isotopes well suited to tracer studies based upon considerations of half-life, radiation characteristics and radiological hazard.

The second group consists of those elements which are more difficult to use as tracers and the third group, those elements which because of short tracer half-life or high radiological hazard should be used only under closely controlled conditions or for short periods of time. These groups are presented in the Table.

Many excellent applications for radioactive tracers are not made because of the lack of communication between those who have the problems and those who have the technique for their solution. Problems which were discussed in the previous section may suggest applications of a similar nature to scientists in other fields. Also, the elements listed in the Table as having suitable isotopes for tracer studies may suggest uses of tracers which have not yet been studied. To overcome this difficulty in communication, the Office of Isotopes Development of the United States Atomic Energy Commission has initiated a long-range

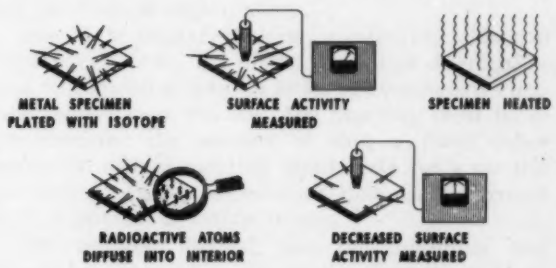


Fig. 7 — Radioactive isotopes for studying solid diffusion.

TABLE — ELEMENTS WHOSE ISOTOPES ARE USEFUL IN RADIOTRACER STUDIES

Group 1. Best Characteristics as Tracer			
sodium	yttrium	cesium	hafnium
scandium	zirconium	cerium	tantalum
chromium	niobium	praseodymium	tungsten
manganese	technetium	samarium	osmium
iron	ruthenium	europium	iridium
cobalt	silver	gadolinium	mercury
zinc	antimony	terbium	thallium
selenium	tellurium	erbium	lead
krypton	iodine	thulium	
rubidium	xenon	lutecium	
Group 2. Average Characteristics			
hydrogen	nickel	rhodium	ytterbium
carbon	copper	palladium	rhenium
phosphorus	gallium	cadmium	platinum
sulfur	germanium	indium	gold
chlorine	arsenic	tin	bismuth
argon	bromine	barium	thorium
potassium	strontium	lanthanum	uranium
calcium	molybdenum	holmium	
Group 3. Difficult to Use			
magnesium	aluminum	polonium	radium

program to encourage the application of radioisotopes to industrial problems.

Organizations having experience in radioisotope research are encouraged to request government funds for the development of tracer applications. Many organizations have already done this and are now working on entirely new and novel applications which will contribute significantly to industrial processes. Some of these studies are short-range programs to develop new instruments utilizing radioactive materials for liquid level gages and density measuring devices. Others are broad surveys to determine the industrial potential for isotope utilization as radiation sources and tracers.

One study which has the potential for wide applicability in industry is determining the feasibility of intrinsic radioactive tracers for industrial process control. An intrinsic tracer is one which is added to the material being processed. Process control would be based upon measurement of the concentration of this radioactive tracer (i.e. specific activity) in the process material. An example of such a system might be the addition of a radioactive isotope of an impurity to iron ore before smelting. A measurement of the radioactivity of the iron would then be substituted for chemical analysis for this constitutent.

Another application would be to tag a particular batch of ore when loaded into the smelter. At the time when melting is complete, and the tagged batch flows from the furnace, a measurement of its radioactivity will serve to differentiate it from the previous batch and the one following it. This type of application may simplify industrial process control of the future, lowering costs and improving the quality of the product.

It is really a misnomer to refer to the above as applications of the future. The isotopes are available now. So are the techniques for handling them and measuring their radiations. What is necessary is that the technologists engaged in the production processes discuss their problems with the scientists trained in tracer techniques. Through this exchange of knowledge between the respective fields, the full potential of this technique can be realized to the benefit of American industry.

PROFIT MANAGEMENT THROUGH COST CONTROL

By R. B. Hill

ABSTRACT

Profits can and should be planned for. If the plan is to be realized, manufacturing costs must be controlled. Cost control is a frame of mind; to be successful it must have top management support. The various specialized techniques for the collection and reporting of costs are important, but really worth-while results can never be obtained until profit-mindedness and cost-consciousness are created in the minds of all company personnel. Then, it is truly possible to obtain the management of profits through cost control.

INTRODUCTION

Profits do not just happen, they must be produced. It is the basic task of management to produce a profit representing a fair return on capital invested.

Profit management means the formulation of a plan resulting in a net profit. If the profit objective is to be realized, control must be exercised to ensure that the plan is adhered to closely.

A major portion of the overall plan for profit must be a plan for manufacturing. When the details are expressed in dollars, they become a manufacturing costs plan. It is only through the control of these costs that the profit goal can be attained.

Webster's dictionary states that *"control* is to exercise a directing influence over." This definition places the responsibility for cost control with management. Cost control is an integral part of management, and must be recognized as a regular part of management's job, not as a special project. When written job descriptions are prepared cost control should be included as a specific responsibility of management at all levels.

TOP MANAGEMENT MUST LEAD THE WAY

How does a foundry get started on a cost reduction program? To be a success, a top management decision to embark upon such a program must come first. This is because the program must have the support of the entire organization, and because it will involve some additional expense particularly in the early stages. It is management's job to see that all levels in the company are sold on the benefits to be obtained from the cost control program.

Cost control is an attitude, a frame of mind, and management must strive to develop a cost-conscious attitude throughout the company. It must be truly sincere in its belief in the value of such a program, for if it is not supervisors and foremen will surely sense it and the program will fall apart.

COSTS CONTROL

There are generally considered to be three main categories of manufacturing costs: 1) direct labor, 2) direct material and 3) overhead.

Direct Labor

There are many text book definitions of direct labor. The definition the author likes is: All labor that can be practicably charged to an individual job should be classified and collected as direct labor. What is practicable to pick up as direct labor will vary from plant to plant, dependent upon the type of operation, its size and the technique or system of labor distribution employed.

One of the primary reasons for collecting labor cost information is for use of the control of direct labor cost on individual jobs. In order to exercise any effective control over this cost, the first step must be to predetermine the amount of time a direct labor operation such as molding should take for a particular casting. This predetermined time can be arrived at in a number of different ways.

The simplest method, and the minimum that should be employed by the smallest foundry, is an estimate of direct labor time by the foreman for each of the direct labor operations such as molding and coremaking. A small, simple job cost card should be used by the foreman when preparing his estimate. Filling out the card ensures a systematic approach to the estimate, and reduces the otherwise strong possibility of overlooking some necessary element of direct labor cost.

When overhead has been applied as a percentage of direct labor, metal cost added, plus an overall allowance for scrap, a cost estimate that can be used by the sales department in establishing the casting price is the result. This is the first use to which management usually puts the predetermination of direct labor costs, and it is an important one. In effect, when a mark-up has been assigned to arrive at a selling price, the development of a profit plan has been extended down to an individual job.

All too many foundries stop at the preparation of this job plan, without taking sufficient steps to exercise control which will enable the profit plan to be realized. In its simplest form, the steps that can and should be undertaken are to advise the direct laborer and his supervisor of the time allowed for in the es-

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timate for the job. The work assignment is then known before the job begins, which gives the direct laborer and supervisor a chance to make the plan workable.

After the first one or two castings have been produced on a new job, the actual time on the job should be checked with the estimate. If the actual has exceeded the estimated time, the job should be studied to see if the pace or method can be improved to a point where cost can be reduced sufficiently to recoup on the remainder of the order. This is an extremely simple system, and one which will exercise a significant degree of control at a low expenditure for the small shop. Its greatest weakness lies in the quality of the informal estimate, which is prepared by the man against whose performance it is to be used as a measuring stick.

Foundry Estimator

To partially overcome this objection, and to enable the handling of a greatest volume of estimates in the larger sized shop, a foundry estimator is appointed. Typically, the estimates are now a compromise between the estimator and the foreman, if the estimator is a man of considerable shop experience. Otherwise, he simply becomes a clerk who enables a larger volume of estimates to be processed.

If the estimator is a man experienced in foundry estimating, and with a thorough knowledge of a particular foundry, the estimated times attain sufficient accuracy to begin the exercise of a more complete type of direct labor cost control.

On the day following the production being considered, a daily report can be prepared for each separate area for which a supervisor is responsible. This report should list the molder's name, the pattern number, the number of pieces produced, the estimated time per piece, the total estimated time for all the pieces produced from that particular pattern, the total actual direct labor time charged for the job and the variance of actual from the estimated time. This is a simple report that can be prepared by the timekeeper. It should deal in man-hours only.

This daily report highlights the individual jobs which actually exceeded estimated time, and which, therefore, need individual attention. If the estimated times for all the jobs and actual times reported are totalled, it enables the company to calculate a percentage of direct labor efficiency for the particular department. This procedure is inevitably rewarded by an increase in effort by the supervisors to improve and then maintain their level of efficiency.

The greater the accuracy of the estimated direct labor time, the more effectively they can be used by management for cost control purposes. The next increase in the degree of control that is possible to exercise over direct labor costs results from the use of measured day work. Measured day work is the term used to describe the direct labor control system, whereby engineered time standards are established and used to predict the direct labor cost of a casting.

A significant increase in the degree of control that can be exercised results from the use of work measurement standards, since the estimates accuracy degree is greatly increased. Since, however, direct labor is still paid on an hourly basis, there is no incentive for labor to reach and maintain a particular production. Direct labor cost per unit will then fluctuate with changes in production from day to day.

Wage Incentive System

The final and most effective direct labor control is provided by a wage incentive system based upon direct labor time standards established through work measurement. The plan of incentive payment which is receiving the most widespread acceptance today from both management and labor is the standard hour plan. It compensates labor for production over standard on a one-for-one basis. Under this incentive plan, if a worker produces 10 per cent more than the standard for the job, he receives incentive pay equal to 10 per cent of his negotiated hourly rate.

Here is the ideal situation for direct labor cost control, as long as the worker at least reaches standard on the job. Above standard, the direct labor cost per unit is constant. Most labor contracts covering incentive plans today permit management to remove a man from a job, if he is unable to meet an accepted standard, within a relatively short period of time. The use of incentives then relieves the supervisor of much of the burden of obtaining adequate rates of production, leaving him more time for other extemely important jobs such as control of departmental expenses.

The exercise of control over direct labor cost is essential if manufacturing cost is to be controlled, and the profit objective realized. There are many methods of controlling direct labor, but the best one to use depends primarily upon the foundry's size. The important thing is for top management to decide to take action and set the stage for an effective program of direct labor cost control.

Direct Material

By direct material it is meant all material which may be practicably charged to an individual pattern number in the foundry, where direct material is made up primarily of cold metal cost. This, of course, includes pig, scrap and alloys.

Typically, there is nothing much a company can do to control the prices that it pays for its cold metals. However, there is something effective most foundries can do to control the cost of their cold metal consumption.

Customers' metal specifications are normally expressed in terms of the chemical analysis of the metal and its physical properties. This leaves the foundryman free to brew his own mixture so long as the end result satisfactorily meets the specification. Often several different mixtures can be used to provide the same specification.

The prices paid for scrap vary in relation to pig, and between the various kinds of scrap. In 1954, for example, steel scrap was selling at only ⅔ the price of cast scrap, whereas in less than two years' time it was commanding a premium over cast scrap.

Where practicable, two or more different techni-

cally satisfactory mixture sheets should be drawn up for each different type of metal required. At least once each month these mixture sheets should be extended into a cost per ton, using the current purchase price of all metals included in the charge. The lower cost mixture should then be selected for use in the melting operation. The quantities of the various metals and alloys to be purchased should then be reestablished according to the planned consumption.

The melting and purchasing supervisors must keep in close touch with one another in a rapidly changing market, in order that the company might produce the lowest cost mixture that will meet the specifications.

Overhead

The remainder of the cost, making up total manufacturing cost after direct labor and direct material, is referred to by several different names. Some of these are overhead, indirect manufacturing expense, burden and on-cost.

In the typical ferrous foundry overhead comprises from ½ to ⅔ of the total manufacturing cost. In nonferrous foundries, the proportion is considerably lower due to the relatively high metal cost of the various alloys produced. However, if we look at the overhead as a percentage of controllable cost, which is the basis for competition between foundries, then overhead becomes about 55 to 80 per cent of total foundry cost of producing all types of metals.

The specific percentage depends primarily upon the degree of mechanization of the foundry. This high proportion of total cost makes the control of overhead of vital importance to the company's success.

Overhead cost, more than any other, is a reflection of top management policies and decisions. The size and type of staff, the amount and type of equipment, the kinds of services supplied to direct labor—and other factors making up a sizeable portion of the overhead cost are determined by management decisions. Some of these policies and decisions concern the type of casting a foundry is to produce, the market to which it plans to sell, the services which it wishes to provide to its customers and the rate at which the company plans to grow.

Once a control program for direct labor has been established, front line supervisors can generally be counted upon to hold relatively tight rein over their cost. During the conduct of their direct supervisory duties, it will become apparent to them when there is insufficient work to be performed and they will make the necessary lay-offs. With the use of the predetermined times, they have means of controlling daily direct labor productivity and become aware on a continuing basis of the responsibilities in this area. The same is not true generally of overhead cost.

The front line supervisor must also be the key man in expense or overhead control. All too often he, along with many members of management, consider overhead to be essentially a fixed noncontrollable cost. This is not true in the foundry.

Only about 10 per cent of total overhead cost is typically made up of such truly fixed items as depreciation, light, heat, taxes and insurance. Even these are only fixed in the sense that they cannot be controlled from period to period. They can be altered on a longer term basis as the result of management decision and action. All of the remainder of overhead cost is variable to some significant degree. Some of the most controllable of these costs are 1) departmental indirect labor; 2) production supplies (indirect materials); 3) equipment maintenance and repairs; 4) pattern and flask expense, and 5) general plant labor.

The most effective way to control departmental labor is through the use of daily, or at least weekly, reports showing the amount of indirect labor in relation to direct labor.

Monthly Cost Statements

This report should be submitted to management as well as to supervision. It is the exceptional supervisor who does not need periodic prodding by management to keep his indirect labor in line with direct labor. To be sure, the ratio of indirect labor is bound to increase somewhat as volume decreases, but by careful planning the supervisor, in the re-allocation of workers' duties and a tightening up of supervision, can do much to keep it in line.

The remainder of overhead costs can be most effectively controlled by management through the analysis of monthly overhead cost statements, which must also be distributed to supervisors for their individual departments. If these monthly overhead reports are to be effective they must be based upon an adequate chart of cost accounts. There must be a sufficient number of accounts with their content clearly defined so that a particular figure on the cost statement may be quickly related to some physical aspect of operation.

Suggested charts of accounts, complete with the description of the items of cost to be charged to each, are readily available from trade associations. In particular, the Gray Iron Founders' Society, the Malleable Founders' Society and the Non-Ferrous Founders' Society have prepared excellent cost manuals available at nominal cost.

The simplest type of cost report lists, for each department and each cost account, the amount of money spent during the period. Unless a budget has been prepared in advance, the one set of figures by themselves are not of too much value. It should be compared with figures for fairly recent previous months. The differences in volume from month to month in the jobbing shop in particular make it difficult to compare absolute dollar amounts between two or more sets of figures. This can be remedied to a significant degree by expressing each one of these items of cost in cents per pound or dollars per ton of good product produced.

This greatly reduces the most serious distortions due to changes in volume, and enables the spotting of other variances for study to determine the reasons for the increase. An effective technique is to develop a target overhead cost statement by recording the

best verified cents per good pound figure for each cost item, and using it as a standard of comparison each month. Each time a new low for an item is achieved, it should be substituted for the previous low on the target statement.

An effective method of cost control for the larger foundry is the preparation and use of an indirect manufacturing expense or overhead budget. The first step in the preparation of this overhead budget is to make a sales budget. Then, by an analysis of current costs and plans for cost reduction in the near future, a manufacturing plan can be made. This plan, expressed in dollars and cents, for each cost account and each department, becomes the manufacturing cost/budget.

Great value is obtained from the preparation of this budget, since it forces all levels of management to plan ahead for at least the period covered by the budget in order to arrive at the budgeted cost figures. The usual practice is to budget on a monthly basis for one year in advance.

Budget Disadvantage

One disadvantage of this type of budget is that the absolute cost figures budgeted for each month are dependent upon the amount and type of work produced. This disadvantage can be minimized by budgeting the amount of direct labor for the particular volume forecast, and then by expressing each item of overhead as a percentage of direct labor. If effective control over overhead is being exercised, these percentages should tend to remain constant as volume decreases.

This obviously does not apply to such fixed costs as depreciation and insurance, but these percentages can be left off the report or ignored, since it is fully recognized that they are not controllable by this method.

Often this type of budget is a forerunner, in a larger company, to the preparation of variable budgets. This type of budget eliminates, or greatly reduces, most of the objections to the fixed budget. It is prepared by analyzing every expense account to determine what cost items are essentially fixed and which are variable. For each variable item a decision is made as to with which readily measurable characteristic of operation it varies most closely. Typical examples are direct labor, tonnage and time or working days per month.

With this system of budgeting cost control, the accounting department calculates how much should have been spent in total dollars for each account on the control statement, under the actual conditions of volume and so on, encountered during the particular month. This has tremendous appeal in dealing with production supervisors who are so well aware of the effect of volume on costs that it tends to cloud their appreciation of significant cost information provided by the fixed budget and its percentage figures to direct labor.

Regardless of which of the above general techniques is used to point out the areas of cost which are out of control, the most effective approach to bringing these costs back into line is the same. Meetings should be held every month between the department supervisors, the cost department and management. Prior to these meetings, all supervisors should have received the cost statements for their own department and should have made some study on their own, and with the cost department, as to the reasons for their out-of-line costs.

At the monthly cost control meetings the supervisors should be called upon to explain as much as possible the reason for their out-of-line costs, what steps they have already taken to remedy the situation, what steps they plan to take and to ask advice or assistance in reducing costs in any area which is out of control. Make sure at these meetings that the supervisors know where they are going, and how they are doing. Do not pass up the opportunity to comment on these items of cost that have been brought back into line by the supervisors, or on items in which a department has shown marked improvement.

CONCLUSIONS

Do not try to analyze and control too much every month, otherwise nothing receives sufficient time and attention to achieve effective results. The setup, or layout of general cost control statements to be submitted each month, should remain the same over a long period of time unless there is an important reason to change. This is necessary to provide an effective measure of progress being made. The number of these reports, however, should be at an absolute practical minimum.

When an individual area of cost, represented by one figure on the basic cost control statement, is out of line, and the supervisor does not know the reason, or feels he has insufficient information to enable him to bring it under control, more detailed special reports should be planned. The layout and type of these reports should be determined by the supervisor requesting them, not by the cost department, although the ideas of the cost department should be solicited. If the supervisor has an opportunity to say what it is he wants reported to him, and in what way, he will make far more effective use of the information when he receives it.

This type of reporting should be flexible, so that concentration can be brought to bear on the troublesome areas. It is important that these special reports be discontinued as soon as the troublesome area has been brought under control, otherwise the cost department gets bogged down with such a heavy load that it has no time left for analyzing cost accounts or working with the supervisors on cost reduction.

When more items of cost are out of line than it is practicable to concentrate on at one time, the order selected for tackling these jobs should be dictated by the absolute dollar magnitude involved. For example, it will be potentially much more profitable to attack a $20,000 item of cost that is 5 per cent out of line, or $1,000, than a $500 item that is 25 per cent out of line for only $100.

DESIGN AND WELDING OF ALLOY CAST STEEL

steam turbine applications

By L. W. Songer

ABSTRACT

The service requirements of modern steam turbine-generator equipment necessitate that castings and cast weldments be of the highest quality and have maximum reliability. The requirements for proof of quality, and control of all phases of manufacture, are far more exacting than the requirements for structural castings most foundries are accustomed to making.

The design, welding and quality control aspects of high quality castings are discussed relative to greater participation by foundries in these important phases of manufacture. Ways of protecting the present market, and regaining lost business through a better understanding of the customers' needs, are pointed out.

INTRODUCTION

The design of modern steam turbine-generator equipment depends to a great degree on castings. The present-day high pressure and high temperature operating conditions require that castings of the highest quality be designed and produced to meet both the ever increasing demands of service and economic manufacture.

Many castings are produced only a few times, therefore, there is little opportunity to gain experience by making and evaluating several castings. Close cooperation is required between the foundries and design engineers since the first casting generally must be used.

Starting in 1905, with temperatures of 500 F, the steam turbine industry today has reached operating temperature levels of 1150 F (Fig. 1). It is interesting to note that this has resulted in an average increase over the years of 12 F per year.

The development of alloy materials and the improved designs of turbines and boilers has permitted a continual increase in the initial pressures for steam turbines from the 500 psig to the 4500/5000 psig of today's highest pressure turbines (Fig. 2). This increase in temperature and pressure, along with many design improvements over the years, has improved the overall efficiency of steam turbines, as shown in the station coal rate (Fig. 3). The improvement in the coal rate has been reflected in the lower cost of electricity.

Recent surveys have shown that the nation's electric power generation is increasing more than twice as rapidly as the rest of the economy. The use of electric power is expected to continue at a high rate of in-

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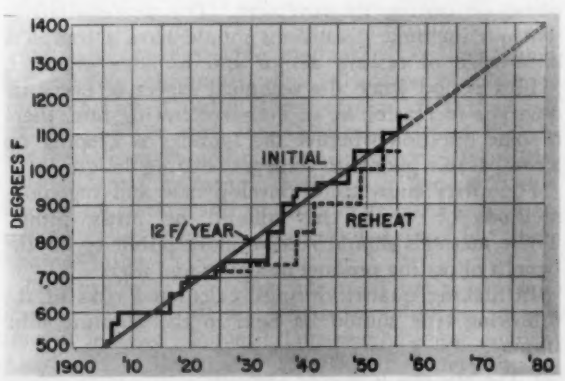


Fig. 1 — Maximum steam temperatures of central station steam turbines.

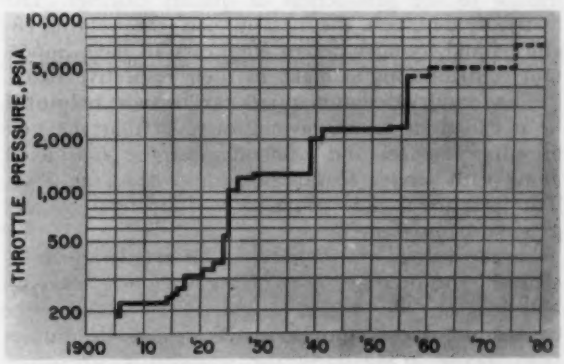


Fig. 2 — Maximum pressure of central station steam turbines.

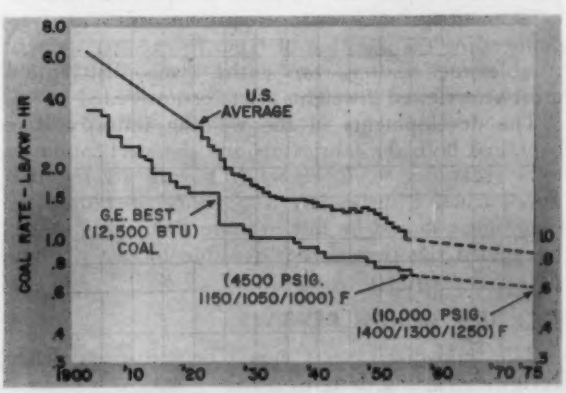


Fig. 3 — United States station coal rate.

crease for the next 20 years, and with this will come further advances in pressures and temperatures. Steel foundries can share in a greater portion of this potential business, and thereby better contribute to our way of life.

Design and welding aspects of large alloy-steel castings for use in steam turbines operating at temperatures up to 1050 F and at pressures up to 3500 psi are discussed. The castings may vary in weight from 1,000 to 85,000 lb. Comments will be given regarding the use of forged and fabricated parts to replace castings, and how business lost to such vendors might be regained through more direct consideration of the customers' needs.

CASTING DESIGN

The design requirements of castings for the power generation industry today are, 1) reliability, and 2) designs which can be produced at reasonable costs. Good designs do not happen by chance. They are the result of a carefully thought out and calculated engineering approach to produce a component which will satisfy a given set of conditions.

The designer must have a thorough knowledge of cast and wrought materials, and must be willing to use these materials either singly or in combination if quality components are to be produced. In addition, he must consider other important design criteria such as structural strength, heat flow and distribution in the component during service, method of casting, welding, machining and nondestructive inspection. He is only one of a group of engineers who must contribute to the production of quality castings. The materials engineers, the metallurgists, the welding engineers and foundry design engineers all have important contributions to make in their respective fields.

The foundry design engineer can be most helpful if he is called in at the layout stage, so that the best foundry practices and principles can be considered along with service requirements and optimum costs. This practice is now being successfully used by progressive manufacturers.

Large castings are inclined to be overdesigned because of the necessity of tapering walls for considerable lengths. This excess weight not only makes the casting more costly but also increases the costs of other components, particularly the supporting structures. Since stress considerations are of the utmost importance, it is necessary to strike an optimum balance between design and casting requirements. When recommending the addition of taper to existing walls of troublesome castings, the entire stress distribution must be reviewed if weights are to be lowered.

The developments in the welding industry have benefited both the fabricators and the steel foundries to a great degree. Since welding with its required nondestructive inspection is a costly operation, no component should be cast in more than one piece to be welded together if it is possible to cast a single part of the required quality.

CASTING DESIGN REVIEWS

The following aspects should be considered when reviewing casting designs:

1) Function of the component and critical areas.
2) Service record of similar castings.
3) Operating temperature and pressure.
4) Material specification and test requirements.
5) Heat treatment.
6) Weldability of material.
7) Welding specification – including code requirements and extent of repair welding permissible.
8) Post-weld heat treatment.
9) Number of castings to be ordered.
10) Type of pattern, core boxes, new or made from an existing pattern.
11) Position in mold for casting.
12) Projections which could possibly cause casting difficulties. Should they be removed, cast separately and welded or bolted on?
13) Should the main casting be made in more than one piece? If so, can an adequate weld preparation be made which is accessible for nondestructive inspection?
14) Junctions of unequal walls, flanges, pipes, etc.
15) What directional feeding is provided by actual drawing dimensions, internal and external padding? Will the padding be removed and how?
16) Internal and external chills (Internal chills used only with purchasers' permission).
17) Tolerances on wall thickness and as-cast surfaces.
18) Stock allowable on finished surfaces.
19) As-cast surface finish requirements.
20) Nondestructive inspection requirements.
21) Tie bars for maintaining dimensions during casting and/or heat treatment.
22) Location of risers, favor placing them on machined surfaces.
23) Should a scale model be made?
24) Sample castings for smaller parts.
25) What new foundry procedures might be used to improve quality or make casting of unusual contours possible, such as the use of exothermic materials in certain areas?
26) The final consideration is a recheck to see if the overall weight can be reduced without jeopardizing the casting.

Foundry Functions

Foundries should have complete control over the design and planning of patterns. Designs and patterns should be developed to suit a particular foundry's facilities. One way to initiate control is to educate the customers' product designers to show wall tapers, feeder pads and other requirements, which will remain on the delivered casting directly on pattern drawings, and to dimension them. Present-day casting problems cannot be left to the skill of the patternmaker and molder alone.

In general, foundries are reluctant to ask for design changes on castings used by the manufacturers of large equipment for fear that some may think they are incompetent to judge. This belief is wrong. Competent product design engineers respect foundry engineering advice, and are willing to discuss any phase of casting improvement. The representatives sent out for these discussions should have a complete knowledge of casting design and foundry practices.

In a period when the technical aspect of every industry is increasing at an ever-accelerating rate, there is some question whether the foundry is keeping up a comparable pace with respect to casting design.

Foundries many times develop new and ingenious methods of which they should be justly proud. These methods should receive more publicity for the benefit of casting product designers and users.

In making quality castings, calculated risks of the following type should be held to the absolute minimum:

1) The heat that is thought adequate.
2) The core that is baked too hard, or burned.
3) The risers that are too small, or not completely filled.

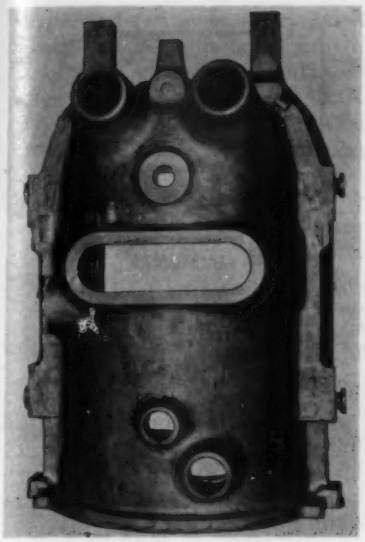


Fig. 4 — (Left) Turbine shell casting designed for separate welding on steam chest.

Fig. 5 — (Below) Steam chest with an integral inlet flange, weld preparations partially machined.



4) The riser that is not properly attached.
5) The flask that is too small.
6) The worn-out and distorted pattern.
7) The broken core that should have been scrapped.
8) The mold that is shaken out too soon.
9) The gating which never worked in the past.
10) The dirty mold.
11) The extra chills that no one will ever find.
12) The casting that is cast in the wrong position just to suit existing flasks.
13) The molding practices which are left entirely to the molder.
14) The metal that is poured too cold.
15) The riser that is burned off without proper preheat.
16) The inadequate heat treatment.

These items can and must be controlled for high-temperature, quality castings. Casting defects are seldom an accident.

Present-Day Designs

High-quality complex castings are produced today by making the casting in more than one piece for subsequent welding. Figure 4 shows a turbine shell designed for use with a separately cast and welded-on steam chest. Figure 5 shows the separately cast steam chest. Figure 6 shows a valve casing with a separately cast flanged elbow, assembly welded. Figure 7 shows a weldment for a steam inlet connection. Three castings and two pipes were welded at the locations shown. The angle inlet tee was designed to save space and eliminate one weld.

Centerline shrink in the walls, and shrinkage at the seat fit, have caused considerable difficulty in many large valve casings. One method of casting which has entirely eliminated the shrinkage problem is shown in Figs. 8 and 9. These casings are cast in the vertical position with the walls tapered and the excess metal is machined out. While this may seem like casting ingots, the overall quality is so good that this method of casting and machining is less costly than making castings in the conventional way and then repair welding.

This example points out the need for a complete understanding of all manufacturing processes. Many times it is more economical to purchase one higher-priced and higher-quality component if the remaining processes can be completed with less work, and the final cost will be actually lower.

Figure 10 shows how a hot spot and the attendant shrink are avoided by making the stock for finish parallel to the finished surfaces on stub pipe and weld preparations.

WELDING

The composition of the alloy steels, the weld deposit and the heat treatment of these materials used for castings and welding at the writer's plant, are shown in Table 1.

Weld Preparations

The guiding thought in the design of all types of weld preparations is to obtain fusion throughout the entire wall thickness, while at the same time keep-

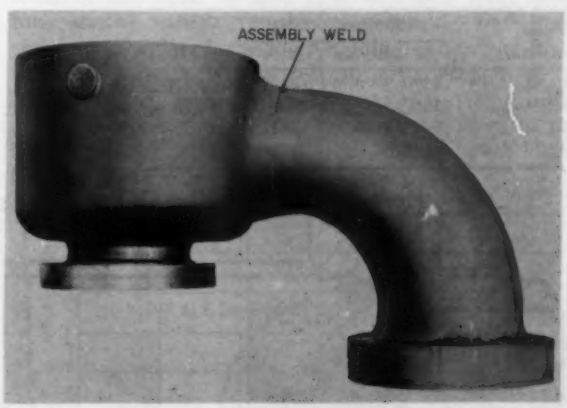


Fig. 6 — Valve casing with separately cast flanged elbow, assembly welded.

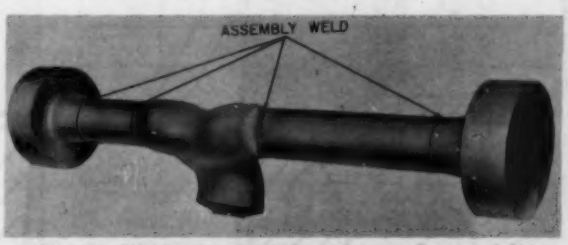


Fig. 7 — Weldment, consisting of cast flanges, inlet tee and two pieces of wrought pipe.

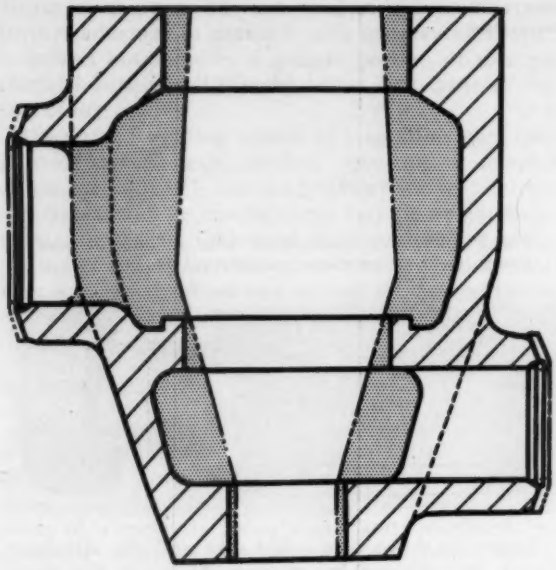


Fig. 8 — Stop valve casing cast with tapered walls which are machined to contour.

ing weld shrinkage and the attendant locked-up stresses and distortion to a minimum. To secure full penetration on a consistent basis, proper access to the root is a prime requisite, which allows the welder to follow the welding progress and to maintain control of the arc. This control is particularly significant where lack of penetration or excessive penetration causes defects on the inside of the joint, which cannot be corrected by removal and rewelding from the inside.

TABLE 1 — COMPOSITION AND HEAT TREATMENT OF MATERIALS

Material	Composition, % Cr	Mo	Mn	C	V	Heat Treatments
Cr Mo	1.25	1	0.70	0.20		Normalized 1920 F min.
Cr Mo V	1.25	1	0.70	0.18	0.20	Tempered 1250 F min.
Cr Mo Weld Metal	0.50	1	0.90	0.12		

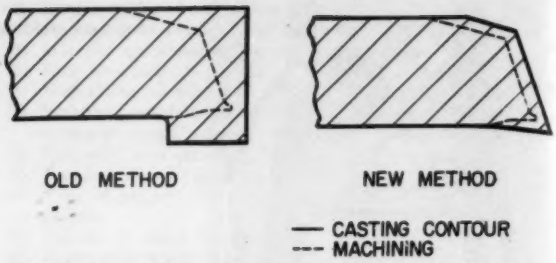


Fig. 10 — Method for adding stock on machine finished weld preparations.

To minimize shrinkage, distortion and fabrication costs, it is desirable to keep the amount of deposited weld metal to a minimum. These requirements, together with the welding process to be employed and any special conditions imposed by specific application, must receive careful consideration.

Figure 11 shows typical weld preparations for joining heavy-walled sections. Figures 11a, 11b, and 11c show weld preparations used for manual metal-arc welding and for manual or mechanized submerged-arc welding. Figure 11d shows the preparation for inert-gas metal-arc welding of the root pass. This preparation can be used either with or without a consumable backing ring. Backing rings of either the integral or separate type are machined out where possible (Fig. 11a). It is also advantageous to machine a secondary angle on the outside diameter of casting (Fig. 11a), to insure uniform weld thickness by having a uniform outside diameter.

WELDING METALLURGICAL ASPECTS

Preheat

Welds which have a high degree of restraint and high residual stresses are encountered during fabrication and repair welding. The size of the planned welds ranges from 1 to 6 in. in wall thickness, many being in the 3 to 6-in. range. Also, the usual contour of casting defect repair welds is of an elliptical cup

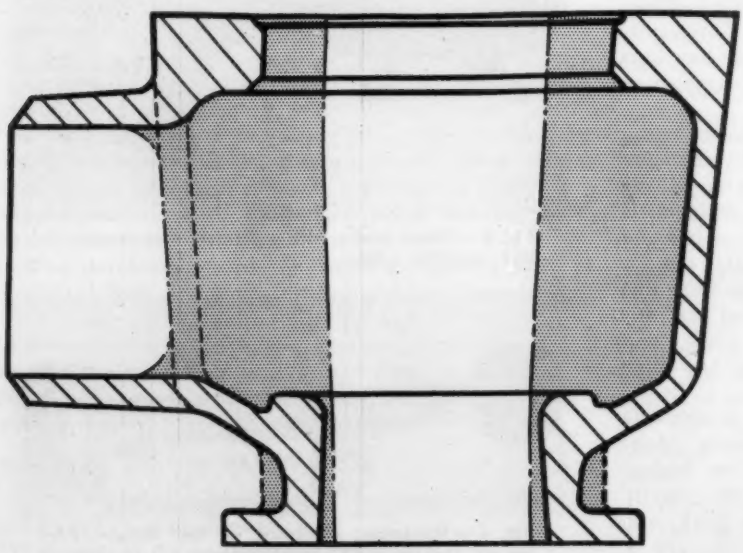


Fig. 9 — Intercept valve casing cast with tapered walls which are machined to contour.

shape, and can be in heavy overall sections such as a flange which might be 24 in. thick. In large section sizes, high multiaxial residual welding stresses can cause material failure with little deformation.

For planned and repair welding, the components are preheated to 450 F minimum. This temperature is above the impact transition temperature of the base materials (Fig. 12).

The impact test is an indication of the resistance of the material to triaxial stresses under a notch. As mentioned previously, repair welds in large sections can have high triaxial residual stresses. The use of preheat above the impact transition temperature holds the material in a temperature range in which its resistance to failure under triaxial stresses is increased.

Experience indicates that the maintenance of the preheat during welding is imperative to the successful processing of these highly restrained welds. Furthermore, it is advantageous to continue the preheat temperature until postweld heat treatment can be effected. This eliminates the possibility of having welds with high residual stress subjected to temperature ranges where their resistance to crack propagation is decreased.

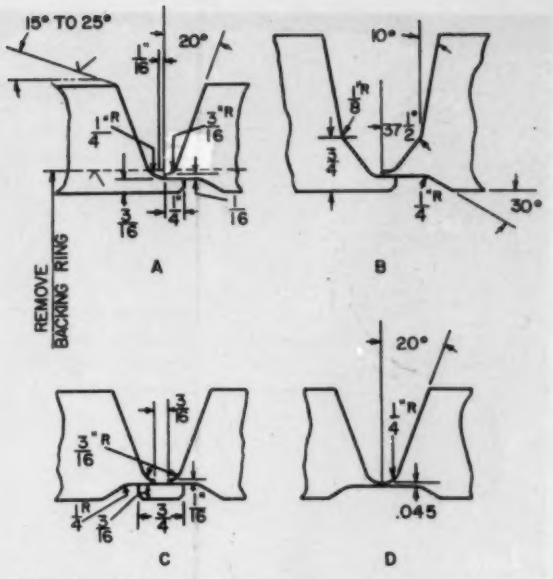


Fig. 11 — Typical weld preparations.

Heat Treatments

The main objectives of subcritical postweld heat treatments are to relieve residual welding stresses and temper hard heat-affected zones. By raising the weld to a high temperature, the yield strength of the material is lowered and a relaxation of the residual stresses is accomplished. The temperature and the time at the temperature are the main variables of a stress-relief cycle. The temperature is more important variable here than the time at temperature.

The Cr-Mo and Cr-Mo-V steels considered herein are given a subcritical stress-relief cycle at 1275 F.

In base materials whose heat-affected zones are subject to secondary hardening, the time at temperature is also an important variable. This time must be long enough to temper the heat-affected zone through its secondary hardening peak to a lower hardness approaching that of the base metal. The Cr-Mo-V steels considered herein are subject to this type of hardening, and are tempered satisfactorily by 8 hr at 1275 F.

The components under consideration have great variation in section sizes. This necessitates carefully planned furnace loading and thermocoupling to assure that all sections receive the proper time-temperature cycle.

Some large turbine castings are processed in the fully annealed condition during repair and planned welding. The advantages of processing castings in this condition are lower as-welded residual stresses, greater crack-propagation resistance and a more complete relief of residual welding stresses by a final above-critical heat treatment.

The annealed chromium-molybdenum castings have an average room temperature yield strength in the range of 45,000 psi, while the normalized and tempered chromium-molybdenum castings have a room temperature yield strength in the range of 70,000 psi. This favors lower as-welded residual stresses for welds made in annealed base metal.

Figure 12 shows typical impact transition curves for the same analysis of annealed and fully heat-treated (normalized and tempered) chromium-molybdenum-vanadium material. The transition temperature is lower for the annealed condition. However, the main difference to be cited is the greater energy-absorption capacity of the annealed material. At 200 F, the normalized and tempered material has a value of 45 ft-lb vs. 74 ft-lb for the annealed material (Note points *A* and *B* in Fig. 12).

This indicates a greater crack-propagation resistance for the annealed condition in this temperature range. It is difficulty to maintain full preheat on some of the large castings (e.g., 80,000-lb large turbine shells), which require considerable handling time to reposition for further welding. Therefore, the annealed material will be less subject to repair weld cracking when it is not possible to maintain a full preheat temperature until a postweld heat-treatment is performed. Experience in the author's plant has generally verified this.

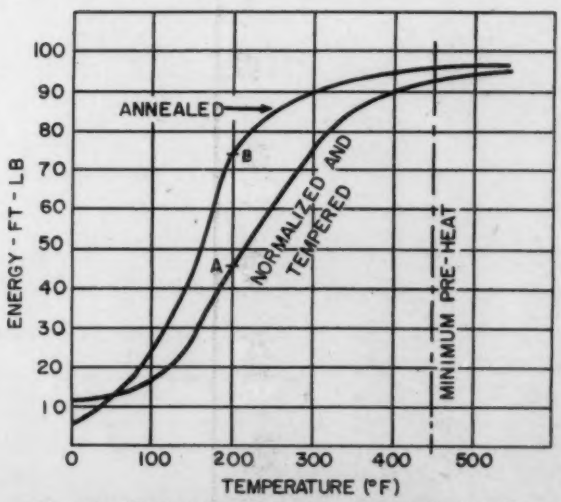


Fig. 12 — Typical impact transition curves for 1¼ Cr; 1 Mo; ¼ V cast material.

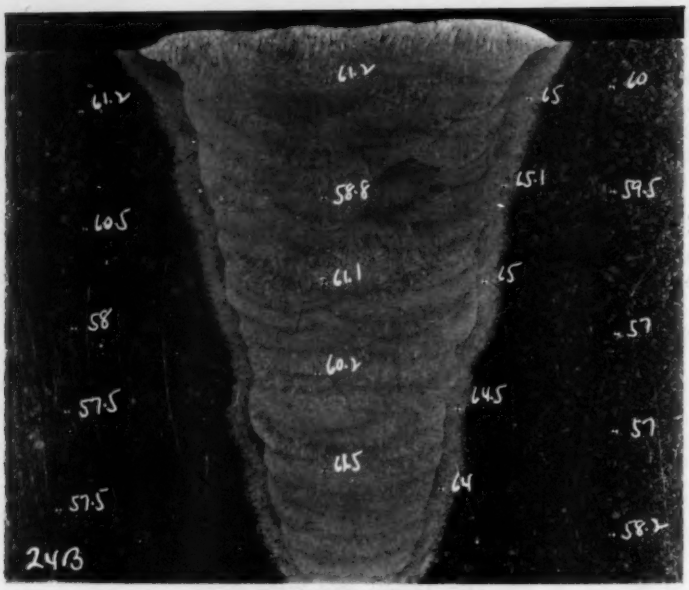


Fig. 13 — Photomicrograph of ½ Cr — 1 Mo Weld in 1¼ Cr — 1 Mo — ¼ V base metal prior normalize and temper heat treatment. Rockwell A hardness survey. 1¼ ×.

The normalize and temper heat treatment, after all planned and repair welding, assures complete relief of residual welding stresses. After a stress-relief cycle, which is a subcritical anneal, the level of remaining stresses approaches the yield strength of the materials at the stress-relieving temperatures. In many cases, these stresses can be as high as the high temperature working stresses. Service experience has indicated sound weld deposits with high-temperature, stress-rupture failures. It is believed that residual welding stresses contributed to these difficulties.

The assurance of complete stress relief by the post-weld normalize and temper heat treatment minimizes cracking. This is of particular importance where heavy section sizes are subjected to high thermal stresses in service. The normalize and temper heat treatment also completely eliminates the heat-affected zone and any "metallurgical notches" associated with it (Figs. 13 and 14).

Distortion and scaling are experienced during the normalize and temper heat treatment. These effects are overcome by the use of tie bars, planned furnace loading and the allowance of extra stock for finish machining.

Intermediate Stress Relief

The removal of casting defects by flame gouging can build up high residual stresses and hard heat-affected zones in one area. By using an intermediate subcritical stress relief, it is possible to minimize these so that subsequent defect removal or repair welding can be done without cracking difficulties.

Also, it has been found beneficial to use an intermediate subcritical stress relief during the repair

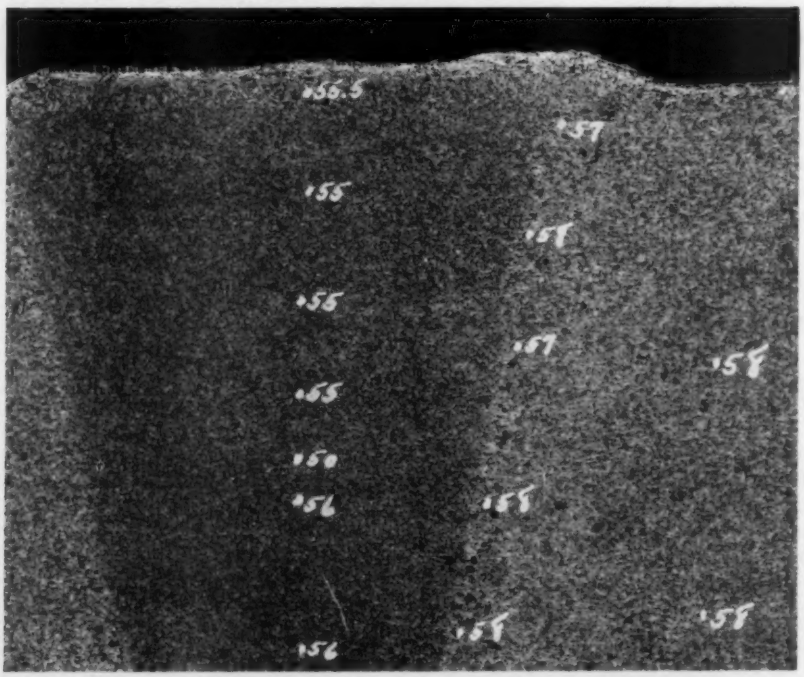


Fig. 14 — Photomicrograph of ½ Cr — 1 Mo weld in 1¼ Cr — 1 Mo — ¼ V base metal after normalize and temper heat treatment. Rockwell A hardness survey. 2 ×.

welding of large areas, in locations where high restraint is encountered.

Electrode Composition

The deposited chemical analysis of the weld metal used for the foregoing casting compositions is shown in Table 1. Only low-hydrogen-type coated electrodes are used for metal-arc welding. However, the majority of the welding is done by the submerged-arc process. At present, work is being completed on the development of chromium-molybdenum-vanadium weld deposits for submerged-arc welding.

WELDING PROCESS TECHNIQUES

The fabrication and repair of large castings are accomplished through the use of the following processes: Manual-arc; manual inert gas metal-arc; mechanized submerged-arc, and manual submerged-arc. These processes are employed so that they complement each other. All submerged-arc welding is done using $\frac{5}{64}$-in. diameter electrode, with currents in the range 300-550 amp and 30-55 arc volts.

It is important to note that a written procedure is prepared for each specific application. Test joints are welded using these procedures, and are evaluated prior to release of the procedure for production use.

All welders are qualified in the processes and procedures which they will use in production. A large portion of planned and repair welding is accomplished using manual submerged-arc welding. For this work, welders have been trained and qualified in both manual metal-arc and manual submerged-arc welding. Qualification procedures are used for manual submerged-arc welding, which includes radiographic evaluation. Assignment of these welders is controlled so that only one welder works on any weld from its beginning to its completion. This method makes one man responsible for the quality of the entire weld.

QUALITY CONTROL

Castings, weldments or forgings used for high pressure, high temperature power generation equipment must be of the highest quality, and the quality must be proved according to written procedures. A difficulty which causes an unplanned outage of a large steam turbine may cost the power producer many thousands of dollars in lost revenue, depending on the size of the unit.

Quality control in the foundry for casting and welding is an effective system for insuring that procedures are established and followed, and a means provided for measuring the result. It gives an up-to-date record of performance and conditions so that quality standards can be maintained continuously. Castings with walls up to 6-in. thick require a greater degree of precaution and closely controlled practices in all phases of manufacture. Therefore, a definite quality control program is imperative. Casting repair welding is a normal, accepted part of the manufacture of high quality pressure-containing castings, and is generally necessary in order to maintain required standards.

Quality control might well be the means of providing information and control, which will aid the most in changing the production of castings from an art to a science.

NONDESTRUCTIVE INSPECTION

No attempt will be made to discuss the various methods of nondestructive inspection. However, all pressure-containing castings used in the manufacture of large steam turbines are inspected thoroughly by more than one method after cleaning, rough machining, planned and repair welding. Certain manufacturer's requirements are more rigid than the normally known codes. Foundries producing these castings must have an intimate knowledge of radiography, magnetic particle inspection, liquid penetrant inspection and ultrasonic inspection. The limits of each method must be thoroughly understood. It is not wise to use one method to the exclusion of all others. It is most satisfactory for foundries to have their own nondestructive inspection facilities.

Specifically, after manufacturing operations noted above are performed, pressure-containing castings are inspected as follows:

1) Visual.
2) 100 per cent magnetic inspection (ferritic materials).
3) 100 per cent liquid penetrant (nonmagnetic materials).
4) Radiographic.
5) Hydrostatic test of pressure components.

Radiographic equipment used by the author's company includes 20 mev betatron, 1,000,000-volt x-ray and 250,000 volt x-ray. Stereoscopic radiographs are used to a considerable extent in determining the exact location of defects with respect to wall surface.

Disposition of Defects

The following evaluations are considered in determining the final disposition to repair or condemn each component:

1) Location and size of defect.
2) Orientation with respect to other defects.
3) Proof that the defect has been completely removed.
4) Service experience.
5) Stress consideration (static and thermal).
6) Limitation of inspection methods.
7) Accessibility for welding.
8) Machining and heat treatment status.
9) Strength of repair weld.
10) Economics.

Shallow surface defects may be faded out and blended into the surrounding surface where location and stress levels permit and where satisfactory appearance can be maintained.

MARKET DEVELOPMENT

The technical growth of foundries in the past ten years has been considerable. Many new processes have been developed, mechanical properties have been upgraded on existing materials and, in general, the quality of castings has improved. In spite of these gains, the total volume of casting business has been dropping for the past seven years. A better under-

standing of the customers' needs, and a closer look at the reason for orders lost to competitors in other fields, may well give the answer.

The steel foundries' main competition in the past has been the fabricators of welded products. Today, forging producers are serious competitors on items up to approximately 5000 lb. Ductile iron also is replacing some carbon steel castings.

Many purchasers are interested in procuring castings which have been completely nondestructive inspected and repair welded to written procedures. Some foundries do repair welding today, but not many are equipped to do machining. The most advantageous time to nondestructive inspect castings is after rough machining. In this state many surface defects have been entirely removed, and excessive repair welding is avoided.

Practically all fabricators and producers of wrought materials have at least rough machining facilities.

Is it not logical that foundries should be deeply interested in the rough machining, nondestructive inspection and repair welding of their own castings in order to deliver the highest quality product to quality-minded customers?

Some castings have been changed to forgings because of poor quality. The fabrications or forging may cost more than the original poor-quality castings, but they are still less costly than the repair-welded and reheat-treated castings. Total casting cost may be doubled because of repair welding. Quality castings can regain lost business.

The producers of fabrications have gone to the manufacturers and looked over castings to see where they could produce a comparable part with improved

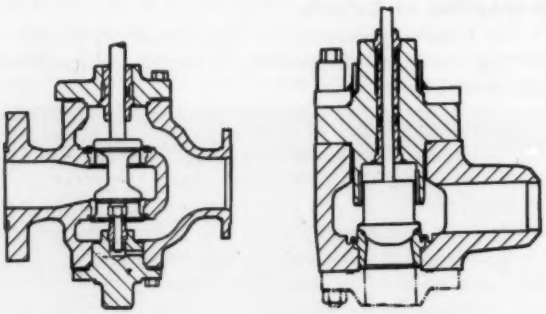


Fig. 15 — Balanced valve design to simplify casting.

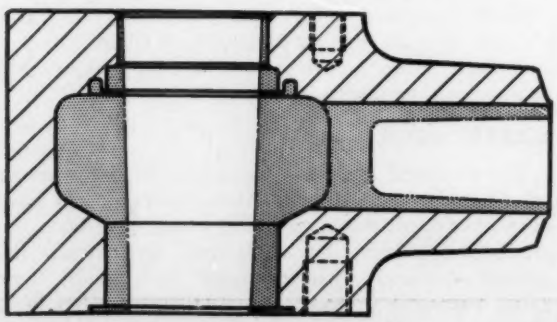


Fig. 16 — Redesigned cast valve casing.

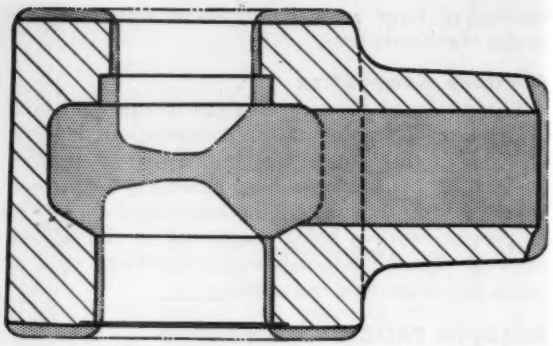


Fig. 17 — Forged valve casing.

quality and lower weight, and sell a satisfactory product at a lower price. Considerable engineering effort goes into each project, with particular attention and study given to all phases of manufacture, including welding, heat treating and nondestructive inspection. Whenever a part becomes difficult or impossible to make with existing equipment, a casting, and generally a complex one, is put in the weldment.

One can only admire any group which is vigorous and far-sighted enough to take advantage of all present-day engineering and manufacturing know-how, and produce acceptable quality products at lower costs. Careful selection of products and intensive engineering effort can produce astounding cost improvements.

New Business Possibilities

Engineering-wise foundries have the same opportunities as the fabricators of welded products, and the producers of wrought materials, to secure new business. They should recommend changes and develop designs to favor the proper use of castings made in the most favorable way. There is nothing wrong with using wrought parts in these predominant cast weldments when the design is improved. Castings made to the generally simple shapes of forgings offer few foundry problems. Concessions are made many times in fabrication designs. These same concessions may be given foundries for casting redesigns if they are requested.

Foundries are losing a lot of their potential business if they produce only the large or complex castings for components which cannot be made by other methods. A good look at the foundry competitors' products will reveal many items which can be produced by casting at equal or better quality and lower costs. It will require some digging to find these items.

Figure 15 shows a balanced valve assembly which was redesigned for both improvement in operation and simplification of the body casting. The body was cast with excess stock on the inside in order to avoid centerline shrink, and directional solidification was provided for as shown in Fig. 16. The castings were sound and no repair welding was required. This rough casting weighed 250 lb.

The forging group was able to produce a forging, as shown in Fig. 17, rough machine it to within $\frac{1}{8}$-in. of finished dimensions, and sell it for less than the rough casting. The foundry should be able

to cast and rough machine this casting for at least the same price. If the foundry does not have machining facilities, why not make price concessions when it is definitely known the customer is machining away stock that was added to insure sound casting? The purchaser does not pay directly for the risers or pads burned off in the foundry.

Many dies for forgings are made from castings where the cavity is practically cast to size with a minimum of machine work necessary. The reduction in cost is considerable.

What has been done by foundries to reduce pattern costs for large castings in the past five years? While some foundries do not make patterns, their costs certainly add to the final cost of castings. Since many patterns for these large castings are used only a few times, what different construction could be used to produce less costly patterns?

Risers are burned off by hand in many foundries, leaving excess stock and/or undercutting. These two casting faults cost the customer a considerable amount of time and money. Machine guided burning torches should be more widely used, and other methods of riser removal, such as sawing, should be considered.

Important Considerations

Orders should be accepted in the casting range that can be made economically. It is important to understand completely the material specification and all test requirements. The intended use, special requirements or limitations must be thoroughly understood.

It is wrong to assume that good castings have been made from the pattern received. You may be given an order because some other foundry has been unable to produce the needed quality. The removal of all telltale evidence of how the casting was gated and risered when returning patterns is still practiced. Orders should be rejected unless you have the knowledge that a satisfactory casting or cast weldment can be made at a fair profit. The first casting of a given type should be evaluated before completing any order.

More scale models should be made and examined, in order to eliminate hot spots and aid in the planning of feeding and gating. Most drawings do not show junctions or section changes clearly. Sample castings cannot be made and sectioned to prove a design when you are producing castings up to 85,000 lb. However, plaster casts can be used effectively in checking many of the smaller parts.

When mistakes are made, do not attempt to repair them in your own manner. Instead, tell your customer. He may be able to save you a lot of work.

The recently established "Value Analysis" groups in many companies offer a new approach to customers. However, the value of discussing design problems with designers first hand is still of utmost importance.

Designers have access to considerably more information in the welding and fabrication fields than in the casting field. There is a general belief among many product designers that fabrications are always less costly because no pattern is involved. The number of parts to be produced often is forgotten. More information on castings should be available to designers.

A vigorous, large scale program is necessary to educate the present-day product designers and users of castings in the current foundry capabilities and the limitations of all foundry processes. The ability to prove the quality of castings by nondestructive inspection in the foundry should be provided and publicized to a greater degree.

CONCLUSIONS

Foundries can contribute much more to proper design of quality castings by a careful evaluation of the customers' needs and by the application of modern foundry engineering practices. Complex components of higher quality can be made by casting better quality individual parts which have been made in the best suited manner and assembly welded.

Sound welds can be made in heavy-walled, low alloy steel castings if closely controlled procedures are followed. Quality control and proof of quality is necessary to guarantee minimum service requirements.

There is a good potential market for high grade castings, but an educational program, to better inform designers and users of castings of the advances made in the casting industry and the present-day status, is needed to fully exploit this potential. In all events, aggressive efforts should be maintained to minimize the possibility of losing even more casting work. Foundries willing to devote engineering effort today to improve and understand the functions of their castings will be the leaders and profit makers of tomorrow.

ACKNOWLEDGMENTS

The author wishes to express his gratitude to N. A. Chapin, C. H. Soldan and other members of the Large Steam Turbine-Generator Dept., General Electric Co., for their assistance in the preparation of this paper.

NOTE: When this paper was written, the author included information and expressed opinions believed to be correct and reliable. Because of the constant advance of technical knowledge, the widely differing conditions of possible specific application and the possibility of misapplication, any application of the contents of this paper must be at the sole discretion and responsibility of the user.

ALUMINUM DIE CASTING METAL COST

its practical and analytical evaluation

By George H. Found and John Lapin

ABSTRACT

Fast developing technological improvements in die casting processes, the expansion of markets and competitive pressures have reduced unit cost greatly during the past five years. Resulting from this, the cost of metal has become a dominant factor in determining whether profits exist in normally resourceful die casting operations. Elementary analysis, supported by the experience of job shop operations, is used to demonstrate that exacting control of metal cost and metal handling are the most important factors in the success of a competitive, modern die casting operation where performance, quality, profit-making motives and competitive pressures must be met on an optimum basis.

A method of establishing the value of the separate factors contributing to the metal cost both in an overall operation and in an individual casting is discussed. A practical review is given of practices for minimizing metal costs in light of the foregoing.

INTRODUCTION

Literature gives extensive coverage of the various phases of die casting aluminum. Despite this, new developments both in the laboratories and on production floors are taking place in this expanding field more rapidly than the literature can cover. As a result, the literature does not emphasize sufficiently the relative effects of these developments on the various elements, particularly on the metal costs.

The scope of this paper is to consider briefly, from the standpoint of practical experience, the elements which appear to be of greatest day-to-day importance, their relative importance to each other and the way this relationship has varied during the evolution of recent years.

The most important cost item, not only from the magnitude standpoint but also from its dominating position in determining profit or loss position of the operation, has recently become the metal cost. The general improvement in speed of casting, the higher injection and locking pressures, improved die construction and changes in metal handling methods, have combined to cause this pronounced emphasis.

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One might easily forecast that metal cost is the big factor, but the degree to which this is so, especially in such a fast-changing industry, is deserving of constant review. The cost structure of castings in the die casting industry five years ago was sufficiently different to have caused direct labor and other elements to have shared the attention with metal costs. Advancement in techniques and requirements has since changed this relationship so that small increments in metal cost supersede all others in importance if it is assumed that foundry performance is at least average in other respects.

RESULTS OF EXPERIENCE

The specific factors accounting for the present emphasis are the increase in casting rates, the increase in casting and poured weights and the reduction in the other cost elements due to improvements in mechanization and handling.

It is not the object of this paper to supply compact answers to the metal cost problems. An analysis will be given which will be suggestive for appropriate treatment of the variables as they may apply to individual plants and operations.

In relatively complicated processes involving work being applied to a material flowing through a plant, the costs of the material initially are superseded by the attention given to improving the process, reducing labor costs, increasing the production rates and the quality and flexibility of the process. This has happened in the die casting industry. Despite this sequence, however, metal costs still have been a significant part of finished part cost.

With process improvements, and attendant improvement in cost, quality, production rates and breadth of market in the end product, the metal cost has evolved into the most important single cost factor. In many die casting operations, the metal cost has become so dominating that all other costs combined are relatively small when compared to the cost of metal.

The metal cost increases in imporance with respect to the total factory cost as production rates and individual casting weights increase. Thus, the current trend in the industry (higher rates and larger castings) has resulted in an increase in the ratio of metal

cost to the factory cost, and, consequently, in relation to the casting selling price.

Continually greater effort must be exerted, first in understanding the elements which contribute to the metal cost, and second, in the means used to control them.

DEVELOPING THE IMPORTANCE OF METAL COST GRAPHICALLY

It is interesting to note how the amount of metal handled per hour drastically increases with size and rate of production of castings in a normal die casting operation.

In Table 1 some typical die casting rates are shown for varying sizes of castings. These rates are compiled from actual operations familiar to the authors. Any die casting operation would show a similar relationship, with the hourly rates displaced upward or downward when plotted against the casting weight as the abscissa.

A general increase in casting rates, such as might result from automatic ladling or improved temperature control of the dies, will result in an upward displacement of the curve without substantially changing its slope.

Figure 1 graphically shows the relation between casting weight and production rate in pounds per hour.

The relative importance of metal cost to total cost can be visualized as pounds per hour increase due to either higher casting rates or larger casting size.

If the metal cost per casting is expressed as per cent of the total factory cost (disregarding at this point metal losses and using a uniform factory cost rate per machine hour), Fig. 2 shows the rise in metal cost with increasing production rates. Here the factory cost is the sum of the metal cost, direct labor and overhead. The ratio, metal cost: factory cost, may be considered to be approximately equal to the metal cost as per cent of selling price (for purposes of illustration, a uniform factory cost of $30.00/machine hr, and an ingot price of $0.25/lb were used). The values derived are shown in Table 2.

If the cost of the metal is deducted from the factory cost, and the resulting cost is termed manufacturing cost, then the ratio metal cost/manufacturing cost plotted against casting rates becomes a much steeper curve than that shown in Fig. 2. However, the curve is perhaps not truly reflective of the realistic relation that exists.

From the above considerations it becomes significant, in a quantitative way, that as the industry

TABLE 1 — TYPICAL CASTING RATES — ALUMINUM DIE CASTING

Casting weight, lb	No. of cavities	Shots/hr	Castings/hr	lb/hr
0.1	8	120	960	96
0.5	4	90	360	180
1.0	3	75	225	225
2.0	2	65	130	260
5.0	1	65	65	325
10.0	1	45	45	450
20.0	1	30	30	600
30.0	1	22	22	660

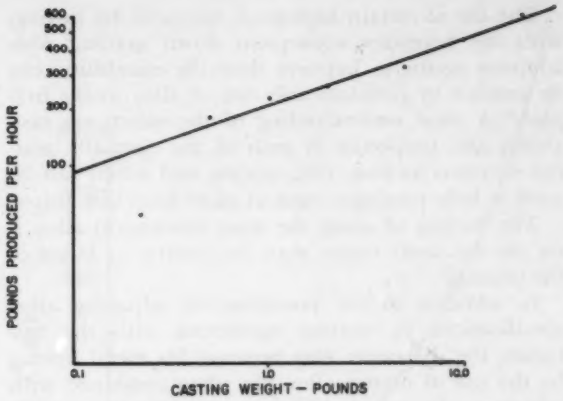


Fig. 1 — Relationship of casting weight to production in lb/hr.

proceeds in its inevitable direction of higher rates and heavier castings, more attention must be focused on the cost of the metal. This can best be done by separating the metal cost into several principal components. These are:

a) Ingot cost.
b) Designed casting weight.
c) Melt loss.
d) Die, machine and operator factors affecting die blow.

METAL COST SAVINGS CONNECTED WITH INGOT COSTS

Since the market price of metal affects all suppliers of die castings alike, it may appear that little can be done by the individual die caster to improve his metal cost/selling price ratio. An all-out effort to buy ingot as cheaply as possible is not an adequate conception for treating the problem.

TABLE 2 — RELATION BETWEEN CASTING RATE AND METAL COST

Production Rate, lb/hr	Metal Cost, % of Total Costs
96	44.3
180	60.0
225	65.2
260	69.0
325	73.0
450	78.9
600	83.2
660	84.8

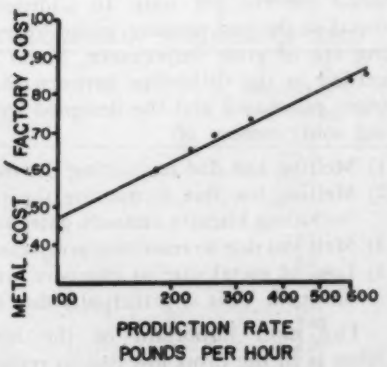


Fig. 2 — Relationship of metal content per unit of total cost to production rate.

The use of certain high-grade alloys in die casting, with the necessary subsequent down grading with addition agents to improve their die castability, can be avoided by judicious selection of alloy in the first place. A clear understanding of the effects on castability and properties of such of the normally present elements as iron, zinc, silicon, and others can be used to help purchase ingot at more favorable prices.

The burden of using the most economical alloy is on the die caster rather than the smelter or buyer of die castings.

In addition to the possibility of adjusting alloy specifications by mutual agreement with the customer, the die caster may improve his metal buying by the use of master alloy pig when combined with adequate alloying facilities and means for rapidly determining chemical composition. Freight savings by shipping castings out and returning ingot by the same truck offer cost savings in many instances.

Metallurgical inspection, including controlled checks for dross loss and nonmetallic content, should be routine.

METAL COST SAVINGS CONNECTED WITH DESIGNED CASTING WEIGHT

Close cooperation between the die caster and the product engineering department of the casting user has been the subject of extensive attention. It has also been pointed out as being an activity which the die casting business especially needs. In several recent meetings of the Society of Die Casting Engineers this was the principal subject. Here again it is the die caster on whom the large share of the burden falls. A carefully made casting drawing, spelling out wall thicknesses, drafts, finishes, tolerances, weight savers and possibilities of some revision in design, makes it necessary for the caster to understand the properties of the material he is working with. Also, it requires his familiarity with the end use.

Such things as minimum wall thickness, minimum finish allowance and no finish on holes which are tapped, require close understanding between supplier and customer. Closely held casting dimensions permit the use of minimum finish, but require good die construction and dependable machine operation.

METAL COST SAVINGS CONNECTED WITH METAL LOSS

The aim of the foregoing points is to reduce the metal content per unit. In addition to conserving metal in the end product, metal losses during processing are of great importance. Metal loss, if it is regarded as the difference between the weight of the ingot purchased and the designed weight of the casting sold, consists of:

1) Melting loss due to melting the ingot.
2) Melting loss due to melting the runaround scrap, including biscuits, runners, gates and finish.
3) Melt loss due to remelting scrap castings.
4) Loss of metal due to excessive weight of castings shipped. This is principally due to die blow.

The most important of the foregoing melting losses is in the dross loss due to remelting runaround scrap. Regardless of the method of melting, and assuming reasonably high quality ingot, the loss due to melting the required amount of ingot is secondary to the sum of the losses due to melting the unit scrap.

For this reason the unit yield of the net casting weight divided by the total weight cast, is of utmost importance.

In addition to making it necessary to have extra melting and metal handling capacity, there is a large overall effect of yield on melt loss due to rehandling of scrap. This is shown in Fig. 3.

The points on the curve in Fig. 3 can be obtained by using the following derivations.

If

L_m = Melt loss in pounds.
I = Ingot used in pounds.
W = Net Casting weight in pounds.

Then from a simple material balance,

$$L_m = I - W \qquad (1)$$

Since L_m is the sum of the unit melt losses incurred in melting the ingot and the runaround scrap, and if

L_i = Unit ingot melt loss (pounds per pound of ingot melted)
L_s = Unit scrap melt loss
S = Pounds of scrap melted

Then,

$$L_m = IL_i + SL_s \qquad (2)$$

Combining (1) and (2),

$$L_m = (W + L_m)L_i + SL_s \qquad (3)$$

The remelt scrap is equal to the difference between the total weight cast and the net casting weight. The yield is defined as the ratio net casting weight to total weight cast per casting.

If

Y = Yield.
C = Gross cast weight.

Then

$$S = C - W \qquad (4)$$

and

$$Y = \frac{W}{C} \text{ or } C = \frac{W}{Y} \qquad (5)$$

If (4) and (5) are combined,

$$S = \frac{W}{Y} - W \qquad (6)$$

Substituting (6) in (3)

$$L_m = (W + L_m)\, L_i + \left(\frac{W}{Y} - W\right) L_s \qquad (7)$$

From relationship (7) the melt loss in pounds per casting can be established (This can be expressed in pounds per pound of net casting weight by dividing L_m by the net casting weight.) Assuming a yield of 62.5 per cent, a unit ingot melt loss of 2 per cent and a unit scrap melt loss of 7 per cent, a typical calculation is shown below, substituting in (7)*

*R. Scmitt, "Die Metallverluste in der Spritz — und Pressgiesserei," *Metallwirtschaft, 22,* no. 18-20.

$$L_m = (1.0 + L_m)\,0.02 + \left(\frac{1.0}{0.625} - 1.0\right) 0.07.$$

$L_m = 0.0632$ or 6.32%.

L_m for other values of Y is shown in Table 3.

A family of curves similar to that shown in Fig. 3 can be constructed for any values for L_i and L_s which might apply in a particular plant operation.

The values L_i and L_s in themselves vary widely with the method used for melting and with the quality of ingot used. The character of the runaround scrap affects the melt loss quite drastically. Massive, dense scrap may have a unit melt loss only two to 2½ times more than good quality ingot, but thin, flaky and otherwise high surface area scrap may have a melt loss ten times the loss for melting ingot.

Expression (7) does not take into account the following three factors:

a) Casting efficiency.
b) Excess casting weight due to die blow.
c) Added remelt scrap weight due to die blow with a reduction in yield.

Other items not considered in expression (7) are processing losses such as machining, grinding or other finishing losses which are strictly additive.

To express the metal loss completely, taking the above factors into consideration, the melt loss L_m as determined by expression (7) must be divided by the casting efficiency, and the weight due to die blow must be added. The unit scrap weight increases by the amount of blow. Ingot weight also increases by the same amount.

If

B = weight due to die blow;
E = casting efficiency as the ratio good pieces/total pieces cast;
L = Total metal loss;

then an approximation for the total metal loss may be expressed thus,

$$L = \frac{L_m}{E} + B \qquad (8)$$

This approximation does not take into account the additional melt loss on the rejected castings, which are now heavier than they should be, nor the additional melt loss on the heavier unit runaround scrap.

In the example casting used to illustrate the application of expression (7), if the projected area of the casting is 50 in.² and the die blow is 0.010 in., then (using 0.1 lb/in.³ as the density of aluminum),

$$B = (50)\,(0.010)\,(0.1) = 0.05 \text{ lb or } 5\%;$$

and if E = 0.80 (80%) then;

$$L = \frac{0.0632}{0.80} + 0.05 \text{ or } 0.1291\ (12.91\%).$$

The drastic effect of both die blow and casting efficiency is illustrated by this example.

More exactly, since 20 per cent of the castings were remelted, (0.05) (0.20) or 0.01 lb must be added to S in determining the melt loss. Furthermore, if the 5 per cent weight added to the casting is also added to the original runaround scrap, then S becomes 1.05S. The melt loss for remelting the runaround scrap and rejected castings becomes $(1.05S + 0.01)L_s$, and can be so used in expressions (2) and (7). The added loss is on the order of 0.3 per cent.

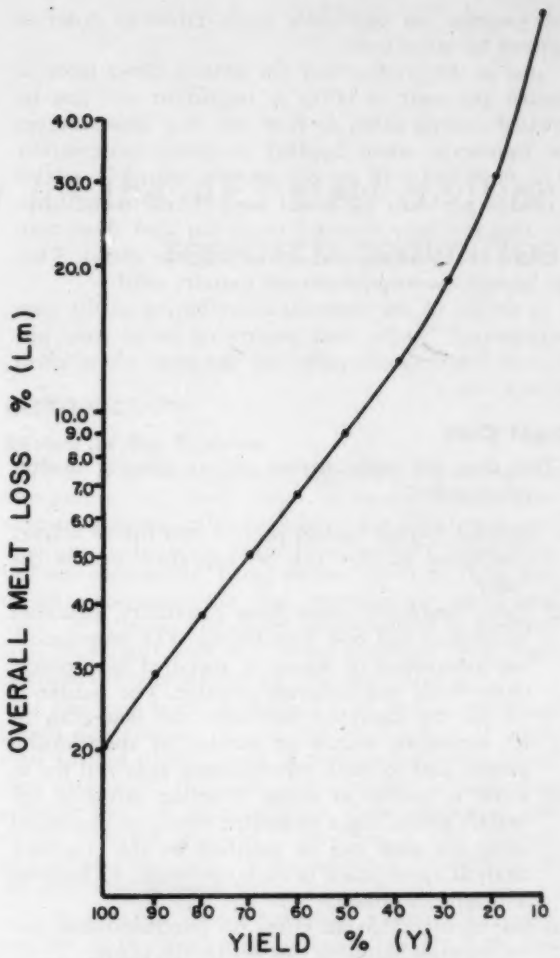


Fig. 3 — Melt loss as a function of yield.

SUMMARY

It has been demonstrated how dominating metal cost components have become in molten die casting practice. The following thoughts represent an orderly attempt at reviewing from a practical standpoint the specific factors on which the producer of

TABLE 3 — MELT LOSS AS A FUNCTION OF YIELD

Yield Net casting wt. Total Wt. cast/casting,%	Melt loss as % net casting wt.
5	133.0
10	66.4
20	30.6
30	18.7
40	12.7
50	9.2
60	6.8
70	5.1
80	3.83
90	2.94
100	2.04

die castings can profitably apply effort in order to control his metal costs.

Just as the productivity per unit of direct labor or return per unit of effort is important and has increased casting rates, so now can this same concept be important when applied to metal conservation. The objective is to get the greatest return in salable castings per unit of metal cost. These metal costs include not only material costs, but also those costs related to handling and conserving the metal. They go beyond the scope of normal foundry yield.

A review of the elements contributing to the costs represented by the four sources of metal costs previously listed should point out the areas where effort is required.

Ingot Cost

(This does not only involve simply beating smelter prices down)

a) *Sensible buying* against present and future market conditions as they can be appraised by the die caster.

b) *Using optimum alloys* from castability, customer acceptance and cost standpoints. The unquestioning acceptance of alloys as specified is a potentially costly and lethargic practice. The burden is on the die caster to determine the best alloy or its variations within or outside of specification ranges, and to work out customer approval for it.

c) *Form of receipt of metal.* Whether prime or secondary alloys, ingot or molten metal, or purchased scrap are used can be justified by the economic analysis appropriate to each operation, its location and type of product.

d) *Size of ingot* has an effect on purchased cost and on material handling cost within the plant.

e) *Freight* costs can be reduced frequently by carrying ingot instead of "dead heading" on return trips from delivering castings.

f) *Ingot quality* is a large factor where high performance, good machinability, close quality specifications or complex castings are involved. The presence of certain subtle but insidious factors, such as segregated intermetallic compounds and dross, can cause increasing metal costs even before foundry yields signal such rising costs.

Casting Weight

a) *Weight control as a concept* must be sponsored and encouraged by the die caster. Coordination between the product engineer, the die caster and the subsequent machining or manufacturing engineer is the starting point for exercising weight saving practices.

b) *Minimum wall thicknesses* must be used. Actual functional service requirements must be known, and excessive tolerances on wall thicknesses must be avoided.

c) *Die blow* or parting line flash must be minimized.

d) *Holes and weight savers* must be cored into the design. These not only improve castability, but frequently save subsequent operations.

e) *Match alloy to castability problems.* Frequently it has been possible to save metal in large amounts by the 0.010-in. or more reduction in walls made possible by increasing silicon content within the specification range.

f) *Die design* as it affects die blow, slide blow, parting line flash and washing out is important in weight control.

g) *Dimensional control* in the die makes thinner finishes for machining adequate. Reduced machining has the bonus advantage of not only saving metal, but of frequently eliminating flange facing and drilling before tapping. Holes and walls on location result in thinner walls and less finish.

Melt Loss

a) *The selection of the method for melting* and the inherent provisions for insulating the metal from an oxidizing environment have as much as a five-fold variation on losses. The type of temperature and atmosphere control, whether or not immersion melting is involved, whether gas or electric reverberatory melting is chosen, whether gas or resistance open pot melting is indicated or whether high or low frequency induction melting are chosen, are not only independently important, but must also be fitted into the composite aspects of the whole operation for best chance of reducing melt losses.

b) *Fluxing and refining techniques* are vital to conservation of metal.

c) *Casting efficiency, foundry yield, and flash* are particularly sensitive in their influence on melt loss. The volume of runaround scrap generated from these sources has considerable leverage on melt loss.

Die, Machine and Operator Factors Affecting Metal Conservation

The principal evidence of losses due to these influences is flashing. Miter and shut-off design which is accessible for die cleaning, lock design, drafts, cooling system, ejection and releasing features in the die design, control the tightness of locking and the losses due to metal escaping the cavity. In the machine, high locking pressures, good alignment techniques and practices and good maintenance are important for conserving metal. Operator vigilance in properly treating the above, as well as in keeping dies free of dirt, solder, abuse and uneven heating, show up in lower metal costs.

CONCLUSIONS

Since the die casting process for aluminum has developed into essentially a rapid transfer of metal through systems, the mechanization of which are in various stages of perfection, the most important economic factor now revolves around the cost of the working material flowing in the system.

The various aspects considered here can be best used if a detailed flow sheet is constructed for any operation, and the details peculiar to that operation are included. This and analyses built around it will serve as the most basic means for understanding and controlling the major cost of the operation.

DEOXIDATION DEFECTS IN STEEL CASTINGS

final report for the AFS steel division research committee

By R. A. Flinn and L. H. Van Vlack

ABSTRACT

Nonmetallic macroinclusions, commonly described as "snotters," are a frequent cause of scrap or lead to inferior machinability of the cope surface of steel castings (Mechanically entrapped sand and scabs are not included in this designation).

The three objectives of this investigation were to:

1) Identify these inclusions.
2) Trace their origin in foundry practice.
3) Suggest methods of elimination.

The results may be summarized under the same divisions:

Nature of Inclusions. In steel deoxidized with aluminum, the cope inclusions contain the mineral species corundum (Al_2O_3) surrounded by reaction products formed by later reactions between the corundum, the mold material, the metal and the atmosphere.

Origin of Inclusions. Detailed explanation is given to prove that the essential feature of the defect, the corundum, can only originate in the metal as a result of deoxidation with aluminum.

Methods of Elimination. Mechanical techniques of settling, skimming and trapping, as well as the use of complex deoxidants to form other deoxidation products, may offer solutions.

PREFACE

Considerable effort has been made in the last four years to identify the nonmetallic cope side defects found on steel castings. Since the results of this work throughout the cast metals industry were largely inconclusive, a fresh approach seemed necessary. Accordingly, the AFS Steel Research Committee sponsored work at the University of Michigan to identify these inclusions and to determine their genesis.

The results of this work are presented in the following paper. The committee feels that the steel foundryman should particularly note:

1) The positive identification of the defect.
2) The positive determination of the genesis of the defect.
3) The importance of the ceramist's polarizing microscope in phase identification of nonmetallic inclusions.
4) The necessity of retaining the surface sinter layer on castings in order to permit positive defect identification.

R. A. FLINN and L. H. VAN VLACK are associated with the University of Michigan, Ann Arbor.

INTRODUCTION

Review of the Problem

There is certainly general agreement regarding the importance of clean cope surfaces. Nonmetallic macroinclusions, called "snotters," are significant contributors to foundry scrap. Possibly more important, but less spectacular, is the serious effect of these materials upon tool life and in reducing endurance limit in fatigue. The latter consideration is particularly important with the recent emphasis upon high strength steel castings for aircraft applications.

It has long been recognized that proper molding, gating and risering lead to the elimination of other cope surface defects, such as mechanically entrapped sand and scabs. However, the "snotter" is somewhat reduced but not eliminated by these otherwise successful foundry practices. Therefore, it was selected as a subject for study by the Research Committee of the AFS Steel Division.*

Although this defect has been a source of great concern to production personnel, relatively little has been published concerning the problem. In most cases, castings are cleaned before inspection and it is then quite difficult to distinguish between the "snotter," mechanically entrapped sand, and even in some cases certain scabs, washes and drops. Sanders[1] has published a good description of the "snotter" defect, and indicated that a low melting eutectic slag from deoxidation might be the cause. He indicated further that this defect was still present in castings poured in graphite molds, thereby eliminating the variable of loose sand.

The following analyses of inclusions from castings were given:[1]

	From Sand Molds	From Graphite Molds
SiO_2, %	84	45
Al_2O_3, %	5	18
FeO, %	4.5	—
MnO, %	4.6	14
CaO, %	1	up to 18

Other discussions are generally fragmentary, but most shop sessions indicate the seriousness of the problem.

Based upon the available literature and many help-

*Present membership: A. J. Kiesler, Ex Officio, W. O. Igelman, C. H. Lorig, J. A. Rassenfoss, G. J. Vingas, S. C. Massari, Ex Officio and G. A. Colligan, Chairman.

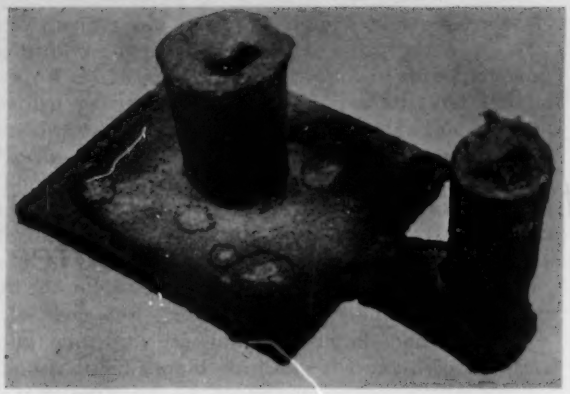


Fig. 1 — Small snotter defect (Source 1). The only evidence before cleaning is the sand discoloration in the circled areas. Plate is 5⅝ in. x 6½ in.

ful personal communications, the objectives of this research were established as:

1) To identify the "snotter" defect in commercially produced sand castings.
2) Based upon the information of Sanders,[1] and a knowledge of the commercial production practices involved, to determine the origin of the nonmetallic inclusions.
3) To suggest means of reduction or elimination.

PROCEDURE

Samples Selection

Samples of "snotters" were provided through the co-operation of the four producers cited in the acknowledgment. The adhering sand was carefully protected during shipment so that the inter-relation between sand and casting would be evident. Details of manufacture were given with each sample, and in all cases aluminum deoxidation and silica sand molds were used. Samples of ladle riser scum were also submitted for examination to determine any inter-relation with the defect.

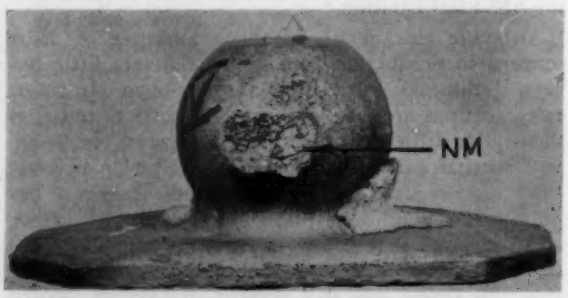


Fig. 2 — Large snotter defect (Source 2). Ball size is 4½ in. in diameter. Nonmetallic material (NM) has been shifted down to expose underlying metal surface.



Fig. 3 — Snotter defect of intermediate size (Source 3).

Examination Method

After receipt, the castings were photographed to illustrate the macroscopic appearance of the defects and then typical sections were cut, impregnated in thermal setting resin, mounted in thermoplastic resin and polished metallographically. The techniques are further described in references.[3,4] In addition, for optical measurement of birefringence and indices of refraction, powdered samples were studied with a petrographic microscope.

DISCUSSION

The results of the investigation may be discussed conveniently in the same order as the objectives, namely, the nature of the defect, its origin and suggested procedures for elimination.

Nature of Snotter Defect

Macroscopic appearance. The photographs of the samples submitted are reproduced in Figs. 1 to 4, inclusive. When a small defect is encountered, the only evidence is given by local discoloration of the adhering sand, Fig. 1. When the inclusions are larger, then a depression in the surface of the metal is encountered, Fig. 2. However, in all cases sand adheres to the inclusion as if a sintering or fluxing reaction is occurring.

Microscopic examination. The key feature of all the samples is the occurrence of the mineral corundum. It should be emphasized that the structures illustrated are not merely evidence of Al_2O_3, but of a particular crystalline species. The distinguishing characteristics are a hexagonal structure, high hardness and the optical characteristics of a uniaxial negative crystal of low birefringence with an index of refraction of $w = 1.76$ (the optical properties were determined by polarized transmitted light).

The corundum is visible as gray basal plates with



Fig. 4 — Snotter defect of intermediate size (Source 4).

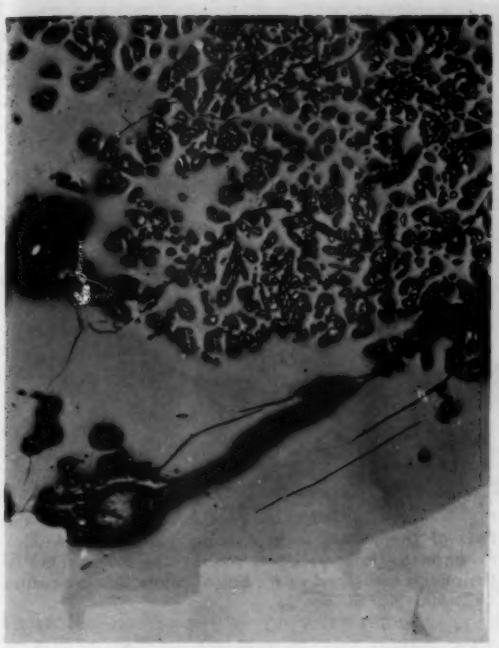


Fig. 5a — Cross-section through metal (bottom) and snotter (top). At top, corundum crystals in a glassy matrix are encountered followed by glass, and then the typical cast steel structure (Source 4). 100×.



Fig. 5b — Typical crystal outlines of the corundum in the glassy matrix are evident in this micrograph (Source 4). 500 ×.

the polishing characteristics of a hard phase in Figs. 5-9.

Other phases encountered in the inclusions may be summarized:

Glassy matrix (siliceous) – this noncrystalline substance surrounds the corundum crystals.

Quartz (SiO_2) – this is the form of most of the original sand grains.

Cristobalite (SiO_2) – sand reacts to form a glass from which cristobalite crystallizes. This is the stable form of silica in the temperature range 2680-3100 F; retained as it is cooled to ambient temperatures.

Magnetite (Fe_3O_4) is encountered in highly oxidized regions of the casting surface.[3]

Wustite (FeO) is present under less oxidizing conditions.

Fayalite (Fe_2SiO_4) is formed by crystallization during cooling from the liquid silicate glass formed at reaction temperatures.

Hercynite ($FeO{\cdot}Al_2O_3$) is a reaction product of iron oxide and corundum.

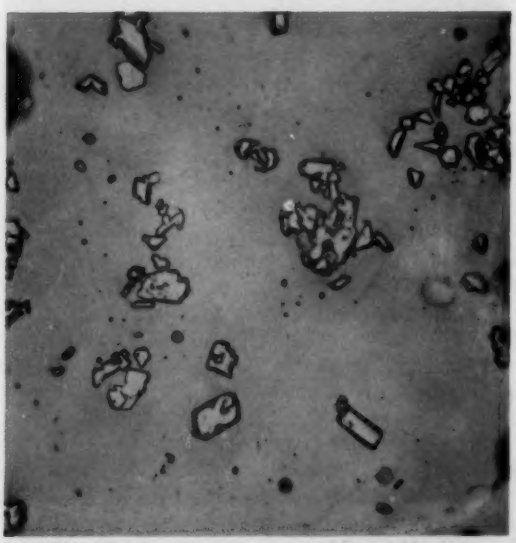


Fig. 6 — Detailed structure of snotter (Source 2) showing same structures as Fig. 5b, but greater solution of corundum by glass has reduced the amount and altered crystal outlines. 500 ×.



Fig. 7 — Detailed structure of snotter (Source 3) similar to Fig. 6, showing corundum. Metal (white) constitutes lower portion of photomicrograph. 500 ×.

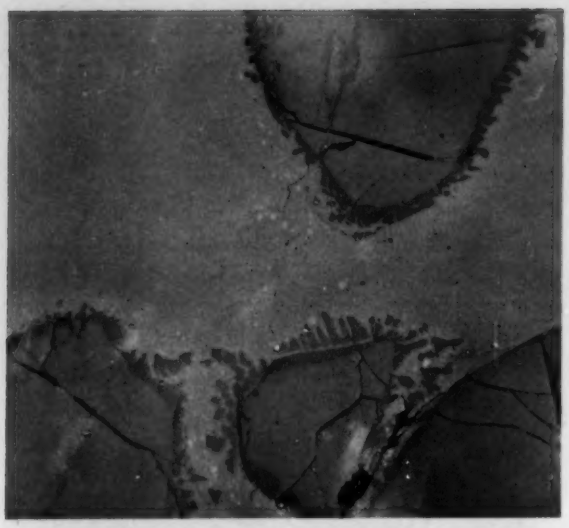


Fig. 8 — Sand grains with cristobalite fringes in glassy matrix (Source 3). Glass consists of SiO_2 dissolved corundum (Al_2O_3) and FeO. 500 ×.

The individual samples are discussed in the captions of the figures. Because of the minerals encountered, the variation in chemical analysis reported by Sanders can be better understood.

In summarizing, it can be pointed out that all the macroinclusions contained corundum accompanied by varying amounts of other minerals.

Origin of Ceramic Macroinclusions

In addition to providing identification of the defects, the petrographic analysis provides interesting information on the source of the inclusions. Corundum is not present in the molding material nor in the ladle refractories. The only logical origin is reaction of the aluminum added for deoxidation with oxygen dissolved in the melt. This is commonly encountered in steel ingots deoxidized with aluminum,[2,5] and many examples of corundum formation have been noted.

Therefore, the following sequence is proposed:

1) Reactions in the liquid metal, while in the ladle, and passing through the mold.

$$2\,\underline{Al} + 3\,\underline{O} \rightarrow Al_2O_3 \text{ (corundum)}$$

Some reaction of Al with FeO may also occur at the ladle surface.

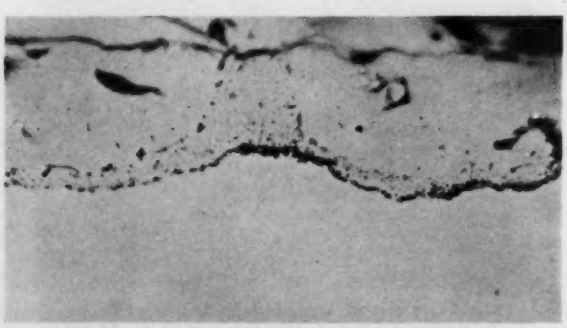


Fig. 10a — Iron oxide surface layer encountered in casting surface from all sources. 500 × (Source 3).



Fig. 9 — The corundum crystals have reacted with the FeO of the matrix to give a hercynite ($FeO{\cdot}Al_2O_3$) overgrowth (arrow). Upon cooling the excess FeO in the matrix combined with silica to crystallize as fayalite (F). 500 × (Source 2).

2) The surface of the corundum reacts with iron oxide forming a rim of hercynite.

$$Al_2O_3 + FeO \rightarrow FeO{\cdot}Al_2O_3$$

This reaction can take place in the ladle, metal stream or at the cope surface (Fig. 9).

3) Corundum reacts at the cope surface with silica, and iron oxide, to form a glass in addition to the hercynite reaction (Fig. 6).

4) Minerals crystallize from the glass upon cooling (Figs. 8, 9, 10b and 11).

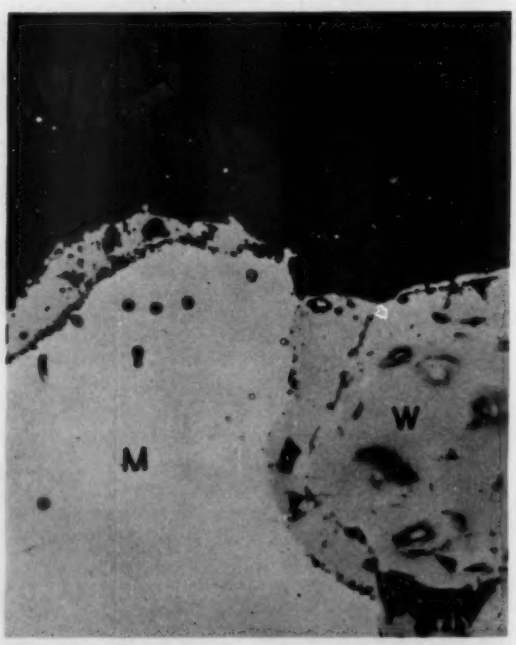


Fig. 10b — Reaction of iron oxide with silica sands. Metal (M), iron oxide (W), fayalite (F) and silica (very dark). Photomicrograph was printed darker than Fig. 11 to show contrast between iron oxide and metal. 500 ×.

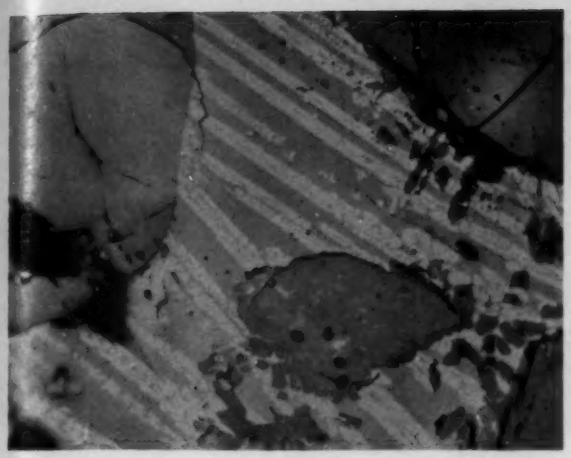


Fig. 11 — Fayalite (light plates) in glassy matrix, penetrating quartz grains. In this sample, as in Fig. 10, the fayalite formed during cooling. 500 × (Source 1).

glass Fe_2SiO_4, FeO, Fe_3O_4 and SiO_2
Fayalite Wustite Magnetite Cristobalite

Appreciable quantities of glass remain, as illustrated in the photomicrographs.

In Figs. 6 and 7 the fluxing of the alumina by silica (in particular) to form a glass is quite evident. By contrast, the regions richer in iron oxide (Fig. 9) contain hercynite and fayalite. The dendritic, idiomorphic crystals of fayalite indicate it formed after the reaction had ceased.

The specimen of ladle deoxidation scum shows the core of corundum being attacked by iron oxide to form the intermediate layer of hercynite (Fig. 9).

To summarize, the chief point is that corundum originates in the ladle as well as in the metal stream, and later reaction products are formed around it. This agglomerate reaches the cope surface and reacts further with silica and metal oxides.

It should be emphasized paticularly that the above sequence of events is proposed only for aluminum deoxidation.

SUGGESTED SOURCES OF ELIMINATION

Since the snotter is clearly related to deoxidation products, both mechanical and metallurgical methods of attack seem to offer promise.

Mechanical methods would involve improved gating systems, as suggested by Sanders.[1] An investigation of the rate of settling out of these inclusions in the ladle seems to offer promise. Obviously, if a metal is poured too soon after deoxidation, the trapped inclusions will not be filtered out by the gating and will rise to the cope surface.

Metallurgical methods would involve the study of alternative deoxidants and the resulting inclusions. The aggravation of the defect by reactions at the cope interface also requires study, since the greater the reaction the more serious the surface imperfection. The amount of aluminum or other deoxidant which is added should also receive attention. The aluminum remaining in the metal after deoxidation can react with either the mold atmosphere or with

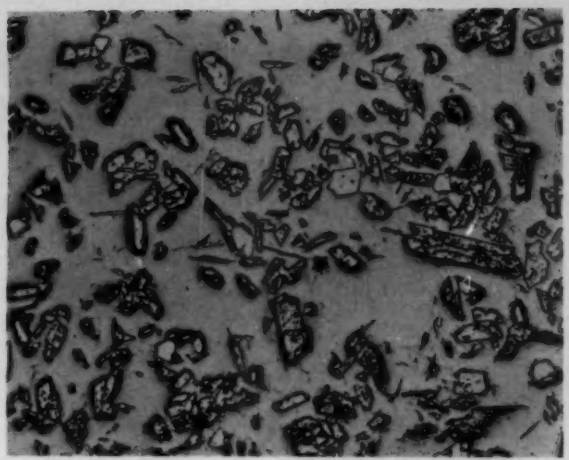


Fig. 12 — Deoxidation scum from receiving ladle. Corundum (Al_2O_3) is shown in glassy matrix identical to products found in cope surfaces of all castings. Some reaction occurred in castings between corundum and silicate melt. 500 × (Source 1).

oxygen from refractories forming additional corundum, which will probably be finally encountered at the cope surface.

CONCLUSIONS

The ceramic macroinclusions investigated here were corundum (Al_2O_3) associated with hercynite ($FeO{\cdot}Al_2O_3$), fayalite (Fe_2SiO_4), wustite (FeO), magnetite (Fe_3O_4), and a silicate glass.

Corundum is formed as a deoxidation product in the ladle (and in the metal stream passing through the mold when excess aluminum is present). Later reactions between corundum, iron oxide and silica result in the final formation of the macroinclusions. The defect is magnified by these mold-metal reactions.

Relief from this defect should be sought from mechanical and metallurgical techniques involving investigation of mold-metal reactions with other deoxidation products.

ACKNOWLEDGMENT

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In addition, the following companies contributed the material for the carefully prepared specimens mentioned in the text: Burnside Steel Foundry Co., Pelton Steel Casting Co., National Malleable & Steel Castings Co. and Wehr Steel Co.

REFERENCES

1. C. A. Sanders, "Casting Finish, Tolerance, Precision," AFS TRANSACTIONS, *vol. 64*, p. 322, 333, 1956.
2. M. Baeyertz *Non-Metallic Inclusions in Steel*, A.S.M., 1947.
3. G. A. Colligan, R. A. Flinn and L. H. Van Vlack, "Effect of Temperature and Atmosphere on Iron-Silica Interface Reaction," AFS TRANSACTIONS, p. 452, 1958.
4. Ralph Wells, L.H. Van Vlack and W. B. Pierce, "Reduction of Silica in Large Shell Molds," AFS TRANSACTIONS, p. 459, 1958.
5. J. R. Rait, H. W. Pinder, "The Origin and Constitution of Certain Non-Metallic Inclusion in Steel," *J. Iron & Steel Institute*, June 1946.

CAST SURFACES EVALUATION FOR ROUGHNESS STANDARDS

By Eldon Swing

ABSTRACT

Developments in the casting industry have progressively produced cast parts with better quality, closer tolerances and smoother surfaces. The casting surface affects the measurement and function of the part. Methods of measuring machine surfaces are not applicable to castings. Foundry produced visual comparator gages for cast surfaces have been used with satisfactory results in many production applications, and have been generally accepted as the most practical method of controlling cast surfaces for the majority of casting applications.

A standard means of evaluating the surfaces of such a roughness gage has proved to be a major difficulty. However, before the use of the gage can be completely successful, it must be possible to standardize casting surfaces.

MIL-STD-10A specifies the mathematical method of evaluating surface roughness, requiring a profile view of the surface. In the method described here, this view is obtained by cutting a section through the surface. A photomicrograph of the edge is made and the surface evaluated by obtaining linear measurements in microinches from the photomicrograph. These linear measurements are then converted to roughness height rating (RHR) values by mathematical average.

It must be noted that this method is suitable only for evaluating a gage. The visual gage is to be used only for sight and feel evaluation. Since there is a wide latitude in the readings of a single surface, there should be sufficient tolerance associated with reading production surfaces to assure that microscopes, magnifying glasses or other aides are not necessary. Values used on a visual surface roughness gage shall be only those practical for visual comparison.

INTRODUCTION

Production engineering is based on standards and specifications which are followed by the design engineer. Most design engineers are not familiar with surface roughness requirements, but follow a manual in an attempt to meet specifications. The evaluation of surface roughness is included in Specification MIL-STD-10A, which is mandatory for military applications. The present practice of cast surface roughness evaluation avoids the exact requirements of this specification, while the photomicrograph method was developed as a means of fulfilling it.

CAST SURFACE ROUGHNESS STANDARDS HISTORY

Due to the nature of the casting process, and the use of various mold techniques and materials, all castings do not have the same surface roughness. When castings were included in the manufacturing processes of the aircraft industry, many factors, including physical specifications, changed in relative importance. At first, castings were rough shapes from which parts were machined, or further processed until they were suitable for aircraft use. The surface roughness was not critical because the contacting surfaces were to be machined. Remaining as-cast surfaces were considered basically for their appearance. Flaws and discontinuities could be repaired or sometimes filled.

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After a time the aircraft castings being produced were of such quality that the as-cast surfaces could, in many applications, be used as mating surfaces. With the acceptance of permanent mold castings in the industry at the beginning of World War II, efforts were made to control surfaces so that machining was eliminated. As experience in the molding processes accumulated, efforts were made to produce the raw castings to tolerances closer to the finished part requirements. Newer developments made smoother surfaces easier to produce. The application of casting was expanded. Sealing requirements, aerodynamics, contact of mating surfaces and stress and fatigue requirements became considerations. These considerations are affected by the surface roughness, therefore, the engineer was required to specify surface roughness.

Reason for Standards

When the engineer states a requirement on the drawing, the inspector must check each detail. When no standard is agreed upon between the producer and the user of the part, a difference of opinion can often cause rejection of entirely acceptable and functional parts. The resulting misunderstandings cause undue delay and extra expense.

It soon became apparent that a suitable method of evaluating cast surface roughness must be established. Surface roughness standards were already in use for machined surfaces. One type of gage, produced by typical machine processes, had been developed to identify various degrees of standard surface roughness. The success of the machine industry's

attempt at standardizing surfaces and identifying them propagated efforts to call out cast surface roughness using the same nomenclature.

SURFACE ROUGHNESS INSPECTION PROBLEM

The previously mentioned gage, universally used for machine surfaces, does not adequately portray the surfaces produced by the casting processes. The pattern of the surface has bearing on its identification. Integrating stylus-type machines developed to read machined surfaces take into consideration the direction of the pattern of the surface roughness. Since the surface roughness pattern of a cast surface has no direction, readings made with these instruments are not reliable. Regular dial indicator instruments are not adaptable. Nevertheless, the engineer continued to use machine surface roughness standards for casting designs expecting all the requirements to be met.

No inspection organization was exempt from the problem of insufficient standards. Since no yardstick was available for cast surface roughness measurements, the only positive way to determine whether or not the part would meet functional requirements was to make a sample supply of parts and evaluate them on an individual basis. This is costly and time consuming. A need for a standardized gage was recognized by both the foundry and aircraft industries.

Visual Gages

Some foundries produced visual gages which were suitable for their operation, and which appeared suitable for use in the aircraft industry. These gages were readily available, simple and low in cost. Surface evaluations were made by their actual use in the aircraft plants. Engineers of the author's company introduced such a gage into their plants in 1953. Foundry engineers, together with liaison engineers and inspectors of the aircraft industry, closely observed the use of this gage.

After a trial period, it was determined that the visual gage met the requirements of the industry better than any gage previously tried. When initial instruction in the use of the gage was given, differences of opinion were negligible. The use of a visual-cast surface roughness gage has, therefore, won universal approval.

Visual Gage Standardization

Even though this system has won universal approval, the evaluation, qualification and control of the many varieties of cast surface roughness gage has not been established. Several projects to provide standard visual gages are underway, and methods of standardizing these gages are under study.

Several paths are available for the establishment of a reproducible visual master gage. One means is to provide large sample plates from which small samples may be cut to make duplication masters. These original samples would be of sufficient number to supply industry needs for some length of time. This method is impractical. A second method is to provide a synthetic surface so hard that no wear would be apparent, and to make visual comparisons of these surfaces from time to time against small duplications.

Specifications for checking these duplications have never been defined in a suitable manner. This method also is impractical. A third plan is conceiving a means of reading cast surfaces within a specified tolerance, which allows the reproduction of a cast surface and the evaluation of its roughness. This plan is the basis of the method herein described.

MIL-STD-10A REQUIREMENTS

It has been standard industry practice to identify surface roughness values in root mean square (RMS) microinch units. The military requirement, MIL-STD-10A, identifies the averaging method values in roughness height rating (RHR) microinch units. Numerically, roughness height ratings are always of a lower value than root mean square values. The ratio, where close accuracy is significant, is approximately 11 per cent, and this value may be used as a conversion factor.

MIL-STD-10A identifies the basic evaluation of surface roughness as shown in Fig. 1.

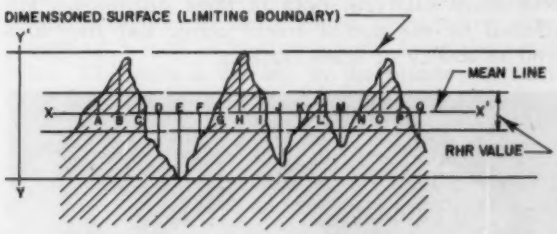


Fig. 1 — Specification MIL-STD-10A basic evaluation of machined casting surface roughness.

This specification applies to either machine or cast surfaces. In the words of this specification:

> "It shall replace all former practices for specifying finishes or surface conditions, and shall apply to any surface of sufficient hardness to be evaluated in terms of microinches under the provisions herein specified."

The mathematical evaluation is stated as follows:

$$RHR = \frac{1}{\ell}\int_{x=0}^{x=\ell} |Y|\,dx$$

An approximation of this value may be obtained by adding the Y increments without regard to sign, and dividing the sum by the number N, of the increments taken. This approximate summation formula is suggested in MIL-STD-10A:

$$RHR = \frac{A+B+C+D+\cdots}{n}$$

A true profile view of the surface is required to obtain the mathematical evaluation of A, B, C, D, etc. There are three means of obtaining this profile view. First, an electronic tracer may be used which graphs a simulated profile as a pointer is moved across the surface. This machine plots an enlarged

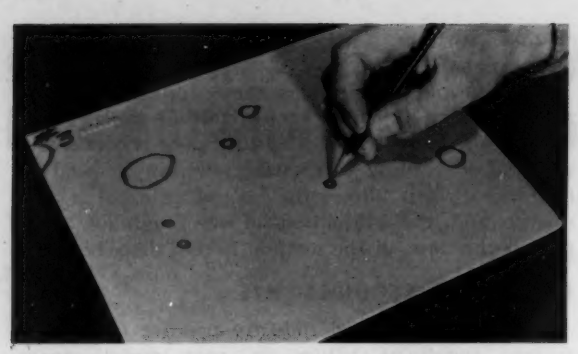


Fig. 2 — Defects in the surface were marked so they could be excluded in the sample selection.

graph of the surface as portrayed from a probe. A second method for duplicating the profile view is the use of a mechanical device. This method employs either a pantograph system, or a micrometer height gage instrument.

The pantograph system plots a graph in an enlarged view similar to the electronic tracer. The height gage spaces and reads heights directly at predetermined intervals. Both of these instruments are affected by the size of tracer point, and they also tend to scratch or mark surfaces.



Fig. 3 — Specimen area was identified by marking the selected location through a small hole in the transparent grid.

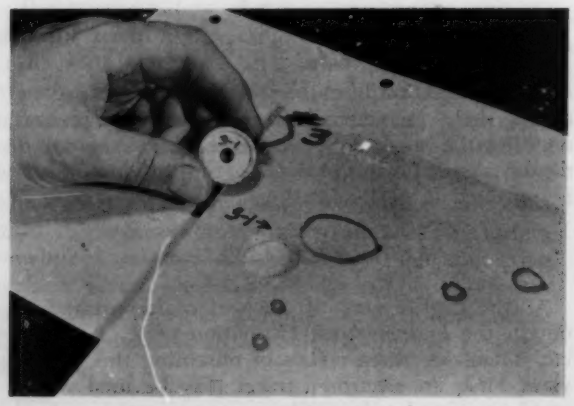


Fig. 4 — Samples cut from the plate were identified as to plate number and location.

PHOTOMICROGRAPHIC TECHNIQUE

A third method, the method reported in this paper, is the use of photomicrographs taken of a cross-section view. The photomicrographs must be carefully produced. The detailed procedure for producing these photomicrographs is:

1) Obtain sample cast plates of approximately a 9 in. x 11 in. area.
2) Identify the plates for subsequent checking as to their production process.
3) Examine the surface area to be checked visually for consistency. Any area which shows defects, either from the casting process or from subsequent handling, is marked so that it can be excluded in the sample selection (Fig. 2).

A random selection system based on probability principles was used, consisting of a transparent grid of numbered 1-in. squares laid out on a sheet of clear plastic. Numbers corresponding to numbers on the transparent grid were printed on wooden beads. The beads were placed in a box of sufficient size to allow mixing, and the transparent grid was laid over a sample plate. Five numbered beads were selected at random from the box.

The specimen area was identified by marking the selected location on the sample plate through a small hole in the center of the grid square with the corresponding number (Fig. 3). This location centered a 1-in. diameter hole saw. Five 1-in. circular pieces were cut from the sample plate by the hole saw, and each sample was identified as to the plate number, and its location on the plate (Fig. 4).

One area of the side of each of these samples was then ground to give a flat edge approximately ½-in. across (Fig. 5). No preference was given to a particular area. The samples were then given an electroless nickel coating 0.002 in. thick. This coating does not change the original surface of the sample, but adds a protective layer for subsequent processing.

Each sample was mounted in a bakelite button of a type used for metallographic studies. After mounting, the sample was polished in the same manner as though metallographic studies were to be made. No etching of the final polished surface was necessary.

A 500 × photomicrograph of each sample was taken using a standard machine. Care was taken that the view was a true edge view. The photos were



Fig. 5 — One edge of each sample was ground flat and phenolic pads were added between samples to aid in alignment and edge protection.



Fig. 6 — Photomicrograph shows the sample surface. White arrow points to the dividing line between the casting and the electroless nickel plating 0.002 in. thick used for protective purposes. 500 ×.

printed on glossy paper, and were the medium on which the linear measurements were taken. The 500 × photo, Fig. 6, brings out the dividing line between the surface and the plated layer with sufficient sharpness for taking measurements.

Most previous industry evaluations of surface roughness have been made using an 0.03 in. roughness-width cutoff value. The 0.03 in. roughness-width cutoff has proved to be one which will include most of the coarse irregularities of the surface for surface roughness values up to 500 RHR; therefore, this standard was used when obtaining the readings reported herein. Photos were spliced as necessary, because the equipment would not provide a single picture of the required 15-in. length.

A mean surface line was estimated using the 15-in. cutoff length on the photo to establish a basic direction of setup and measurements. Referring to the basic evaluation diagram, increment measurements of deviation from the mean surface, A, B, C, etc., are taken perpendicularly to the estimated mean line at 1-in. intervals.

A telereader was used to take the actual measurements. This device eased handling, and reduced human error incurred in reading scales. The machine consists of a lantern projection device wherein the image is projected twice size on a ground glass reading screen. Horizontal and vertical cross hairs are provided. An electrical counting device provides 854 increments of measurement per inch of movement of each cross hair.

Readings are taken by hand positioning the cross hairs and recording the position by an interconnected IBM computer. This provides data which are easily referred to. Sixteen readings were taken on each sample; a total of 80 readings were taken per sample plate.

RHR COMPUTATIONS

Using the previously stated summation formula taken from Specification MIL-STD-10A, and using the individual measurement readings as recorded by the digital computer, each set of readings was compiled on the basis of an 0.03-in. cutoff width. This was done by first totaling the measurements for each sample and dividing the sum by the number of readings. The results are the mean value.

The mean value then was subtracted from each measurement to get the deviation from the mean as indicated by the values A, B, C, D, etc. Each of these computed figures are again added, disregarding signs, providing a sum of the deviations from the mean value. This sum is divided by the number of readings giving an average value. Measurements and computations shown in the Table have been corrected for the 500 × magnification of the photomicrograph and converted to microinches. Values obtained are actual roughness height readings and are in accordance with the formula in Specification MIL-STD-10A.

ACCURACY EVALUATION

Based upon experience to date, it is felt that the five samples from each plate will give sufficient accuracy to indicate the RHR value of the entire

TABLE — COMPUTATIONS TO EVALUATE ROUGHNESS HEIGHT READING

Individual Measurement Readings		Deviation from Mean	
1	1838.4	A	295.8
2	1693.2	B	150.6
3	2105.4	C	562.8
4	2805.6	D	1263.0
5	2658.1	E	1115.5
6	1932.1	F	389.5
7	1740.1	G	197.5
8	1894.6	H	352.0
9	1391.1	I	151.5
10	1016.4	J	526.2
11	16.4	K	1526.2
12	1440.3	L	102.3
13	1084.3	M	458.3
14	1000.0	N	542.6
15	927.4	O	615.2
16	1138.2	P	404.4
	24681.6		8653.4

$$\frac{\text{SUM}}{\text{NOR}^{\bullet}} = \frac{24681.6}{16} = 1542.6 \text{ (Mean Value)}$$

$$\frac{\text{SUM}}{\text{NOR}^{\bullet}} = \frac{8653.4}{16} = 540.8 \text{ (Average Value)}_{\text{RHR}^{\bullet}}$$

•Number of readings.
•(RHR) — Roughness height readings.

plate. A means is thus provided to certify the surface roughness of the plate, sections of which could be used as inserts for a visual surface roughness gage.

Variations in repeat readings of photomicrographs are within 10 per cent. Allowances for lens distortion in the photomicrograph machine, and other miscellaneous variations, indicate that the RHR readings obtained by the photomicrograph process will repeat within 15 per cent. The values obtained by this photomicrograph method were compared with other reading methods. There was a wide latitude between the photomicrograph readings and those taken with the integrating stylus-type instrument (profilometer). Readings were also compared with those taken by an electronic tracer machine which plots a graph.

The readings from the tracer-type machine compared favorably with the photomicrograph readings, but no conclusions as to the accuracy of this machine were attempted. Readings were also taken with a dial indicator-type instrument, and readings checked well in line with those of the photomicrograph method. Certain mechanical features of this machine were undesirable.

VISUAL GAGE LIMITATIONS AND UTILITY

Since the visual cast surface roughness gage has proved satisfactory for production use, the photomicrograph method is proposed for use only in evaluating and standardizing a visual gage. Some misunderstanding has been experienced as to the ability of the visual gage meeting the requirements of inspection. It is of major importance that education program data for the use of the visual cast surface roughness gage accompany the distribution of the gages.

Three basic rules will improve the utility of the visual gage. These are:

1) The visual gage should be used exclusively for a sight and feel evaluation, and shall not require microscopes or various reading devices to compare surfaces. Due to the tolerance allowances involved, no more accurate readings are feasible.

2) Often too much importance is placed on whether or not a cast surface meets the roughness specification. When it is difficult to visually determine that the production casting meets the surface specification, it would be equally difficult to prove that the surface would not adequately meet the functional requirement. The tolerance of the functional requirement is somewhat arbitrary. There is wide latitude in surface roughness readings over a so-called uniform surface.

3) Values used on a visual surface roughness gage should be those practical for visual readings. Cast surfaces smoother than 63 RHR may be measured by the means established for machine surfaces, and additional standards should not be required. When cast surfaces rougher than 500 RHR are permissible, it should not be necessary to check the surface roughness.

It is not the intent of this discussion to leave the impression that casting surface roughness is an insurmountable problem. Surface roughness should create practically no problems in the foundry. Cast surface roughness will probably always be a source of irritation for inspection and sales personnel. The information available to the engineer is being increased regularly, and no doubt he will be able to evaluate the surface he needs and will know what to expect from the foundry.

It is essential that the basic problem of creating a gage with accurate and practical values be achieved. Coordination between the foundry and aircraft industry should assure this ultimate goal.

ACKNOWLEDGMENT

The material presented in this paper constitutes many hours of work by supporting members of the Boeing Airplane Co., without whom it would have been impossible to cover the latitude of work required.

The Engis Equipment Co. and Tubular Micrometer Co. are also commended for their cooperation in supplying data and equipment for comparison readings.

POROSITY, INCLUSIONS AND PINHOLES IN MALLEABLE CASTINGS

By C. A. Sanders

ABSTRACT

An approach to the porosity, inclusions and pinhole problem in the malleable foundry is put forth. Inclusions are recognized too slightly as having a direct relationship toward the creation of pinhole porosity. Hot sand and brittle sand, in the author's opinion, are the most common sources of inclusions that may create pinhole porosity in production malleable foundries. Hot sands are believed to create a chain of reactions which become difficult to control. Other reasons for porosity or pinholes in malleable castings are those of improper melting practices. Also, mold wall movement tends to aggravate a surface pinhole condition.

MALLEABLE SCRAP CAUSES

There is no concrete agreement concerning porosity in malleable metal. Removing undesirable gases in malleable metal is a problem so complicated, and attached to so many theories, that one may easily develop strong arguments to support individual claims. Porosity is one continual difference of opinion supported by little evidence which has been proved on this foundry scrap problem. No single remedy works universally for all foundries. Many foundries do not have a standard definition concerning this scrap defect. Sand technicians have always carried a principal share of the blame, perhaps more than other departments.

Pinholes and gas porosity are not always associated with inclusions (Fig. 1). The inspection of the pinhole, or gas hole, is most important or they may incorrectly be identified as to cause. Inclusions may cause as many pinholes as entrapped gas.

In many malleable foundries inclusion porosity is linked with the scrap defect known as "dirt." The author is strongly against this nomenclature, as "dirt" generally indicates an inclusion (Fig. 2). If the inclusion cannot be identified properly, how can it be corrected? The inclusion must be slag, sand, refractory, carbides, sulfides, nitrides or dross from the products of combustion or oxidation, etc.; the word "dirt" must not be used in foundry nomenclature.

Since there are so many direct and indirect causes contributing to pinhole porosity, it is incorrect for one to make a positive statement that it is principally caused from a sand condition, or metal condition or by a specific department. There are too many cases in which the accuser of the sand department ends up the guilty party.

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Usually, porosity originates from a collection of incorrectly applied practices, or incorrectly applied fundamental foundry laws. Even though slightly defective castings may be salvaged, the expenditure of time and materials to do so affects foundry profits. There is a series of combinations which may cause porosity, such as hot sands causing brittle sands which promote inclusions which develop pinholes.

It has been claimed that the number of operating factors that create porosity or gas defects usually is in excess of six. If an operation was constantly in error only 10 per cent, at the end of ten operations there could be a 100 per cent source of error.

It is often wise to invite service engineers to the foundry when difficulty arises, to determine whether



Fig. 1 — Typical gas porosity defect which appeared in the machine shop. Entrapped gas from some source is the cause of this scrapped casting.

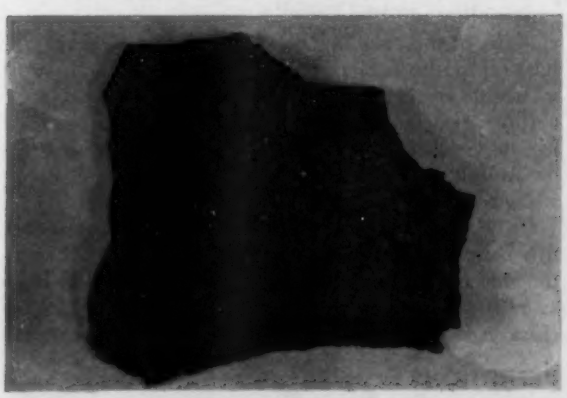


Fig. 2 — The word "dirt" does not properly identify the inclusion. Was this surface inclusion due to slag, sand, refractory or other source?



Fig. 3a — Inclusions may occur from skimmer cores. A soft, poorly prepared skimmer core may be the direct cause of sand inclusions.

or not the condition may be pointed toward sand, metal, ladles, melting practices or to other variables.

Many debates have been heard where the discussion of nonmetallic inclusions in oxidized metal was the issue. Surface oxides on the liquid metal must be trapped or skimmed, otherwise scrap may result (Figs. 3a and 3b). Metal furnace refractories erode easily under thermo-chemical activity, temperature, pressure and abrasion. The malleable foundryman is faced with the same problem as the steel foundry.

A new name is on the horizon, "ceroxides." These so-called ceroxides are thought to be products of combustion, and are low eutectic slags generally created during the deoxidizing period.

CERMETS

It has been found that certain pinhole fractures are associated with carbides and sulfide formations. In most cases the segregated areas are definitely dendritic, but tiny inclusions can also be seen at the base of the pinhole.

Numerous malleable castings have been studied that show manganese sulfide particles associated with stringy slag at the bottom of pinhole craters.

When the manganese sulfide inclusions are seen, the sulfur content is generally in excess of 11 points. Most malleable foundries use the formula for manganese content as twice the sulfur content plus 15 points. As the sulfur advances, so does the ferro-manganese alloy addition.

Many of the dark blemishes shown at higher magnification at the bottom of some dendritic fractures in malleable castings appear to be manganese sulfide inclusions associated with thin, stringy slag. The author has previously termed these "cermets." Some malleable foundries have stated that they now define this as a slag condition, and not as a pinhole gas defect.

Only by severe choking and gating the metal can this condition be eliminated. In some cases, the author has observed choking the malleable metal as little as 1 lb poured per sec to eliminate such surface inclusions.

DEFINITIONS

Experts have tried to define separately pinholes, porosity and gas defects.

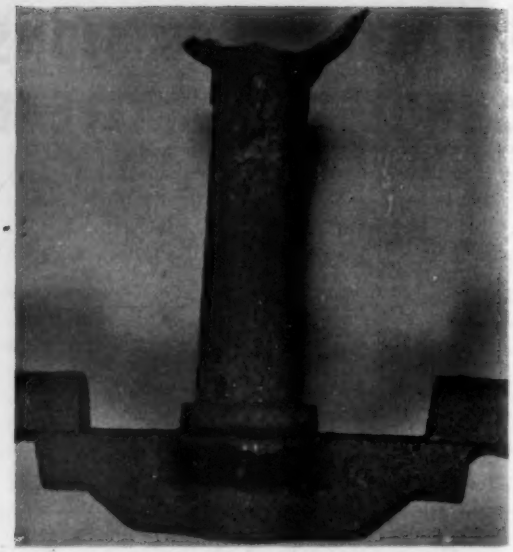


Fig. 3b — Evidence that soft rammed and poorly prepared skimmer cores will cut any road in the area below the sprue.

Porosity is most generally associated with the internal wall structure of the casting. It may be caused by gas or steam passing through the metal, which may be accompanied by inclusions of nonmetallic material or dross (Fig. 4). Porosity may appear as a rounded cavity or as a dendritic structure; it is easy to confuse with metal shrinkage.

When the surface of a casting is pitted with small holes it is known as "pinhole porosity." Surface pinholing may indicate that the sub-surface metal may contain larger porosity or gas-hole areas (Fig. 5).

Gas holes are usually rounded cavities which appear spherical, flattened or elongated (Fig. 6). They vary in size and even in color, and are usually formed as smooth depressions rather than irregular surfaces.

Some investigators claim there are "evolution pinholes," which result when gas is dissolved in the metal and released because of lack of solubility with decreasing temperatures. They claim that "reaction pinholes" are caused by particles other than metal which form gas within the metal casting (Fig. 7).

Inclusions are recognized too slightly as having a direct relationship toward the creation of pinhole po-



Fig. 4 — Another surface porosity may be the direct result of gas passing through the metal accompanied by inclusions of nonmetallics or dross.

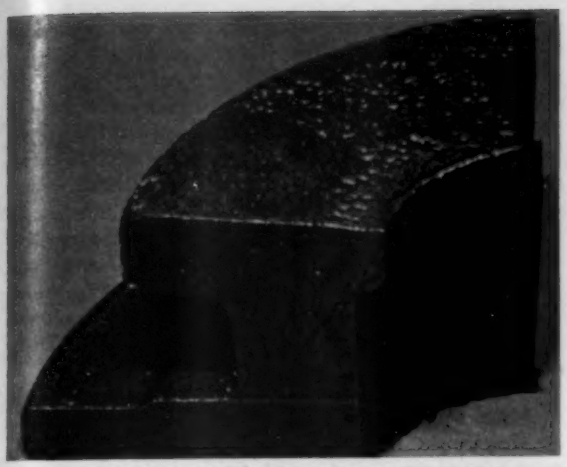


Fig. 5 — Surface pinholing may also indicate sub-surface enlarged porosity areas. Only by destructive testing can the areas be studied clearly and identification made properly.

rosity. It is the author's opinion that hot sand and brittle sand are the most common source of inclusions that may create pinhole porosity in production malleable foundries.

Hot sands are believed to be the chief influencing factor in pinhole porosity, setting into motion a chain of reactions that becomes too difficult to control, hence porosity and pinholes can occur.

Hot sands generally occur from lack of proper sand storage, and this is where the defect begins. To overcome hot sands, the foundry technician may use excess water or air to reduce the temperature (Table). Molding sands insulate themselves; therefore, it is common to find sands returning to the muller at temperatures exceeding 350 F. These sands consume gallons of water to reduce the temperature. If excess air is applied at the muller, the foundry has a chance of avoiding this defect. If there is not enough air to drive the steam from the saturated molding sand, an epidemic of pinholes can develop.

When excess water is applied to hot sand, the temperature is generally reduced to approximately 212 F

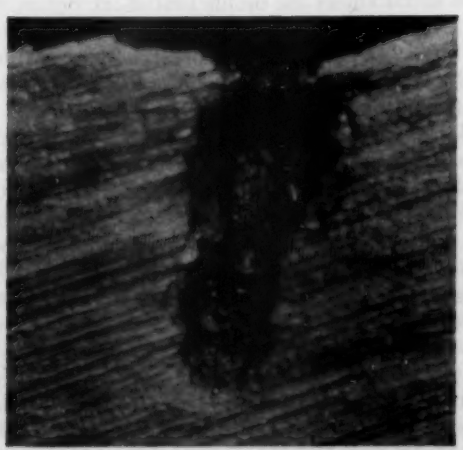


Fig. 7 — This pinhole may be called a "reaction pinhole," which may be caused from sand or other inclusions.



Fig. 6 — Many gas holes appear as cavities which may be spherical flattened or elongated, and may vary in size and color.

by conversion of water to steam. If a sand technician immediately tests sand at this stage, the laboratory temperature may show only 120 F, or perhaps less. The result is that as the sample is handled excess air is present; therefore, further cooling of the sample occurs which is not the exact condition of the sand when molded.

Actually, hot molding sand which has been decreased somewhat in temperature is still saturated with steam. When this steam-soaked sand covers the pattern and is jolted into place, numerous offending occurrences develop in the mold. Aside from sticking to metal patterns and sand handling devices in the foundry, it soon becomes the mold. The combination of metal turbulence in the mold and steam trying to escape, develops a brittle sand at elevated temperatures. Brittle sand rains sand grains onto the metal during pouring (Fig. 8).

Spalling of the sand grains can occur quite easily in a mold, which is one of the sources for inclusions, and may develop into pinholes as solidification progresses. Here is the perfect development of a pinhole. The same inclusion may be the direct cause, or hydrogen from steam in the hot sand may be carried directly into the metal by the molding sand, which rains into the mold as the casting is being poured.

TABLE — PROPERTIES OF MOLDING SAND

Properties	Molding Sand Temperature: Room Temp.	Molding Sand Temperature: 160 F	Heated Sand Comments
Green Properties			
Moisture content, %/wt.	4.0	3.6	Hot sand gives a variable moisture reading
Density, grams, 2 in. spec. ...	163	160	Penetration may result
Permeability number	75	75	
Mold hardness at 3 rams.....	91	91	
Compression strength, psi ...	14.0	12.1	more clay required
Deformation, in./in.	0.020	0.018	sand brittle
Dried and Fired Properties			
Dry compression strength, psi	150	100	loose sand inclusions
Hot compression strength, psi			
at 1650 F	160	80	
at 1850 F	100	60	thermal stability
at 2000 F	90	30	affected greatly

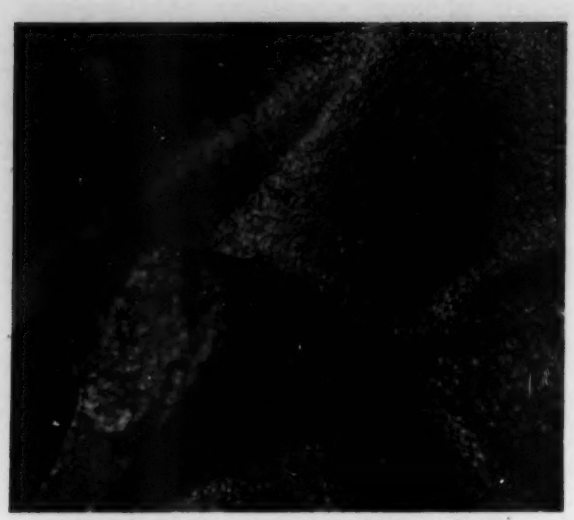


Fig. 8 — Hot sands become brittle sands, which in turn allow sand grains to wash with the inflowing metal.

Every foundryman has heard of "free moisture" and "combined moisture," but where there is an excess of water associated with an excess of bond there is always a danger of increasing the combined water. The combined water becomes an integral part of the bonded sand grain. If these fine bonded sand grains become dislodged and spall, they carry a considerable amount of mechanically combined water. When these grains are heated and converted to steam, the water expands 1600 times its own volume. The combined water then becomes a direct source for hydrogen and the chain reaction develops.

Hot sands develop faster in malleable foundries than gray iron foundries, since twice as much metal is poured by malleable foundries than gray iron per ton of salable castings, and the temperature of pouring is much higher. Management soon pays for hours of excess cleaning time, grinding time and excess labor due to this direct hot sand problem.

Scrap always increases when hot sands prevail with short mulling time. The author has seen scrap jump from a morning low of 3 per cent to as much as 37 per cent scrap in the afternoon due to the heating of the sands when it becomes unsatisfactory to mold.

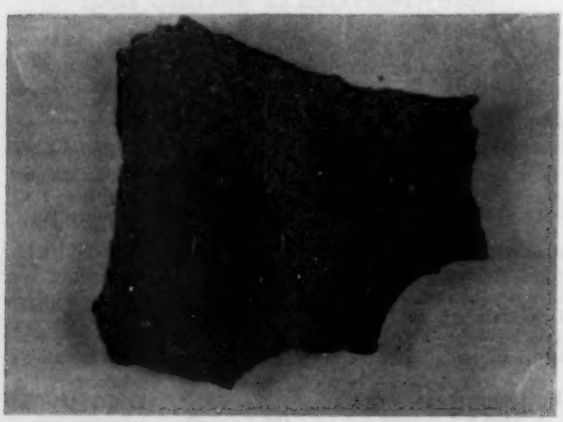


Fig. 9 — Excess additions of coarse cellulose additives tend to make sand mixtures more brittle. Brittle sands cut and wash.

Ignoring excess scrap, high cleaning costs and rougher castings due to hot sands, management may be compelled to increase the use of corn flour where hot sands have created brittle sands in order to make it workable. The author has visited foundries that increased the addition of corn flour approximately one ton per day because of hot and brittle sand mixtures. Consider these additional corn flour costs at $125.00 per day, then how many days are required before that foundry has spent enough money to have purchased a cooling device, a second or third muller or additional storage bins?

It is quite probable that foundries with hot sands and short mulling cycles are required to use more cellulose products, bond additions and fine additions, with perhaps a second or third bonding agent to consume the large amounts of water necessary to overcome hot sands. This further enhances brittle, crumbly sands at elevated temperatures.

Hot Sand Effect

Brittle sands are most generally caused by hot sands. The excess water required in cooling hot sands is normally free water, unless an addition of cellulose is added to soak up these larger additions of cooling water. Excess additions of cellulose help control the excess cooling water (Fig. 9). As the sands tend to cool, the cellulose makes a more brittle sand mixture. As the sand mixture becomes more brittle it calls for more bond material, whether it be fire clay, bentonite or bond of any type. With the increased addition of bond, soon the end of the line has been reached as far as additions to the sand mixture can be made.

In many cases, when excess bond has been added to overcome brittle sands, it has become necessary to shorten the mulling time to some extent, otherwise the sands would become stiff and lack flowability.

If metal tonnage increases slightly then the only remaining solution towards furnishing sand promptly to the molders is by further shortening the mulling cycle.

Brittle sands are unstable at elevated temperatures. When metal enters the molds heat shock occurs, causing brittle sands to break and curl at the edges on the metal flow lines. Where this breaking and curling of the thin sand layer occurs, it will appear as a rattail or buckle on the casting (Fig. 10). The danger of this breaking and curling of unstable sands is the gathering of loose and free sand grains which mix with the metal as the mold is being poured. For this reason it is best to learn where flow lines of the metal occur over the face of the mold, and to disperse the gating system to avoid this phenomenon.

Malleable foundries with these problems would be better off perhaps if they added naturally bonded sand to maintain proper temper and bond. The naturally bonded sands do not require increased mulling as they possess some strength when added to the sand mixture. It is the author's opinion that naturally bonded sand added to hot sand, or brittle sands, is a better choice than to attempt to add a fire clay or bentonite and still shorten the mulling time.

When there is a large addition of core sand entering the system, so much bond is required to maintain strength that naturally bonded sands may be best for maintaining properties. It is not believed possible to rebond core sand entering the system without a sufficient mulling cycle. It is the author's opinion that unless a fire clay or bentonite is properly mulled into the system, more harm can be done than is recognized.

Hot, brittle sands are dangerous to each foundry that tries to force them to work by such efforts as described. At first, one claims a hidden success, then after some months an apology is usually made, and the fault is always pointed towards a particular raw material in the sand system. This is a regrettable error.

As larger castings or increased melting capacity occurs, the hot sand problem magnifies. Soon the mulling time is so shortened that it becomes only a mixing cycle where any bond would do the job (Fig. 11). The bond available to the foundry would probably work best. Under these circumstances all economies must be forgotten.

Bonding Material

There are cases where a certain bonding material has worked satisfactorily at a given foundry for as long as 10 or 12 years. Yet, after this foundry has grown in production and is confronted with hot sands, pinhole porosity or other gas defects, it wishes to blame the bond for the defect. Or, it may not be the bond accused but it may be the pig iron, the alloy, etc. The direct reasons, hot sands and short mulling time, are rarely accused.

Many foundries actually delight in stating that they have worked hot sands for years and have made them work. They state, "If they have worked in the past, why not in the future?" The medical profession has the answer to that one. Many people may live around a contagious disease for years and never catch it, but there is always a first time for this eventuality. Why take the chance?

As closer tolerances and improved casting finish are demanded by the customer, these hot sands will be corrected by the foundry. When machine shop scrap becomes out of reason due to under-surface porosity, that foundry may lose the pattern shortly.

One foundry that had molding sands so hot that molders were compelled to use gloves to handle flasks and patterns claimed they had no difficulty at these temperatures except for short periods at a time. They did not consider the amount of scrap being lost daily because of these hot sands. When it becomes necessary to heat patterns in order to get proper draws, sands are too hot to work.

The author has seen cases where hot sands have been so difficult to draw from the pattern that excess liquid parting was added. Consequently, these hot sands soaked up the parting as rapidly as it was applied (Fig. 12). One can recognize how much volatile gases were present at the mold-metal interface when the hot metal ran into the mold. Excess use of liquid parting is always dangerous, particularly when there are puddles in the various contours of the pattern. Hot molding sand picks up this liquid, and soaks it



Fig. 10 — When metal enters brittle molds, the sand surface tends to break and curl at the edges of the flow lines. This will appear as a rattail or buckle. Loose, free sand grains follow the metal into the mold and are direct causes of inclusions.

up even more completely when these hot sands carry excess additions of cellulose and bond materials. Excess volatility is dangerous to good melting practice, and is a regular source of hydrogen gas porosity.

Here lies the confusion of malleable pinholes in most cases. Foundries will continue to resist the use of facings to overcome hot sands, but when facings become necessary the added cost will be accepted. They

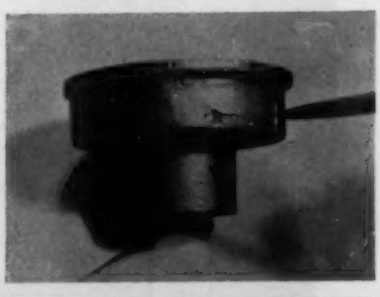


Fig. 11 — Before condemning a foundry practice, determine the reason or reasons causing the defect. Which caused this one?



Fig. 12 — Mold gas can produce pinhole porosity. Excess liquid parting compound that produces volatile gas at the mold-metal interface may contribute to this scrap problem.

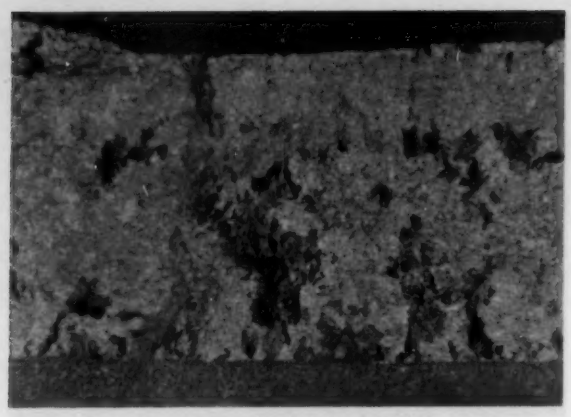


Fig. 13 — A typical spike defect. Graphitic hot tears appear on the surface when such pinhole porosity exists.

will continue to resist the suggestions of longer mulling times and the idea of purchasing an additional muller for the job. Foundries will argue against more sand storage or more aerating and cooling equipment, but will pay thousands of dollars in scrap due to this neglect.

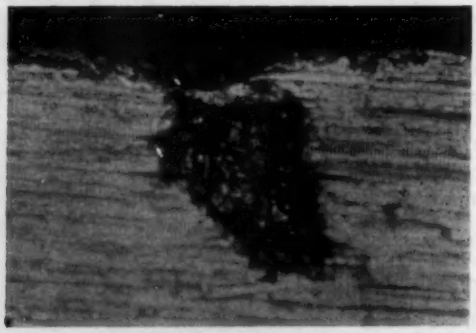


Fig. 14 — A pinhole which does not show on the surface except as a small depression. It will open upon annealing.



Fig. 15 — A pinhole. They may be round in shape and have an oxidized surface. They may appear in clusters just below the surface skin of the casting. 40 ×.

Management has lost track of why synthetic sands first entered the foundry industry; it was to maintain control and economy, and this involved the purchase of mulling equipment. If mulling capacity is too short to do the job, it is the author's opinion that management is paying the bill for excessive additives which normally would purchase mulling capacity and proper cooling and aerating equipment several times per year.

Management seems too inclined to want to purchase raw materials to add to hot sands to overcome their lack of workability rather than go directly to the source to overcome the problem. No doubt management shall continue to pay the bill for such scrap, but this bill when accepted is more likely to be lost or covered by other costs. Hot sands create a chain of reactions which become difficult to control.

There are many other reasons for porosity, or pinholes in malleable castings, particularly those of improper melting practices.

MALLEABLE MELTING CONTROL

Much of the surface pinhole porosity in malleable castings are termed "spears" or "spikes" due to their appearance (Fig. 13). Graphitic hot tears appear on the surface when this pinhole porosity condition exists.

This condition may occur in either a duplex operation or with a cold melt furnace. Larry Emery, Chief Met., Marion Malleable Iron Works, Marion, Ind., states that "the cold melt foundries which melt down charges very high in graphitic carbon run into pinhole defects quite frequently." Usually they incorrectly diagnose the cause of the difficulty as being sand. Only recently further detailed information has been accumulated to counter these claims. Many malleable heats from cold melt furnaces have not been properly super-heated and refined; thus the solidification temperature of the metal is changed (Fig. 14).

A slow rate of solidification permits gases which become entrapped in the casting to evolve. Nucleation is greater in malleable iron which is not properly refined, and pinholes may more often be present (Fig. 15). These pinholes are rounded in shape and have an oxidized surface. They may occur in clusters just below the surface skin of the casting.

Duplex Foundries

The duplex foundries encounter a similar problem if a proper oxidation loss is not maintained in both the cupola and air furnace. If a desired amount of oxidation does not occur, the pinhole defects may be predominant (Fig. 16). The defect shows immediately on vertical sidewalls of heavy sections near ingate and heavy riser areas. The responsible mechanics is the lowering of the solidification temperature, permitting the gases that are dissolved in the metal to come out of solution and become entrapped under the skin of the casting. In some cases the gas does get to the surface of the casting, and is entrapped at the mold-metal interface.

A good example of the above is in the casting of ductile iron. Solidification temperature is greatly re-

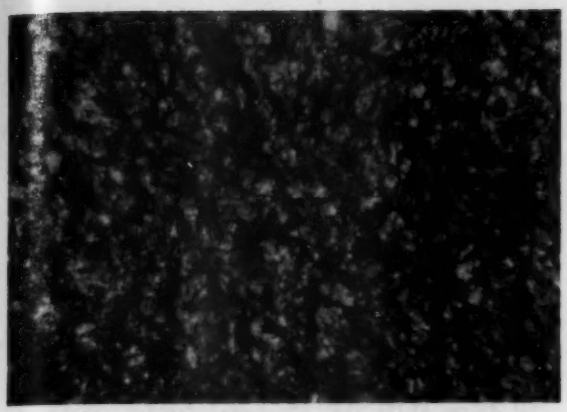


Fig. 16 — Small surface pinholes where the gas does not reach the surface before solidification occurs.



Fig. 17 — It is always difficult to analyze a surface pinhole. Inclusions must be magnified at least 40 × to be identified properly. If no inclusions are present, the condition may then be due to melting practices.

duced, and at these lower temperatures the gases present in the metal begin to work out of solution and may be entrapped on the surface of the casting (Fig. 17).

In an experiment reducing the sulfur content of a malleable metal, by reducing the sulfur to less than 0.06 per cent with soda ash, a considerable amount of pinholes developed. This was identified as a loss of temperature combined with excessive gas content in the metal.

In melting malleable iron, whether it is in a cold melt furnace or a duplex operation, it is important to maintain a minimum amount of oxidation loss. Reducing atmospheres in air furnace practice possibly are common faults in creating pinhole porosity.

Excessive additions of ferromanganese or ferrosilicon to white iron is a method of creating pinholes. If excessive additions reduce the solidification temperature of the metal being cast, pinhole porosity thus can be produced. The composition of the base metal in many cases is on the border line of producing pinhole porosity. If there is insufficient oxidation when the metal is being held in a reducing atmosphere, pinhole porosity may be magnified (Fig. 18).

Any addition of ferro-alloys that tends to reduce the solidification temperature of any metal prior to being poured should always be seriously considered, and must be viewed as a possible source of pinhole porosity. Atmosphere control in superheated or refining furnaces is now possible. The foundry continuous gas analyser records the combustion conditions in the superheat furnace. It is an important part of good melting practice to control the amount of combustibles in the furnace, as well as the introduction of excessive oxygen.

A practice that can easily influence pinholes is one that makes use of a reducing slag cover on the bath of malleable metal. It is important to keep the bath free from slag while the cupola is in operation, that is, if it is a duplexing operation. If the heat is prolonged, and it becomes necessary to use a cover slag, the degree of iron oxide in the slag must be kept to a predetermined level.

Many examinations of malleable pinholes at 250 × and 500 ×, unetched and etched with 10 per cent am-

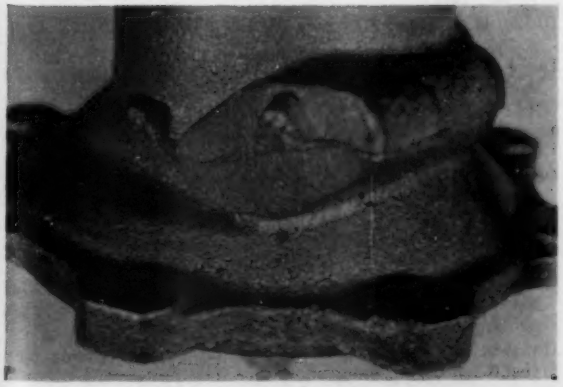


Fig. 18 — A typical surface inclusion which has been termed "pinholes." It may be an inclusion or may be gas which has occurred from the inclusion.



Fig. 19 — The area adjacent the pinholes is pearlitic. A filigree network of oxides containing primary carbides directly surrounds the pinhole cavities. There are some entrappd manganese sulfide inclusions, but not in all cases. The retention of carbides is due to the porosity and oxides. Manganese sulfide is not prominent or thought to be the cause of the pinhole.



Fig. 20 — Some surface segregation areas definitely appear dendritic. The formation of carbide needles at the fractures may be due to hot spots on the surface or by gas seeping into cracks in the semi-molten metal. Gas may also occur from an inclusion producing hot spots that are dendritic.

monium persulfate, have revealed networks of oxides containing primary carbides directly surrounding the complete pinhole area (Fig. 19). In most cases there are areas of pearlite surrounding the oxide network. In some pinhole cavities small amounts of entrapped manganese sulfide inclusions can be detected, but not in all cases. Some pinholes appear to be associated with a complicated manganese sulfide and iron sulfide compound, which also have a network of "slag bond" at the surface of the pinhole. These the author has referred to as "cermet inclusions."

A duplex operator stated that it is better to pass from the cupola with a higher carbon rather than a lower carbon content, thus reducing the carbon content in the air furnace rather than at the cupola. For example, he states it is better to come from the cupola at 2.80 per cent carbon, than 2.60 per cent carbon if a 2.50 per cent is the end point. In the air furnace the 2.80 per cent carbon may be reduced to approximately 2.50 per cent carbon, whereas the 2.60 per cent cupola iron is already too close to the 2.50 per cent, which may force a practice of holding the heat with a reducing slag.

This synthetic slag used to block the heat is dangerous, and if a too reducing furnace atmosphere develops more centers of nucleation, primary graphite may be present. It is this primary graphite that causes the carbon rings on certain malleable castings, and these hot tear areas may also show the black spears or spikes common with certain types of pinholing.

MOLD MOVEMENT POSSIBILITY

Many times a gray iron or malleable casting may be oversize and overweight. This may encourage "apparent shrinkage." These surface segregated areas definitely appear dendritic (Fig. 20). The formation of carbide needles at the fractures may be due to hot spots on the surface created by this mold movement. Furthermore, the mold movement may tend to encourage hot tearing, which may further result in a dendritic structure.

Any movement of the mold tends to aggravate a surface pinhole condition, and certain malleable foundries are now aware that seacoal seems to minimize this unusual mold movement.

REFERENCES

Andrew, J. H., Lee, H. and Quarrel, A. G., "The Determination of the Solubility of Hydrogen in Iron and Iron Alloys," Iron and Steel Institute, Adv. Copy, Sept., 1942.

Adadurov, I. E., Pevnyi, N. I., Rivlin, I. I. and Kushta, G. P., "Absorption of Hydrogen by Chromium," *Trudy Kharkov Khim-Teknol. Insti. S. M. Kirova, vol. 1,* pp. 12-18, 1939.

Allen, N. P., "Experiments on the Influence of Gases on the Soundness of Copper Ingots," *Journal of the Institute of Metals, vol. 43,* pp. 81-124, 1930; also *Engineering, vol. 129,* pp. 457-461, 1930.

Ash, E. J. and C. M. Saeger, Jr., "Methods for Determining the Volume Changes Undergone by Metals and Alloys During Casting," AFA TRANSACTIONS, *vol. 38,* p. 107, 1930.

Ash, E. J. and C. M. Seager, Jr., "Volume Changes of Cast Irons During Casting," AFA TRANSACTIONS, *vol. 40,* pp. 172-200, 1932.

Batty, G., "The Most Potent Variable," AFA TRANSACTIONS, *vol. 38,* pp. 309-331, 1930.

Boyles, A., "Influence of Atmosphere and Pressure on Structure of Iron Carbon-Silicon Alloys." *A.I.M.M.E. Transactions,* Iron and Steel Div., *vol. 135,* pp. 376-395, 1939.

Boyles, A., "The Freezing of Cast Iron," *A.I.M.M.E. Transactions,* Iron and Steel Div., *vol. 125,* pp. 141-203, 1937.

Baker, T., "Gases Occluded in Steel," Iron and Steel Institute, *Carnegie Scholarship Memoirs, vol. 3,* pp. 249-259, 1911.

Bohm, W. F., "Mold Materials Are Factors in Gray Iron Shrinkage," AMERICAN FOUNDRYMAN, *vol. 19,* no. 1, pp. 26-29, Jan., 1951.

Batty, George, "The Relation of Molds and Cores to Porosity in Steel Castings," AFA TRANSACTIONS, *vol. 42,* p. 339, 1934.

Bardenheuer, P., and M. Kunkele, "Influence of Silicon Content and Cooling Rate on the Constitution of the Phosphide Eutectic in Cast Iron," *Mitteilugen,* Kaiser-Wilhelm Institute fur Eisenforschung, *vol. 12,* No. 4, pp. 33-38, 1930.

Benedicks, C., N. Ericcson, and G. Ericson, "Bestimmung des Spezifischen Volumes von Eisen, Nickel and Eisenlegierungen um Geschmolzenen Zustand," *Archiv für das Eisenhuttenwesen, vol. 3,* p. 473, 1930.

Caine, J. B., "A Method of Control of Oxides and Gases in Steel Castings," *Steel Foundry Facts,* pp. 2-11, May, 1940.

Chipman, J., and Murphy, D. W., "Solubility of Nitrogen in Liquid Iron," *A.I.M.M.E. Transactions,* Iron and Steel Div., *vol. 116,* pp. 179-190, discussion, pp. 190-196, 1935.

Caine, J. B., "A Method of Control of Oxides and Gases in Steel for Castings," Unpublished paper presented at round table discussion, AFA Convention, May, 1940.

Chipman, J. and Murphy, D. W., "Solubility of Nitrogen in Liqui[illegible]n," *Metals Technology, vol. 2,* Jan., Tech. Publ. 591, 12 [illegible]935.

Dickie, H. A., *Journal, Iron and Steel Institute, vol. 159,* no. 4, p. 360, 1948.

Derge, G., W. Peifer and J. H. Richards, "The Sampling and Analysis of Steel for Hydrogen," *Metals Technology,* June, 1948, T. P. 2362.

Foseco Foundry Practice — Many copies *Technical Service* — Foundry Services, Ltd., England.

Gercke, E., "Experimentelle Thermische and Metallographische Untersuchung uber das System Eisen-Phosphor," *Metallurgie, vol. 5,* pp. 604-609, 1908.

Good, R. C., "Some Recent Developments on Iron and Steel Castings." *The Foundry, vol. 68,* no. 11, pp. 42-44, 103-105, 1940.

Goerens, P., and W. Dobbelstein, "Weitere Untersuchungen uber das Ternare System Eisen-Phosphor-Kolenstoff," *Metallurgie, vol. 5,* pp. 561-566, 1908.

Hultgren and G. Phragmen, "Solidification of Rimming Steel Ingots," *A.I.M.E. Transactions, vol. 135,* pp. 134-254, 1939.

Hayes, A., and Chipman, J., "Mechanism of Solidification and Segregation in a Low-Carbon Rimming Steel Ingot," *A.I.M.M.E. Transactions, vol. 135,* pp. 85-132, 1939.

Herasymenko, P., and Dombrowski, P., "Hydrogen Equilibria in the Production of Steel", *Archiv für Das Eisenhuttenwesen, vol. 14,* pp. 109-115, 1940.

Honda, K., and H. Endo, "On the Volume Change in Cast Iron During Solidification," *Science Reports,* Tohoku Imperial University, *vol. 16,* p. 19, 1927.

Honda, K., T. Kase and Y. Matsuyama, "On the Change of Volume of Cast Iron During Solidification," *Science Reports,* Tohoku Imperial University, *vol. 18,* p. 699, 1929.

Kempa, G. E., and R. W. Heine, "Effects of Some Melting Variables on Properties of Malleable Iron," AFS TRANSACTIONS, *vol. 63,* p. 19, 1955.

Kunkele, M., "Formation and Composition of the Phosphide Eutectic in Cast Iron," *Mitteilungen, Kaiser-Wilhelm Institut für Eisenforschung, vol. 12,* pp. 23-31, 1930.

Konstantinow, R., "Ueber die Phosphide des Eisens," *Stahl und Eisen, vol. 30,* pp. 1120-1121, 1910.

Klopf, Arthur S., "A Study of One Cause of Casting Defects," AMERICAN FOUNDRYMAN, p. 9, March, 1941.

Lange, E. A. and R. W. Heine, "Effects of Melting Furnace Atmospheres on Fluidity, Hot Tearing Tendency, and other Properties of Malleable Iron," AFS TRANSACTIONS, *vol. 60,* p. 169, 1952.

Lorig, C. H., and Samuels, M. L., "Some Effects of Hydrogen on the Time of Malleablization," AFA TRANSACTIONS, *vol. 50,* pp. 107-125, 1942.

Luckemeyer-Hasse, L., and Schenck, H., "Solubility of Hydrogen in Several Metals and Alloys," *Archiv für Das Eisenhuttenwesen, vol. 6,* pp. 209-214, 1932.

MacKenzie, *Pig Iron Rough Notes,* Winter and Spring, 1936.

McFerrin, W. B., "Casting Defects Common to Automotive Gray Iron," AFA TRANSACTIONS, *vol. 48,* pp. 1-30, 1940-1941.

McCabe, H. H., "Relationship of Chemical Composition & Mechanical Properties of Malleable Iron," unpublished M.S. Thesis, University of Wisconsin, 1953.

Marsh, J. S., *Principles of Phase Diagrams, Alloys of Iron Monographs, Engineering Foundation,* McGraw-Hill Book Co., New York, 193 pp., 1935.

Müller, F. C. G., "Ueber die Gasausscheidugen in Stahlgüssen," *Stahl Und Eisen, vol. 2,* pp. 531-542, 1882.

Martin, E., "Consideration of the Question of the Absorption of Pure Iron and Several of Its Alloying Elements for Hydrogen and Nitrogen," *Archiv für Das Eisenhuttenwesen, vol. 3,* pp. 407-416, 1929.

Pourcel, A., "Notes on the Manufacture of Solid Steel Castings," *Journal Iron and Steel Institute, vol. 22,* pp. 509-533, 1882.

Pearson, T. F., W. A. Baker, and F. C. Child, "Gas Porosity in Nickel Silver Castings," *Foundry Trade Journal, vol. 85,* No. 1667, pp. 145-152, Aug. 12, 1948.

Principles Involved in the Making of Rimming Steel, Iron & Steel Inst., "Fourth Report on the Heterogeneity of Steel Ingots," pp. 68-99 (See page 88), 1932.

Rosenthal, Philip C., "Ferrous Metallographic Practice", AFS TRANSACTIONS, *vol. 64,* p. 709, 1956.

Rote, F. B., "Phosphorus in Nickel-Molybdenum Gray Cast Iron," Ph.D. Thesis, University of Michigan, Jan., 1944.

Report — "A Defect Peculiar to Malleable Castings" — *Giesseripraxis,* 58.245, June 20, 1937.

Rote, F. B., and W. P. Wood, "Segregation of Molybdenum in Phosphorus-Bearing Alloyed Gray Cast Iron," *A.S.M. Transactions, vol. 35,* pp. 402-434, 1945.

Schenck, H., *Physical Chemistry of Steel-Making,* J. Springer, Berlin, *vol. 1,* 306 pp., 1932; *vol. 2,* 274 pp., 1934.

Swinden, T., and Stevenson, W. W., "Some Experiments on Gases in Iron and Steel and Their Effect on the Solidification of Ingots," *Iron & Steel Institute,* "Sixth Report on Heterogeneity of Steel Ingots," pp. 137-150, 1935.

Sims, C. E., "Preparation of Steel to Avoid Porosity in Castings," AFA TRANSACTIONS, *vol. 42,* pp. 323-338, 1934.

Saklatwalla, B., "Phosphoreisen, seiene Konstitution," *Metallurgie, vol. 5,* pp. 711-713, 1908.

Sims, C. E., "Why Deoxidizers Are Necessary," *The Foundry,* pp. 37, 109-111, Jan. 1941.

Sims, C. E. and C. A. Zapffe, "The Mechanism of Pinhole Formation," AFS TRANSACTIONS, *vol. 49,* pp. 255-281, 1941.

Sanders, C. A., and C. C. Sigerfoos, "Gray Iron Shrinkage Related to Molding Sand Conditions," AMERICAN FOUNDRYMAN, *vol. 19,* no. 2 pp. 49-55, Feb. 1951.

Saklatwalla, B., "Eisen and Phosphor, die Konstitution ihrer Verbindungen," *Metallurgie, vol. 5,* pp. 331-336, 1908.

Schwartz, H. A., Guiler, G. M., and Barnett, M. K., "The Significance of Hydrogen in the Metallurgy of Malleable Cast Iron," *A.S.M. Transactions, vol. 28,* pp. 811-829, 1940.

Stead, J. E., "Iron and Phosphorus," *Journal of the Iron and Steel Institute, vol. 58,* part 11, pp. 60-155, 1900.

Schneidewind, R., and C. Harmon, "Microstructure of Silvery Pig Iron," AFS TRANSACTIONS, *vol. 55,* pp. 602-606, 1947.

Sims, C. E., and Zapffe, C. A., "The Mechanism of Pinhole Formation," AFA TRANSACTIONS, *vol. 49,* pp. 255-270; discussion pp. 270-281, 1941.

Sieverts, A., "The Absorption of Gases by Metals," *Zeitschrift für Metallkunde, vol. 21,* pp. 37-46, 1929; Abstract, *Metallurgist, vol. 5,* pp. 168-172, 1930.

Sanders, C. A., "One Answer to the Snotter Problem," MODERN CASTINGS, May, 1956.

Smith, Eugene W., "A Solution to Pinholes," *Pig Iron Rough Notes,* Reprint — no date.

Timmins, A. A., "Shrinkage and Contraction of Cast Iron," *Foundry Trade Journal, vol. 59,* pp. 191-193, Sept. 15, 1938.

Timmons, G. A., "Theory of Shrinkage in Gray Cast Iron," *The Foundry, vol. 70,* no. 12, pp. 82-83, 1942; No. 1, p. 80, 1943.

Turner, T., "Volume and Temperature Changes During the Cooling of Cast Iron," *Journal of the Iron and Steel Institute, vol. 69,* p. 48, 1906.

Vogel, R., "Ueber das System Eisen-Phosphor-Kohlenstoff," *Archiv für das Eisenhuttenwesen, vol. 3,* pp. 389-381, 1929-1930.

Von Keil, O. and R. Mitsche, "Der Einfluss des Siliziums auf das System Eisen-Kohlenstoff-Phosphor," *Archiv für das Eisenhuttenwesen, vol. 3,* pp. 149-156, 1929-1930.

West, W. and Hodgson, C. C., "Porosity and Sinking in Cast Iron," *Proceedings of the Institute of British Foundrymen,* paper No. 699, *vol. 32,* pp. 421-441, 1938-1939; discussion, pp. 442-448, 1938-1939.

West, W., "Phosphorus in Cast Iron," *Metallurgia, vol. 7,* pp. 83-88, 1933.

Wust, F., "Beitrag, Zum Einfluss des Phosphors auf das System Eisen-Kolenstoff," *Metallurgie, vol. 5,* pp. 73-87, 1908.

Zahn, Charles, "Porosity or Segregation in Gray Iron Castings," AMERICAN FOUNDRYMAN, p. 14, July, 1942.

Zapffe, C. A., "An Effect of Dissolved Gases on Casting Structure," to be published.

Zapffe, C. A., and Sims, C. E., "Hydrogen Embrittlement, Internal Stress, and Defects in Steel," *A.I.M.M.E. Transactions,* Iron and Steel Div., *vol. 145,* pp. 225-261, 1941; discussion pp. 261-271; *Metals Technology, vol. 8,* Technical Paper 1307, 37 pp., 1941.

LINEAR PROGRAMMING APPLIED TO ECONOMIC SELECTION OF MATERIALS FOR FURNACE CHARGES

By Gideon I. Gartner

ABSTRACT

Many combinations of furnace input materials exist which will meet a set of given furnace output requirements. The linear programming technique can be used to choose that feasible material combination whose material cost is a minimum. The technique can be used to calculate optimum prescriptions for individual heats, or to determine which materials to buy to stock in inventory. This paper is concerned with the "what to buy" decision; a hypothetical problem is solved on a high-speed digital computer, and several implications relating to the problem are discussed.

INTRODUCTION

At some point in the operation of a foundry, decisions must be reached which specify the particular furnace input materials (ferroalloys, bar stock, pig, etc.) to be purchased. The materials which are bought are eventually drawn from inventory, and melted together to form a product of some specified composition and gross weight.

It is unfortunately true that the decision concerning "what to buy" is often made without consideration as to the infinite number of possible combinations of different input types (drawn from the catalogs of the many furnace input producers) which will meet the specified requirements. How is one to choose among these many feasible combinations? Does a criterion in fact exist which allows these different combinations to be compared? Obviously, yes, since each feasible combination will generally involve a different overall material cost, and this cost should be minimized.

It will be shown that the technique called "linear programming" can be used to systematically evaluate the many possible combinations of inputs according to the least-cost criterion, thereby allowing the solution to the best-input-mix problem to be readily obtained.

LINEAR PROGRAMMING

Linear programming* has previously been successfully applied to "best mix" problems as described. For example, a Columbia economist has computed the least-cost mix of foods that a person could consume and yet meet all the accepted nutritional requirements; farms have applied the method practically in the determination of least-cost cattle feeds; and the petroleum industry in particular has realized substantial cost savings in applying L. P. (linear programming) to gasoline blending problems.[1,2,3]

In general mathematical terms, the technique applies to those processes or activities which are subject to a set of linear restrictions in the form of equalities or inequalities. If the number of feasible solutions within the constraints is large, L. P. can choose that particular solution among all feasible ones which optimizes an objective linear function.

Furnace Charge Application

The charging of metals in a furnace is an activity which fits the above description. One could wish to choose the least-cost mix of raw materials which, when melted, will give a product of known total weight and chemical composition. Since each raw material is of different composition, the linear restrictions are those which relate the contents of each chemical element in each raw material to the required content of that element in the end product. If X_i is the unknown weight to be charged of each raw material i, and P_{ij} is the per cent by weight of each element j in each raw material i, then

$$\sum_i X_i P_{ij} \left[\begin{matrix} \geq \\ = \\ \leq \end{matrix} \right] R P_j$$

where R is the total weight of the heat and P_j is the required percentage of element j. The equations represented by the above symbology are in fact element balances; some appear as equalities, for example the restriction that manganese content equal some specified amount; others appear as inequalities, e.g., the requirement that sulfur (an impurity) content shall be less than or equal to some specific limit. Others are also in inequality form, but appear as part of a two-equation set stating that the final content shall lie within a specified range.

There is another linear restriction which relates the weight of each raw material charged to the total weight of the heat:

$$\sum_i X_i = R$$

*The Appendix contains a geometrical explanation of linear programming for the interested reader.

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TABLE 1 — CHARGE MATERIALS INVENTORY

No. Corresponding to Materials Listed in Table 2	Materials Stocked in Inventory	Composition, % C	Mn	Si	Cr	Ni	Cost, $/lb	Lb Required in prescription per 1000 lb
2	FeMn	2.0	82.5	1.5			0.25	22.3
5	FeSi	0.15		50			0.20	23.4
9	FeCr	0.04		1.5	70		0.41	277.7
11	Ni	0.1				99.9	0.74	139.1
14	Pig	4.42	0.13	0.7			0.034	30.1
17	Generated Scrap	0.22	1.8	0.65	25	20	0	300.0
18	Stock	0.25	0.85	0.30	0.50	0.50	0.085	207.4

Total cost of 1000-lb prescription: $245.55

Still other restrictions appear in the metal-charging problem, according to the particular problem being solved. For example, one might limit the utilization of specific raw materials according to their availability. For example, if home-scrap is used, then the amount generated must limit the amount utilized.

The linear restrictions mentioned in the foregoing make up a set of simultaneous linear equations and/or inequations,* with the X_i's being the unknown variables. Linear programming extracts an optimum solution to the foregoing set of equations from an infinity of feasible solutions. The condition for more than one unique solution to exist is that the number of unknowns exceed the number of equations. It should be pointed out that whenever an inequality exists, an additional unknown is introduced which converts the inequality to an equality.

This unknown is called a "slack variable," since it is introduced to take up the slack (the extent of the inequality) between both sides of the equation (Appendix 1 clarifies the slack concept).

Once it is established that a great number of possible solutions exist, an "objective" linear function of the unknown variables (raw materials) must be set up to enable the L. P. technique to evaluate the many different possible solutions. Since the objective in this problem is the minimization of cost, a linear expression is constructed:

$$\sum_i X_i C_i = \text{Minimum}$$

where c_i is the cost per pound of the ith raw material.

With the problem formulated, the L. P. computational procedure† is used to extract a solution. The procedure is iterative, requiring a systematic introduction of variables. As much of each chosen variable is brought into solution as the restrictive equations will allow, while amounts of other variables already in solution are adjusted. The method used to choose the variable being introduced at each step causes the results to converge toward a least-cost solution.

SOLUTION OF A TYPICAL PROBLEM

A steel manufacturer is producing S.A.E. 310 stainless steel with the following charged composition (furnace losses are added to the required composition, to give the charged composition):

Composition, per cent						
C	Mn	Si	Cr	Ni	P	S
0.25 max	2.00	0.75	24-26	19-22	0.03 max	0.03 max

*An inequation represents an inequality.

He carries in inventory the materials shown in Table 1, in which compositions and costs are listed. To meet the required end composition, and with the stipulation that generated scrap can contribute no more than 30 per cent of the total charged weight, a melter's calculation gives the prescription per 1000-lb heat shown in the right hand column of Table 1, at a total material cost per 1000 lb of $245.55.

The manufacturer wishes to know if he can reduce the average cost per heat by carrying a different set of raw materials in stock. He studies the market, and lists alternate materials to the ones being used. From this point, knowing the compositions and costs of the alternates, the least-expensive combination of inputs that will meet all the requirements can be found by L. P.

This was done for a total of 19 materials (listed in Table 2), including such alloy-rich metals as silicomanganese and ferro-chrome-silicon, two types of bar stock and several grades each of FeMn, FeSi, Cr, FeCr, Ni. The 11 equations (evaluating 25 unknowns) which are used in the L. P. solution are listed in Fig. 1. The equations are analogous to those symbolically represented earlier with a few additional refinements; each equation includes at least one slack variable with which is associated a high cost (M) in the objective function.

High Cost Slack Variables

These high-cost slacks are inserted for computational purposes only, and will be driven out of solution during computation, hence the name "artificial" slack variables. Equations a, c, d, f, g, h and i contain additional real slack variables, because these equations are in fact inequalities. The cost associated with real slacks is zero, and, ergo, these slacks do not appear in the objective function.

There are two equations written for the chrome balance, and two for the nickel balance. This is because requirements for these two elements allow a range of possible composition values. For each of these elements, one equation represents the upper limit (content to be less than or equal to x), and the other represents the lower limit (content to be greater than or equal to y).

A matrix of coefficients is constructed, and information from the matrix is encoded in machine language for the computer to operate on. Hand or desk calcu-

†Several algorithms have been developed to solve L. P. problems. The most widely used of these is the simplex technique, and the interested reader may learn the theoretical and computational details of this method from the literature.[4,5,6]

lator computation could be attempted depending on the size of problem, and for this a simplex tableau would be constructed.* For the particular problem under discussion, hand computation is infeasible.

The machine solution to this problem is shown in Table 3 which corresponds to Table 1, but lists the materials which L. P. says will, when combined, satisfy the requirements and restrictions at least cost. This least cost is $220.88, or about $24.00 saved per 1000 lb: ($48,000 per 1000 tons). The final cost is most sensitive to those materials used in greatest quantities. So if several other bar stocks were evaluated in the problem, a lesser-cost solution might well have been found.

In general, the more materials evaluated, the more likely are we to find a lower cost solution. There are two reservations to the above statement. First, it is entirely possible that certain materials can be eliminated from consideration by inspection (aside from those materials that will not even do from technological considerations). Second, it is assumed that no extra computational costs are incurred when more materials are evaluated. This assumption is a reasonable one, especially when the computation is by machine.

*The second half of the Appendix shows the simplex solution to a simplified problem.

FIG. 1 — EQUATIONS USED IN THE LINEAR PROGRAMMING SOLUTION OF THE LEAST-COST-MIX PROBLEM

A. DEFINITION OF TERMS

Let X_i represent the number of lb of input i used in the solution. Subscripts (i's) correspond to the numbers associated with the list of inputs in Table 2.

S_{jk} represent slack variables inserted for computational purposes in equation j.

M be a high cost associated with certain artificial slack variables, so that these "artificial" variables will be driven from solution.

R be the total weight requirement.

B. THE EQUATIONS

a) Carbon Balance
$$0.50X_1 + 2.00X_2 + 0.10X_3 + 0.15X_4 + 0.15X_5 + 0.50X_6 + 1.50X_7 + 0.10X_8 + 0.04X_9 + 1.0X_{10} + 0.10X_{11} + 0.25X_{12} + 0.75X_{13} + 4.42X_{14} + 0.05X_{15} + 1.5X_{16} + 0.22X_{17} + 0.25X_{18} + 0.13X_{19} + S_{a1} + S_{a2} = 0.25R$$

b) Manganese Balance
$$95X_1 + 82.5X_2 + 82.5X_3 + 0.13X_{14} + 66.5X_{16} + 1.8X_{17} + 0.85X_{18} + 0.50X_{19} + S_b = 2R$$

c) Phosphorus Balance
$$0.25X_1 + 0.35X_2 + 0.35X_3 + 0.05X_4 + 0.05X_5 + 0.06X_6 + 0.15X_6 + 0.019X_{14} + 0.03X_{17} + 0.04X_{18} + 0.025X_{19} + S_{c1} + S_{c2} = 0.03R$$

d) Sulfur Balance
$$0.10X_1 + 0.05X_2 + 0.05X_3 + 0.04X_4 + 0.04X_5 + 0.04X_6 + 0.06X_7 + 0.02X_{11} + 0.05X_{12} + 0.07X_{13} + 0.026X_{14} + 0.03X_{17} + 0.04X_{18} + 0.025X_{19} + S_{d1} + S_{d2} = 0.03R$$

e) Silicon Balance
$$1.0X_1 + 1.5X_2 + 1.25X_3 + 42.5X_4 + 50X_5 + 27.5X_6 + 16X_7 + 1.5X_9 + 1.5X_{10} + 0.70X_{14} + 43.5X_{15} + 19X_{16} + 0.65X_{17} + 0.30X_{18} + 0.30X_{19} + S_e = 0.75R$$

f) Chrome Balance, upper limit
$$96X_8 + 70X_9 + 70X_{10} + 37.5X_{15} + 25X_{17} + 0.50X_{18} + 3.5X_{19} + S_{f1} + S_{f2} = 26R$$

g) Chrome Balance, lower limit
$$96X_8 + 70X_9 + 70X_{10} + 37.5X_{15} + 25X_{17} + 0.50X_{18} + 3.5X_{19} + S_{g1} - S_{g2} = 24R$$

h) Nickel Balance, upper limit
$$99X_{11} + 98.9X_{12} + 97.7X_{13} + 20X_{17} + 0.50X_{18} + 1.6X_{19} + S_{h1} + S_{h2} = 22R$$

i) Nickel Balance, lower limit
$$99X_{11} + 98.9X_{12} + 97.7X_{13} + 20X_{17} + 0.50X_{18} + 1.6X_{19} + S_{i1} - S_{i2} = 19R$$

j) Equation limiting the scrap used to 30% of total weight
$$X_{17} + S_j = 30R$$

k) Total weight balance
$$X_1 + X_2 + X_3 + X_4 + X_5 + X_6 + X_7 + X_8 + X_9 + X_{10} + X_{11} + X_{12} + X_{13} + X_{14} + X_{15} + X_{16} + X_{17} + X_{18} + X_{19} + S_k = 100R$$

l) The objective function (cost equation)
$$28X_1 + 25X_2 + 38X_3 + 24.2X_4 + 20X_5 + 17X_6 + 14X_7 + 131X_8 + 41X_9 + 37.5X_{10} + 74.2X_{11} + 73X_{12} + 71.6X_{13} + 3.4X_{14} + 33X_{15} + 14.5X_{16} + 8.5X_{18} + 10.5X_{19} + MS_{a1} + MS_b + MS_{c1} + MS_{d1} + MS_e + MS_{f1} + MS_{g1} + MS_{h1} + MS_{i1} + MS_j + MS_k = \text{MINIMUM.}$$

Further Cost Reductions

It is possible to reduce the average material cost per heat still further, although probably not to any great extent. The number of materials utilized in a prescription is limited in the L. P. procedure by the number of equations, or rows in the matrix of coefficients. The number of rows equals material inputs plus slack variables. If one slack variable could be driven out, an extra material input could be brought in, which might reduce the overall material cost.[7] A slack can generally be driven out by reducing a requirement coefficient by the amount of the slack which came out of solution, and recomputing.

This was done, and the new materials to be charged (8 in all, as against 7 before) are listed in Table 4. The new cost per 1000 lb is $219.21. If such a meagre cost saving (compared with the $220.88 prescription) is typical, the extra ordering and handling costs imposed by the extra input would negate the use of this last prescription, unless other considerations prevailed.

It is a fact that almost always materials exist which are substitutes for others. The L. P. method tests the economic suitability of using such substitutes. For example, in the last prescription a more expensive bar stock (No. 19) was chosen which, because of its rich alloy content, lessened the amounts prescribed of certain ferro-alloys. Thus, the final cost was still comparable (in fact less, because of the additional input allowed) despite the use of 341 lb of an "expensive" material. It often happens that utilization of the expensive input results in a cheaper prescription.

All L. P. computations were accomplished on an I.B.M. 650 machine. A linear programming canned program was available, which made programming the computer unnecessary. L. P. routines have been written for several of the available medium and large scale digital computers, and if the foundry does not own or share such a machine, the service bureaus of the larger computer manufacturers will make them available for a fee. Once again, an economic decision must be made before embarking on such a program. However, since for a machine comparable to the I.B.M. 650, the calculation lasts but a few minutes, and will be required several times a year at most, the computational costs are likely to be reasonable.

CONCLUSION

Those foundries which have a daily choice of charge inputs and a computer on hand might give some thought to the calculation of charges on a day-by-day basis. Even without a computer, there are successful manual techniques which have been developed, usually incorporating simplifications to ease the computational burden. Descriptions of these techniques can be found in the literature.[6,7]

The L. P. application described in this paper allows a foundry to make a choice of inputs to stock in inventory. The actual day-by-day charges will generally be calculated separately, unless day-by-day conditions are identical to those incorporated in the L. P. model. For foundries using L. P. to calculate the actual charges, D. E. Debeau[8] points out a limitation in the use of the technique: The composition of each charge input must be known within narrow limits. In those foundries where non-uniform purchased scrap is a major ingredient, L. P. is likely to prescribe a substantial amount of this low-cost scrap, which because of the variability of its composition will result in a relatively high number of off-heats.

A foundry may wish to know the best materials to

TABLE 2 — LIST OF THE 19 MATERIALS WHICH WERE EVALUATED IN THE DETERMINATION OF BEST INPUTS TO STOCK IN INVENTORY

Available Materials	Composition, %							Cost, ¢/lb
	C	Mn	P	S	Si	Cr	Ni	
1 FeMn	0.50	95.0	0.25	0.10	1.0			28.0
2 FeMn	2.00	82.5	0.35	0.05	1.5			25.0
3 FeMn	0.10	82.5	0.35	0.05	1.25			38.0
4 FeSi	0.15		0.05	0.04	92.5			24.2
5 FeSi	0.15		0.05	0.04	50.0			20.0
6 FeSi	0.50		0.06	0.04	27.5			17.0
7 FeSi	1.50		0.15	0.06	16.0			14.0
8 Cr	0.10					96		131.0
9 FeCr	0.04				1.5	70		41.0
10 FeCr	1.0				1.5	70		37.5
11 Ni	0.10			0.02			99.0	74.2
12 Ni	0.25			0.05			98.9	73.0
13 Ni	0.75			0.07			97.7	71.6
14 Pig	4.42	0.13	0.019	0.026	0.70			3.4
15 FeCrSi	0.05				43.5	37.5		33.0
16 SiMn	1.5	66.5			19.0			14.5
17 Scrap	0.22	1.8	0.03	0.03	0.65	25	20.0	0.0
18 Stock	0.25	0.85	0.04	0.04	0.30	0.50	0.50	8.5
19 Stock	0.13	0.50	0.025	0.025	0.30	3.5	1.6	10.5

TABLE 3 — MACHINE SOLUTION OF PROBLEM

No. Corresponding to Materials Listed in Table 2	Materials to Stock in Inventory	Composition, %					Cost $/lb	Lb Required in prescription per 1000 lb
		C	Mn	Si	Cr	Ni		
1	FeMn	0.50	95.0	1.0			0.28	9.3
16	Si-Mn	1.5	66.5	19.0			0.14	3.6
10	FeCr	1.0		1.5	70.0		0.375	64.1
9	FeCr	0.04		1.5	70.0		0.41	169.3
11	Ni	0.1				99.5	0.74	129.0
17	Generated Scrap	0.22	1.8	0.65	25.0	20.0	0.0	300.0
18	Stock	0.25	0.85	0.30	0.50	0.50	0.085	324.7

Total cost of 1000-lb prescriptions: $220.88

TABLE 4 — RECOMPUTING THE PROBLEM

No. Corresponding to Materials Listed in Table 2	Materials to Stock in Inventory	Composition, %					Cost $/lb	Lb Required in prescription per 1000 lb
		C	Mn	Si	Cr	Ni		
1	FeMn	0.50	95.0	1.0			0.28	9.8
16	SiMn	1.5	66.5	19.0			0.14	4.5
10	FeCr	1.0		1.5	70.0		0.375	121.7
9	FeCr	0.04		1.5	70.0		0.41	96.9
13	Ni	0.75				97.7	0.72	7.1
11	Ni	0.1				99.5	0.74	118.2
17	Generated Scrap	0.22	1.8	0.65	25.0	20.0	0.0	300.0
19	Stock	0.13	0.50	0.30	3.5	1.6	0.105	341.7

Total cost of 1000-lb prescription: $219.21

stock in inventory for a specific production schedule, or to meet a fluctuating demand. In the first instance the required chemical analyses for the individual company products can be pooled together, and the optimum raw-material mix to be stocked may then be calculated by L. P. The foundry may wish to recompute the optimum raw-material mix whenever there is a cost-change in one of the inputs, when there is a product requirements change, when new materials become available or periodically.

If the foundry employs economic lot-size formulas for the determination of stock quantities, a new prescription calling for substitute materials could not be put into effect until old materials are utilized. Therefore, the foundry might wish to standardize the inventory cycle* for all charge input materials, and calculate the optimum charge by L. P. at each cycle.† Whether or not such a standard cycle is used, depends on a cost analysis which needs to be undertaken.

In general, the total variable inventory cost for all charge inputs using the standard inventory cycle time, will exceed the total cost when each input is stored at its own optimum level. Before adopting a change in the inventory system, this extra expense must be compared with the estimated savings from linear programming prescriptions, and the convenience of ordering all charge materials together.

Since each application is individual, it remains for each foundry to determine if such methods as were described in this paper could be incorporated (inevitably with alterations) into their system to advantage. The systems analysis should be done carefully, lest some significant hidden cost or benefit be overlooked.

ACKNOWLEDGMENT

The cooperation of the I.B.M. Corporation during the computational phase of this program, is gratefully acknowledged.

*Ordinarily the inventory cycle is different for each material.

†It is assumed that the charge inputs represent a relatively small fraction of the foundries total purchasing volume, so that ordering all 8 inputs at once does not overtax the space available.

REFERENCES

1. A. Charnes, W. W. Cooper and B. Mellon, "Blending Aviation Gasolines — A Study in Programming Interdependent Activities in an Integrated Oil Company," *Econometrica, XX,* 2 (April 1952).
2. D. C. Lennon, "Operational Research: Its Applications to the Oil Industry," *Petroleum, 17,* No. 4 (April 1954).
3. G. H. Symonds, "Linear Programming Solves Gasoline Refining and Blending Problems," *Industrial and Engineering Chemistry, 48,* (March 1956).
4. A. Charnes, W. W. Cooper and A. Henderson, "An Introduction to Linear Programming," John Wiley & Sons, Inc., New York, (1953).
5. T. C. Koopmans (ed.), "Activity Analysis of Production and Allocation," Cowles Commission for Research in Economics, No. 13, John Wiley & Sons, Inc., New York, (1951).
6. O. R. Bibliography, "A Comprehensive Bibliography on Operations Research," Publications in Operations Research, No. 4, John Wiley & Sons, Inc., New York, (1958).
7. John Coakley, I.B.M. Corp., personal discussion (1958).
8. D. E. Debeau, "Linear Programming Isn't Always the Answer," *Operations Research,* (June 1957).

APPENDIX

Explanation of the L. P. Technique

Let us study a grossly simplified version of the metal-charging problem. Assume that a foundry must produce 100 lb of a product with specifications as follows: Percentage composition of element 1 not to exceed 0.05, element 2 not to exceed 0.07, element 3 not to exceed 0.07. Two raw material inputs called *A* and *B* are available, with element compositions and costs as specified in Table 5.

TABLE 5 — RAW MATERIAL INPUTS

Element	Composition of Input A	Composition of Input B	Max. Allowable content in end product
1	0.02	0.07	0.05
2	0.05	0.06	0.07
3	0.08	0.04	0.07
Cost ¢/lb:	5	4	

100 lb required.

Based on the given information, three inequations can be constructed, each inequation representing an element balance. These balances are the "linear restrictions" referred to in the body of the report:

Balance for

Element 1: $0.02X_A + 0.07X_B \leq 0.05(100)$ *(1)*

Element 2: $0.05X_A + 0.06X_B \leq 0.07(100)$ *(2)*

Element 3: $0.08X_A + 0.04X_B \leq 0.07(100)$ *(3)*

where X_A is the no. of lb of input *A* to be prescribed, and X_B is the no. of lb of input *B* to be prescribed.

Another linear restrictive equation can be written specifying that the total weight of the heat be 100 lb.

$$X_A + X_B = 100 \qquad (4)$$

Because expressions (1), (2) and (3) are inequalities, slack variables (S_1, S_2, and S_3) are introduced which convert the expressions to equations.

$$0.02X_A + 0.07X_B + S_1 = 0.05(100) \qquad (1)$$

$$0.05X_A + 0.06X_B + S_2 = 0.07(100) \qquad (2)$$

$$0.08X_A + 0.04X_B + S_3 = 0.07(100) \qquad (3)$$

$$X_A + X_B = 100 \qquad (4)$$

The slack variables increase the number of variables to five. Because only four equations can be written, a multitude of feasible solutions exists. These many solutions are evaluated according to the "objective function," which represents in this case the minimization of cost:

$$\text{Cost to be minimized } (\$) = 0.05X_A + 0.04X_B$$

The geometrical representation of the above problem is as follows:

In Fig. 2, let the ordinate represent lb to be used of input *A*, and the abcissa represent lb to be used of input *B*.

If we were to satisfy the element 1 requirements entirely with input *A*, we would require at most 5/.02 = 250 lb of *A* (5 is the max. no. of lb of element 1 re-

quired, 0.02 is the percentage of element 1 found in input *A*). If we were to satisfy the element 1 requirements with input *B*, we would require at most 5/.07 = 71 lb of *B*. The line connecting the 250-lb point for *A* and 71-lb point for *B*, represents the linear restriction written out previously as equation (1).

Any combination of inputs *A* and *B* described by a point on this straight line, will meet the maximum allowable requirement for input 1. Since this is a maximum requirement, the area enclosed by the two axes and this straight line represents the "feasible space." Any point within this area satisfies the first restriction.

Similarly, one can draw the restrictive lines for elements 2 and 3. These lines are properly labeled in Fig. 2. For each of these lines (restrictions), feasible space includes the triangle formed by the line and the axes. It can readily be seen that feasible space for the three restrictions when taken together includes only the area which is shown shaded in the figure.

There is one additional linear restrictive equation which must be considered [Equation (4)]. It says that 100 lb of input material must be used. There are no maximum or minimum limits here, exactly 100 lb are called for, and the straight line representing combinations of *A* and *B*, which total 100 lb, is shown dashed in Fig. 2. This line reduces feasible space from the shaded area to that portion of the straight dashed line which intersects the shaded area.

The problem now is to find the point along the feasible portion of the dashed line at which the foundry should operate; e.g., the most economical mix of *A* and *B* to use, while still satisfying all the restrictions which were set up.

One does this by plotting the objective function on the same graph. The dotted line shows the cost function describing a cost of $1.00. Any point on this line represents a combination of inputs *A* and *B* whose cost equals $1.00. Other cost functions are straight lines parallel to the $1.00 line. If one finds the line closest to the origin which just intersects feasible space, that line gives the lowest cost which may be obtained while satisfying the requirements.

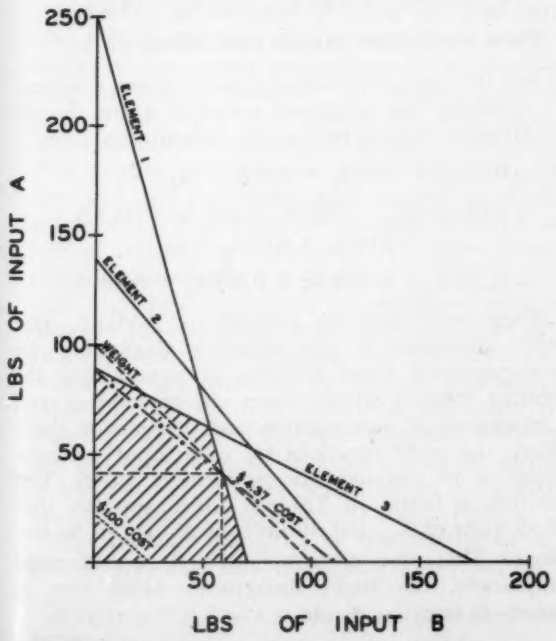


Fig. 2 — Geometrical representation of the linear programming problem.

The dotted and dashed line is the closest to the origin which intersects feasible space, and represents a cost of $4.37. It intersects feasible space at the point (61, 39), which says that the most economical mix for the foundry to use is 61 lb of input *B* and 39 lb of input *A*. This is the graphical solution to the simplified L. P. problem.

If another input were to be considered along with inputs *A* and *B*, we would need a third axis to represent it. The problem would then be in three dimensions, and we might still be able to visualize it geometrically. If *n* inputs were to be considered however, the problem would be one in *n* dimensional space, and analytic techniques which correspond to the geometrical representation would have to be used to extract a solution. Virtually all real problems involve the evaluation of many variables, and such analytic methods must be employed.

THE SIMPLEX SOLUTION

For the interested reader, we will now solve the simplified problem shown above by the simplex method* an iterative algebraic procedure developed by George Dantzig.†

The algebraic statement of the problem [Equations (1) to (4)] is again listed:

$$0.02X_A + 0.07X_B + S_1 = 0.05(100)$$
$$0.05X_A + 0.06X_B + S_2 = 0.07(100)$$
$$0.08X_A + 0.04X_B + S_3 = 0.07(100)$$
$$X_A + X_B + S_4 = 100$$

The meaning of S_1, S_2 and S_3 we are familiar with; they are slack variables designed to allow the final contents of elements 1, 2 and 3 to be less than the maximum limit imposed on these final contents. S_4 is what is called an artificial slack variable. Expression (4) is not an inequation, so although for computational purposes it is desirable to include with it a slack variable, this slack will not be allowed to appear in the final solution (by associating with it a high cost, M).

The equations as listed are put in matrix form, as shown in Table 6. The coefficients of the slacks are 1, and the initial solution is that indicated by the symbols in the first column. This first solution indicates that slack is used to the exclusion of products. This is

TABLE 6

Initial Solution	INPUTS						Amount in Solution
	S_1	S_2	S_3	S_4	X_A	X_B	
S_1	1				0.02	0.07	5
S_2		1			0.05	0.06	7
S_3			1		0.08	0.04	7
S_4				1	1	1	100

*See Bowman and Fetter, *Analysis for Production Management*, Richard D. Irwin, Inc. (1957).

†G. B. Dantzig, "Maximization of a Linear Function of Variables Subject to Linear Inequalities," appearing in Ref. 5.

TABLE 7

		INPUTS							
	Profit:	0	0	0	−M	−5	−4		
Profit	Initial Solution	S_1	S_2	S_3	S_4	X_A	X_B	Amount in Solution	Determination of Smallest restriction
0	S_1	1				0.02	0.07	5	/0.07 = 71.4
0	S_2		1			0.05	0.06	7	/0.06 = 116.6
0	S_3			1		0.08	0.04	7	/0.04 = 175
−M	S_4				1	1	1	100	/1 = 100
		0	0	0	−M	−5	−4		
						−0	−0		
						−0	−0		
Profit						−0	−0		
Function:						−(−M)	−(−M)		
						+M	+M		

TABLE 8

	Profit:	0	0	0	−M	−5	−4		
Profit	Solution	S_1	S_2	S_3	S_4	X_A	X_B	Amount in Solution	Determination of Smallest restriction
−4	X_B	14.3				0.286	1	71.4	/0.286 = 250
0	S_2	−0.86	1			0.0328		2.72	/0.0328 = 83
0	S_3	−0.57		1		0.069		4.14	/0.069 = 60
−M	S_4	−14.3			1	0.714		28.6	/0.714 = 40
		0	0	0	−M	−5	−4		
Profit		−14.3(−4)				−0.286(−4)			
Function:		−(−14.3)(−M)				−0.714(−M)			
		−M				+M	−4		

TABLE 9

	Profit:	0	0	0	−M	−5	−4	
Profit	Solution	S_1	S_2	S_3	S_4	X_A	X_B	Amount in Solution
−4	X_B	−20	0	0	1.4	0	1	59.95
0	S_2	0.2	1	0	−0.4	0	0	2.59
0	S_3	0.43	0	1	−0.046	0	0	3.86
−5	X_A	−2	0	0	−0.097	1	0	40
Profit Function:		−90	0	0	−M	0	0	

equivalent in the geometrical representation, to starting at the origin.

The proper input to be brought in must now be determined. Below each column in Table 6 is shown the profit (costs are negative profits) to be realized for each unit of input brought into solution. The coefficients in Table 6 represent rates of substitution between the various inputs. So, for example, the profit for each unit of X_A brought in is its profit per unit, −5, less the profits lost when substituting X_A for each of the inputs in the present solution, the slacks.

This is $-5-[-0.02\times0-0.05\times0-0.08\times0-(-M)]$, equals approximately $+M$ (since M is a large cost and outweighs the other factors). X_B has a similar net profit per unit of $+M$, so it is immaterial at this stage whether X_A or X_B is introduced into solution. X_B was chosen. Because we are dealing with linear relationships (there is no falling off of utility when introducing X_B), we bring in as much X_B as is technically feasible. If we divide each capacity restriction by the substitution rates of the product going in, we get limits to the number of units of X_B which can go into solution.

The smallest capacity restriction is the largest number of units which can be brought in, and we substitute X_B for that product in solution which offers the restriction. In the right hand column of Table 7 we see that 71.4 is the smallest capacity restriction. What we have just done is equivalent in the graphical solution to moving out along the abcissa (B axis) from the origin until the first capacity restriction is reached.

The other products no longer have their original number of units in solution. We derive the new solution by multiplying the number in the present solution by the proportion of the capacity remaining:

$$S_2 = 7 \times \frac{116.6-71.4}{116.6} = 2.72$$

$$S_3 = 7 \times \frac{175-71.4}{175} = 4.14$$

$$S_4 = 100 \times \frac{100-71.4}{100} = 28.6$$

These new values are shown in the solution column of Table 8. In Table 8, the new substitution rates (coefficients in the Table) must be calculated, because we have new row headings (the inputs in the new solution), and the products which can be brought into solution must be expressed in terms of this new set of variables. The procedure is as follows:

1) Express the rates of substitution from Table 7 for the product brought into the new solution:

$$X_B = 0.07\,S_1 + 0.06\,S_2 + 0.04\,S_3 + S_4$$

2) Solve for the product being replaced (S_1).

$$S_1 = 14.3X_B - 0.86\,S_2 - 0.57\,S_3 - 14.3\,S_4$$

These are the new rates of substitution for S_1.

3) For the new substitution rates of other products, substitute the calculated values of S_1 in the expressions describing the old substitution rates.

$$\begin{aligned} X_A &= 0.02\,S_1 + 0.05\,S_2 + 0.08\,S_3 + S_4 \\ &= 0.02\,(14.3X_B - 0.86\,S_2 - 0.57\,S_3 - 14.3\,S_4) \\ &\qquad + 0.05\,S_2 + 0.08\,S_3 + S_4 \\ &= 0.286X_B + 0.0328\,S_2 + 0.069\,S_3 + 0.714\,S_4 \end{aligned}$$

These new rates are inserted in Table 8. The entire procedure is now repeated: Evaluating the most profitable input to bring in, determining the limiting product which comes out of solution and computation of new solution and substitution rates. When the profit functions for each input are computed to be negative, the problem is solved. The solution is found in Table 9 which specifies that 59.95 units of X_B, and 40 units of X_A are to be employed. This solution corresponds to the one obtained graphically, the slight discrepancy being due to round-off errors.

COPPER IN CAST IRON

principal considerations

By Albert de Sy

ABSTRACT

Copper is a graphitizer at the stage of solidification as well as an anti-ferritizer through transformation range cooling. Because of this, copper additions to gray cast iron prevent thin section chilling and formation of free ferrite in heavy sections. A uniform structure is thus imparted to gray iron by the copper additions which has neither hard nor soft spots, and has good overall strength and hardness properties.

Thus, copper additions offer a distinct advantage in relation to machinability and wear resistance which are important for the many applications and engineering properties of gray iron castings. Also, compared to other alloying elements, copper is relatively inexpensive.

I. THEORETICAL CONSIDERATIONS

It may be said that the art of using basic or general scientific knowledge is to analyze complicated matter systematically, and to look at it carefully in the simplest possible way. This should undoubtedly apply to gray cast iron, which is a complicated alloy, mainly because of the superposition of two crystallization systems and owing to the allotropic transformation of iron.

Moreover, any element added to the basic ternary alloy Fe-C-Si affects the relative importance of the two crystallization systems and influences the transformations on cooling. Therefore, any element, even when present only as a trace, affects the final structure and, consequently, the physical properties of the alloy. The physical properties are determined by the structure, and thus it is important to analyze the structure and structure modifications resulting from the addition of an alloying element such as copper, which is to be considered here in particular.

In studying the structure of a multiphase alloy it is desirable to proceed in two parallel but slightly different ways, viz., by considering the different constituents on the one hand and the different phases on the other hand. Apart from inclusions, and even apart from iron phosphide and free cementite which will not be considered here, gray cast iron is an aggregate of the constituents pearlite, free ferrite and graphite; or of the phases ferrite, cementite and graphite.

The physical properties of any multiphase alloy, and of gray cast iron in particular, depend on the physical properties of the different phases present, their amount, the form and size of the particles of each phase and on the distribution of the different particles and their association in the so-called constituents.

Thus, it is evident that the study of the effect of copper, or of any other element on the properties of gray cast iron, requires a careful examination of the effect of that element on the above mentioned factors of influence. In reviewing the various copper effects on gray cast iron, it is desirable to distinguish between its direct effect (effect on the properties of the individual phases) and its indirect effects (those concerning the other factors of influence).

Copper Direct Effect

Copper has practically no direct effect on the properties of the graphite phase in which it is insoluble, and may only be present in graphite as a few entrapped atoms. Moreover, the physical properties of graphite compared with those of the metallic matrix are negligible. This also applies to its effect on the physical properties of cementite. Copper is not a carbide-forming element, and its partition coefficient is such that only a few copper atoms can be dissolved in cementite whose physical properties thus remain practically unaffected.

The physical properties of the ferrite phase, by far the largest in amount, on the contrary are affected greatly by copper additions. Ferrite is hardened by the normal solution hardening in any case, and eventually, in a second step by structural hardening.

Figure 1 gives indications of the solution hardening effect of copper, according to A. Kussman and B. Scharnow,[1] compared with that of several other common elements, according to E. C. Bain.[2] Thus, it appears that copper has a strong hardening effect. Figure 1 as such does not apply for the ferrite phase of cast iron, because the effect of the various elements is not simply additive. Moreover, although elements such as Cu, Si and Ni are entirely dissolved in ferrite this is not so for the carbide forming elements Mn, Mo, V and Cr, which dissolve partly in ferrite and partly in cementite.

The direct effect of copper on gray cast iron thus is limited to its hardening effect on the ferrite phase. On this basis alone, it can be expected that copper additions to gray cast iron will increase both hardness and strength characteristics.

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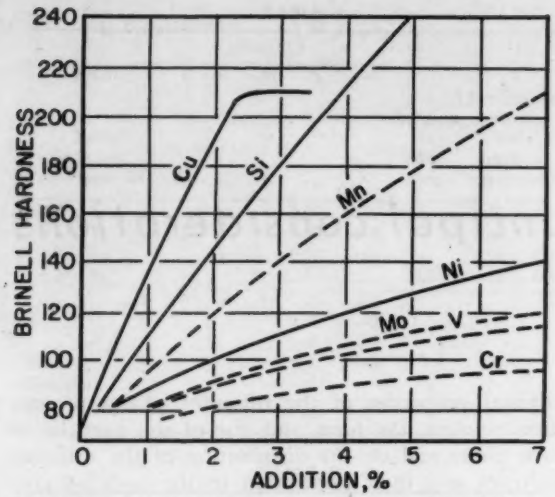


Fig. 1 — Brinell hardness of α — Fe and solid solution hardening effect of various elements.

Copper Indirect Effects

When considering the indirect effects of copper we have to deal with all other factors of influence, viz., the influence of copper on the amount of the different phases, its influence on the form, size and distribution of the different phase particles and its influence on the association of phase particles in constituents. All these influence factors have to be studied in close correlation to the formation and transformation of the different phases. Consequently, it is necessary to review these copper effects in intimate correlation to the mechanism of solidification and of transformation in the solid state.

At the Solidification Stage: Copper Is a Mild Graphitizer. This effect is well known. According to technical literature, the graphitizing effect of copper is estimated between ⅓ and ⅕ that of silicon. According to the author's experience, it should be rather close to ⅕ that of silicon. Consequently, copper is a milder graphitizer than nickel whose graphitizing ratio is ⅓-¼ that of silicon.

The graphitizing effect of Si, Ni, Cu . . . is believed to be in relation to atom prearrangements in the melt just above and at the solidification temperature. As these elements have no tendency to form carbides they increase the carbon activity of the melt through a dilution effect. Moreover, Si because of its affinity for Fe has a tendency to form Fe-Si prearrangement in the melt. Thus, C atoms tied up by Fe atoms are liberated and the carbon activity of the melt is once more increased.

However, apart from its graphitizing effect, copper also has a structural refining effect. Figures 2 and 3 show the influence of copper on the structure of the chilled zone in rollers and wheels according to P. Van Bleyenberghe and R. Coussement.[3] This effect is of course beneficial, since hardness is lightly increased and brittleness is considerably decreased.

Copper Refining Effect

The copper refining effect on graphite is less apparent because of the difficulty in estimating it in the case of flake graphite. However, it is distinctly apparent in the case of spheroidal graphite. The refining effect of copper on cementite (and ledeburite) and graphite is an indirect result of an adverse influence of copper on the carbon diffusion rate, and consequently a decreased growth rate of carbon (graphite) or carbon-rich (cementite) crystals. Smaller individual crystals necessarily mean a larger number.

It is not known, at least to the author's knowledge, whether copper has a direct influence on graphite nucleation or (very unlikely) on cementite nucleation. *Transformation in the Solid State: Copper is a Powerful Anti-ferritizer, and Consequently an Equally Powerful Anti-graphitizer.* On the cooling of gray cast iron from solidification to room temperature free ferrite formation may occur in two different ways:

1) Direct ferritizaton according to the reaction
austenite ⟶ ferrite + graphite
or precipitation of ferrite from austenite and also precipitation of the corresponding stoichiometric carbon which crystallizes on the already existing graphite crystals.
2) Indirect ferritization according to the reaction
austenite ⟶ pearlite ⟶ ferrite + graphite
or ferritization and graphitization as a result of pearlite decomposition.

The influence of copper on the first stage reaction



Fig. 2 — Cu = 0%; Rc = 55-57.

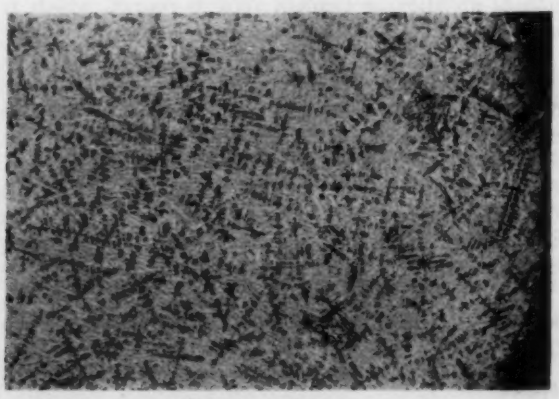


Fig. 3 — Cu = 1.5%; Rc = 60-65.

Figs. 2 and 3 — Copper effect on structure of chilled cast iron. 50 ×.

(direct ferritization) has been carefully studied by A. de Sy, J. Vidts and J. Van Eeghem,[4] and by A. de Sy and J. Foulon.[5]

Figures 4 through 7 show some suggestive results of these investigations. They present a good idea of the anti-ferritizing power of copper as compared with nickel and with tin. Figure 4 shows the structure of a spheroidal graphite iron as-cast in ¾-in. diameter section. It is a high carbon, low manganese and, for that type, a rather low silicon iron. However, the matrix is at least 70 per cent ferritic.

The large ferritic bulls-eyes undoubtedly result from the direct ferrization-graphitization reaction. In order to comparatively study the anti-ferritizing power of Ni, Cu and Sn, various amounts of these elements were added to identical melts prior to the magnesium treatment which was done with a magnesium-ferrosilicon alloy.

Nickel additions of 1.6 per cent, 3.2 per cent, 4.8 per cent and 6.4 per cent were successively made, and none of these additions succeeded in eliminating the ferrite bulls-eyes. Figure 5 shows the result obtained with a 1.6 per cent Ni addition. The result obtained with the 4.8 per cent Ni addition was equivalent to the 0.2 per cent Cu addition. The 6.4 per cent Ni addition resulted in a more or less acicular matrix structure, but still did not completely eliminate the ferritic bulls-eyes.

Thus, it appears that nickel acts as a mild anti-ferritizer. Copper, on the contrary, is a more powerful anti-ferritizer. Copper additions of 0.2 per cent, 0.8 per cent, 1.6 per cent and 3.2 per cent were made and, as shown Fig. 6, the 0.8 per cent copper addition was sufficient to eliminate the bulls-eyes and give a fully pearlitic matrix. Thus, it appears that copper is a powerful anti-ferritizer.

As is well known, high copper contents have an adverse influence on spheroidal graphite formation. It is worth while stating that the 3.2 per cent copper addition completely destroyed the graphite spherulites.

In the same series of experiments the fully pearlitic matrix was obtained with a 0.06 per cent Sn addition, resulting in a 0.03 per cent residual tin content. This powerful anti-ferritizing tin effect has been recently confirmed by J. A. Davis, D. E. Krause and H. W. Lownie.[6]

Of course, the anti-ferritizing effect of copper, tin and other elements is not limited to spheroidal graphite iron. A. de Sy and J. Foulon[5] have shown that it also applies to flake graphite iron, as confirmed recently by J. A. Davis, D. E. Krause and H. W. Lownie.[6] There is only a quantitative difference. Flake graphite irons necessitate larger amounts of anti-ferritizing additions because of the much larger graphite-austenite interface surface.

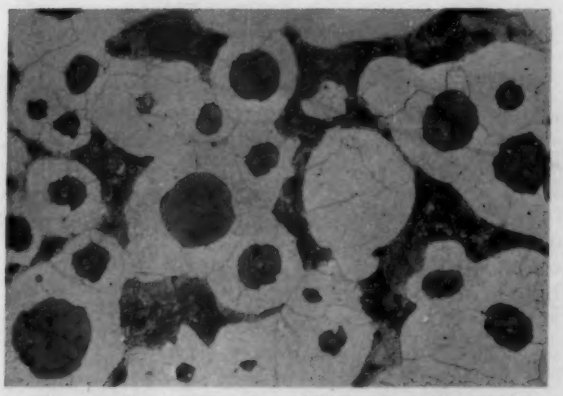


Fig. 4 — C — 3.93%; Mn — 0.27%; Si — 2.02%; Mg — 0.051%.

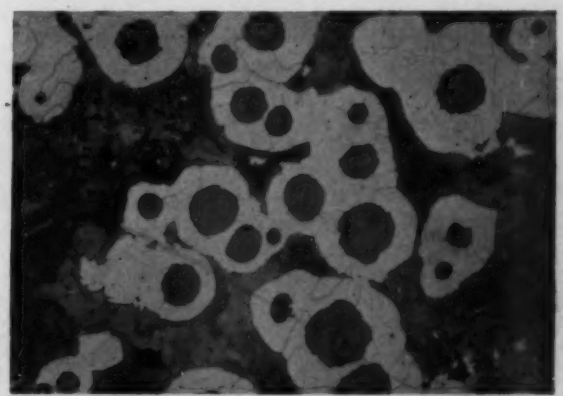


Fig. 5 — Same as Fig. 4, with addition of 1.6% Ni.

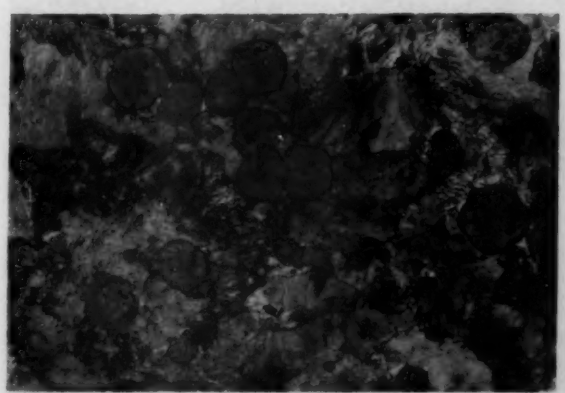


Fig. 6 — Same as Fig. 4, with addition of 0.8% Cu.

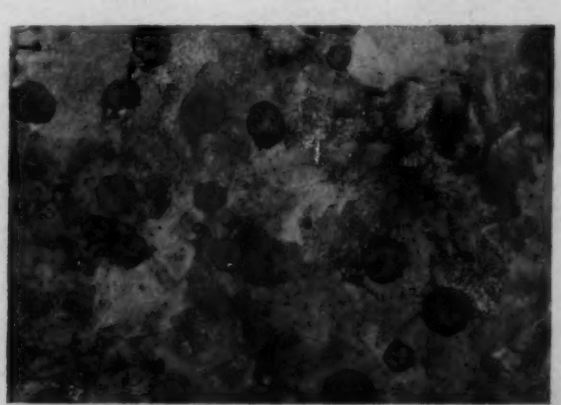
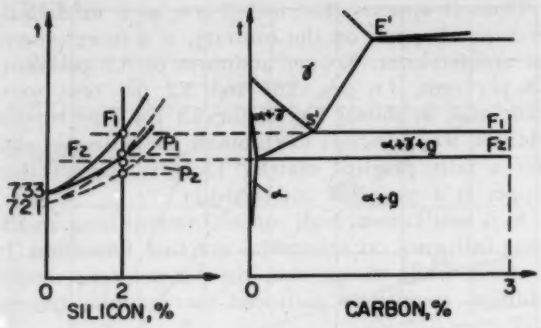


Fig. 7 — Same as Fig. 4, with addition of 0.03% Sn.

Figs. 4, 5, 6, and 7 — Anti-ferritizing power of copper as compared with nickel and tin in a spheroidal graphite iron. 200 ×.



Figs. 8a (left) and 8b (right) — Fe-C-Si system, 2% Si (8a) and 3% C (8b) sections, according to Hanemann and Schrader.[7]

The author does not claim to have discovered the anti-ferritizing effect of copper, only, perhaps, the term. This effect of copper has long been known, but it is stated in the wrong way and generally misunderstood. Indeed, in many technical papers it is said that copper is a pearlite stabilizer, and that is fundamentally wrong. However, it is correct to say that copper is a pearlite formation promotor, as may be easily understood from the study of the mechanism of austenite transformation on cooling.

Figures 8a and 8b represent, respectively, the 3 per cent C and the 2 per cent Si sections of the Fe-C-Si system, according to H. Hanemann and A. Schrader.[7] From Fig. 8a it appears that the direct transformation of austenite to ferrite and graphite may proceed in the temperature range comprised between the F curves (from F_1 to F_2 at 2 per cent Si). If an iron sample is held in this temperature range, transformation will proceed until equilibrium, provided the holding time is sufficiently long.

The mechanism of this direct transformation is illustrated by Figs. 9a, 9b and 9c, in which graphite is supposed of the spheroidal type for schematic representation facilities only. The reaction invariably starts at the existing graphite-austenite interfaces (Fig. 9a). This results in graphite deposition on the already existing graphite crystals, and in the formation of ferrite layers tending to separate the graphite crystals from the austenite.

Figure 9b represents a stage where the graphite crystal is already completely surrounded by a ferrite layer. From this stage on any further progress of the reaction not only implies ferrite precipitation at the austenite-ferrite interface, but also carbon diffusion from austenite through the ferrite layer to the graphite crystal.

The transformation of, say 100 g austenite containing 1 per cent carbon, results in the formation of approximately 99 g of ferrite and 1 g of graphite. Direct ferritization and corresponding graphitization are stoichiometrically bound. The former is impossible without the latter. The mechanism of this reaction being now clearly understood, it is worth while examining how the presence of copper is able to retard this reaction.

As is well known, copper is an austenite stabilizer. Therefore, as the ferritizing reaction proceeds further, copper must concentrate in austenite as, for an analogous reason, silicon concentrates in ferrite. Thus, increasing copper concentration in austenite corresponds to increased austenite stability, and consequently to a lower reaction rate.

However, the retarding effect based only on austenite stabilization is probably unimportant, since nickel has an equal or even higher austenite stabilizing effect but does not substantially retard direct ferritization.

Diffusion Barrier

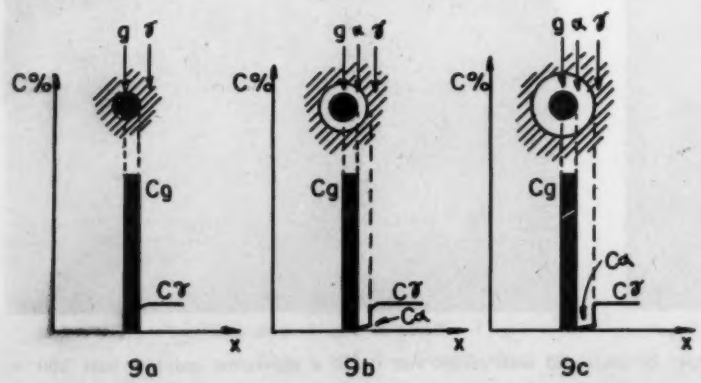
Therefore, it is believed that copper, contrary to nickel, diffuses slowly in austenite, thus accumulating in austenite at the austenite-ferrite interface, to create a diffusion barrier. This theory which may probably be checked in the near future by x-ray diffraction or x-ray fluorescence gives, at least, a plausible explanation of the anti-ferritizing influence of copper, and perhaps also of tin and some other elements.

If, as is conclusively proved by the investigations of A. de Sy and J. Foulon[5] and of A. de Sy, J. Vidts and J. Van Eeghem,[4] copper retards or prevents direct ferrite formation during the cooling of gray cast iron, it thus indirectly promotes pearlite formation in the temperature range between the P curves (from P_1 to P_2, Fig. 8a).

Indeed if austenite does not transform according to the stable equilibrium system, it must necessarily transform according to the metastable equilibrium system in a slightly lower temperature range.

Thus, copper is a strong pearlite promoter, which is quite different from saying that copper is a pearlite stabilizer.

In concluding the study of the copper-graphite



Figs. 9a (left), 9b (center) and 9c (right) — Schematic illustration of the mechanism of direct transformation of austenite to ferrite and graphite.

correlation in gray cast iron it may undoubtedly be stated that:

1) Copper is a mild graphitizer at the solidification stage.
2) Copper is a powerful anti-graphitizer (which means the same as anti-ferritizer) on cooling through the allotropic transformation range.
3) Copper slightly refines the graphite.

To review completely the copper effect on structure, and consequently on the properties of gray cast iron, it is also necessary to examine its influence on the structure of the constituent pearlite.

As already stated, copper is an austenite stabilizer, from which it automatically follows that the pearlite formation temperature is lowered and the pearlite structure must be finer. That already formed pearlite is stabilized by the presence of copper appears unlikely, but accurate data on this point are not yet available. In reviewing, on the basis of the theoretical considerations given above, the overall effect of copper on gray cast iron, it may be stated that:

1) Copper, being a mild graphitizer, at the solidification stage, tends to decompose or to prevent the formation of ledeburitic carbides.
2) Copper, because of its anti-ferritizing influence during cooling through the transformation temperature range, promotes pearlite formation and, because of its austenite stabilizing influence, refines the pearlite.
3) Copper is a powerful hardener of the ferrite phase, which is the main phase representing over 80 per cent in weight.

These structural effects must necessarily correspond to an overall increase in hardness and strength of the alloy without creating hard spots, and without a material reduction in machinability. Moreover, it may also be anticipated that copper additions will result in an important decrease in the so adverse section sensitivity of gray cast iron. To what extent these various structural effects may improve the physical properties must be determined experimentally.

II. EXPERIMENTAL RESULTS

In earlier literature dealing with the influence of copper additions on the physical properties of gray cast iron some contradictory results are reported, and even contradictory conclusions are drawn. However, this is not surprising as many investigators have tried to study the influence of copper by comparing irons of different origin and even different analysis, or even different casting sections.

A good and almost complete review on the effects of copper on the properties of gray cast iron appears in the publication *Copper as an alloying Element in Steel and Cast Iron* by C. H. Lorig and R. R. Adams.[8] The experimental results reported by these investigators are those from almost identical irons, except for the copper content. Such results are readily comparable and, if experimental conditions were kept constant, as is practically certain in the cases referred to, they may be considered reliable.

The results reported by C. H. Lorig and R. R. Adams show that copper increases tensile strength, transverse strength and Brinell hardness, but lowers deflection. The beneficial effect of copper additions on strength and hardness properties is proved up to copper contents of 3-3.5 per cent, which corresponds to the "practical solubility" of copper in gray cast iron.

More recently these conclusions of C. H. Lorig and R. R. Adams have been confirmed by A. de Sy and J. Foulon,[5] who studied the influence of copper on the basis of 140 experimental melts carefully controlled under laboratory conditions. Phosphorus and sulfur were kept constant while copper, carbon and silicon varied between the extreme limits:

Cu : 0 to 4.8%
C : 2.76 to 3.53%
Si : 1.48 to 2.04%

Manganese was kept at 0.50 per cent in one series and at 0.90 per cent in the other series.

The numerous values of the tensile strength determinations have been used in a broad statistical study, the results of which appear in Fig. 10.

Thus, it clearly appears that tensile strength increases with copper content up to approximately 3 per cent. The same conclusion applies for transverse strength and Brinell hardness.

Experimental Verification of Quantitative Influence of Copper on Gray Cast Iron Physical Properties

In recent experiments at the author's laboratory, the quantitative influence of copper additions to gray cast iron has been checked once more, and this time with the maximum possible accuracy.

In order to study the specific influence of copper, identical charges, except of course for Cu content and corresponding balanced Si content, of accurately constant raw materials were melted in a high frequency induction furnace according to a normal and constant melting procedure, and cast at constant temperature in 1¼-in. and ⅝-in. diameter test bars.

Tensile strength and Brinell hardness values were

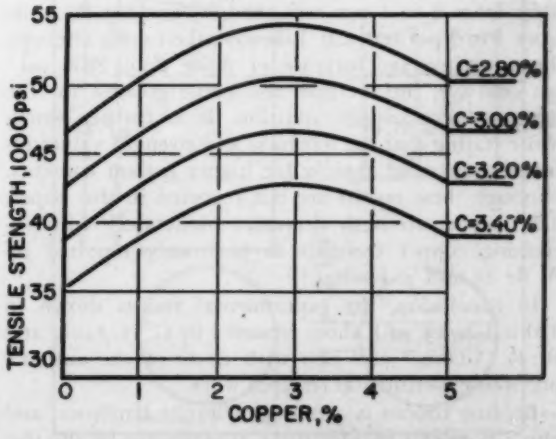


Fig. 10 — Tensile strength vs. copper content for different amounts of carbon. Test bar — ¾-in. diameter, as-cast. Si — 1.8 to 2.0%; Mn — 0.50%; P < 0.10%.

TABLE 1 — RESULTS OF TENSILE STRENGTH AND BRINELL HARDNESS TESTS

Melt No.	Composition, % C	Mn	Si	Cu	Ni	S	P	Brinell Hardness 1¼-in. dia.	5/8-in. dia.	Tensile strength, 1000 psi	Chill depth, mm	20 mm Impact, kgm,
S-11	3.20	0.9	1.86	—	—	0.065	0.10	187	246	39.0	13	2.5
								181	234	39.0		2.6
								187	222			
								av. 185	av. 234	av. 39.0		av. 2.55
S-13	3.16	0.88	1.75	0.50	—	0.065	0.10	200	257	39.9	12	2.7
								198	257	42.4		2.3
								195	256			2.3
								av. 198	av. 257	av. 41.2		av. 2.43
S-14	3.16	0.90	1.60	0.91	—	0.065	0.10	205	255	42.3	10	2.4
								198	263	42.2		2.2
								210	262			2.4
								av. 204	av. 260	av. 42.3		av. 2.33
S-15	3.19	0.91	1.54	1.93	—	0.065	0.10	224	281	44.7	5	2.2
								219	278	45.1		2.0
								214	268			2.0
								av. 219	av. 275	av. 44.9		av. 2.07
S-16	3.17	0.89	1.35	2.96	—	0.065	0.10	229	270	46.0	4	2.0
								227	263	46.0		2.2
								231	265			2.0
								av. 229	av. 266	av. 46.0		av. 2.07
S-17	3.15	0.88	1.52	—	1.82	0.065	0.10	207	260	42.9	9	2.4
								204	253	43.9		2.3
								207	246			2.2
								av. 206	av. 255	av. 43.4		av. 2.30

determined according to the recommendations of the ISO/TC 25 Committee. Chill depth was measured on the Le Thomas test piece, and impact strength was determined by the Charpy method on a 20-mm diameter unnotched test piece with 100-mm distance between supports.

The results of this investigation are shown in Table 1.

From the description of the experimental procedure, and from the results of the chemical analysis of the different melts, it appears that the experimental accuracy is largely sufficient to permit drawing valuable conclusions as to the quantitative influence of copper additions on the physical properties of gray cast iron.

For this medium carbon iron cast in 1¼-in. diameter cylindrical bars, Brinell hardness increases by 20 units from 0 to 1 per cent, and by another 20 units from 1 to 3 per cent Cu. Likewise, the tensile strength shows an average increase of more than 2000 psi/per cent Cu, the increase also being greatest for the first per cent copper addition. It is further worth while stating that the hardness and strength values increase even more rapidly for higher carbon contents, although these results are not reported in this paper. The impact strength decreases moderately with increasing copper contents as previously reported by A. de Sy and J. Foulon.[10]

In conclusion, the experimental results shown in Table 1 agree with those reported by C. H. Lorig and R. R. Adams,[8] and also with those of the author's previous experimental research work.

Because copper is a powerful ferrite hardener, and also a powerful anti-ferritizer as explained in the theoretical part of this paper, it may logically be expected that copper additions will also substantially increase hardness and strength properties in heavy sections.

Beneficial Effect of Copper in Heavy Section Castings

The increase in hardness, as well as the increase in uniformity of hardness distribution, has been reported by R. H. McCarroll and J. L. McCloud[9] on iron of constant Si content and variable Cu content cast in pyramid-shaped castings. One should, however, compare irons of different Cu contents but with constant carbon equivalent. This has been done at the author's laboratory by keeping total carbon constant and varying the Si content in order to balance it against the Cu contents. The carbon content chosen was 3.20 ± 0.03 per cent, and all melts with a carbon content out of this range (C analyzed on a chilled sample) were eliminated.

The carbon contents, together with the results of a complete analysis (except P which is: 0.05-0.07 per cent), are shown in Table 2. Extreme carbon limits are thus 3.19 and 3.23 per cent.

Meltings were done in a high frequency induction furnace under constant metallurgical conditions with the same raw materials, and the pouring temperature was also kept constant at 2460 F. Two test bars 1¼-in. diameter, and one cylindrical bar of 7 in. x 12 in. were cast from each melt.

The mechanical properties were determined as recommended by the ISO/TC 25 Committee, employing the diabolo-shaped, or type A tensile test piece, machined from the 1¼-in. diameter test bars and from test bars extracted from the 7 in. cylindrical bar, as shown in Fig. 11. The complete results of the mechanical tests are shown in Table 2, in which all Brinell hardness figures are the averages of 4 determinations.

TABLE 2 — CARBON CONTENT AND COMPOSITION ANALYSES FOR TEST BARS

Melt	Composition, %								7 in. dia. Bar: Tensile strength, 1000 psi					7 in. dia. Bar: Brinell Hardness, H_B					1¼-in. dia. Bar	
No.	C	Si	Mn	Cu	Ni	Cr	Mo	S	o	m	c	m	o	o	m	c	m	o	TS, 1000 psi	H_B
1	3.22	1.66	0.82	0	0	0	0	0.065	28.3	23.9	22.7	24.4	29.6	164	163	156	165	167	40.7	193
2	3.19	1.49	0.78	0.94	0	0	0	0.064	31.0	28.0	24.4	25.9	33.0	167	167	171	167	168	42.3	211
3	3.20	1.39	0.82	1.48	0	0	0	0.067	33.0	28.3	27.3	28.6	33.5	186	181	178	180	186	46.9	225
4	3.21	1.28	0.82	2.02	0	0	0	0.065	32.8	28.1	26.9	28.3	33.7	184	178	178	179	186	44.5	224
5	3.20	1.24	0.81	2.03	0	0.23	0	0.063	37.3	32.7	29.6	31.8	36.9	198	196	194	195	194	46.5	229
6	3.24	1.27	0.82	2.02	0	0	0.24	0.064	36.9	30.6	27.4	31.3	38.0	190	182	182	183	191	47.4	232
7	3.22	1.32	0.83	2.04	0	0.25	0.24	0.064	40.4	34.2	31.0	34.0	38.7	210	199	195	197	208	50.1	253
8	3.22	1.19	0.82	0	1.93	0	0	0.065	29.4	25.7	24.4	25.7	29.4	170	166	165	167	174	42.8	213
9	3.23	1.19	0.82	0	1.93	0.24	0	0.064	32.7	28.4	26.6	27.3	32.8	180	176	172	174	180	46.3	224
10	3.23	1.17	0.83	0	1.95	0	0.23	0.063	32.5	26.7	26.3	26.2	32.3	178	168	169	164	176	46.2	241
11	3.23	1.16	0.86	0	1.96	0.23	0.23	0.067	39.7	34.0	31.7	33.7	39.5	198	188	187	183	195	54.0	253

In comparing the results from melts 1 to 4, it appears that tensile strength and hardness increase regularly with increasing copper content. Considering the properties in the 1¼-in. diameter test bar, Brinell hardness increases by 30 points, and tensile strength approximately 4500 psi in passing from 0 to 2 per cent Cu, all other possible influencing factors being kept constant. These results confirm those of the first experiments shown in Table 1.

More interesting are the results obtained in the heavy section castings. Here again, when Cu passes from 0 to 2 per cent Brinell hardness is increased by 20 points and tensile strength by approximately 4500 psi. Moreover, the hardness is uniform over the section, with only a few points variation (lower) in the central part.

Hardness vs. Casting Section

A good general survey of the influence of copper on hardness increase, and uniform hardness in thick sections, is given in Fig. 12. Figure 12 shows hardness values of melts 1 to 7 plotted against casting section. The same figure also shows the influence of supplementary additions of either 0.25 per cent Cr or 0.25 per cent Mo, or both (Cr and Mo additions are without correction of the carbon equivalent, melts 5, 6, 7 and 9, 10, 11).

A remarkable composition is that of melt 5. The hardness in the heavy section is almost perfectly uniform and high, nearly 200 Brinell. The alloy addition of 2 per cent Cu combined with 0.25 per cent Cr is believed beneficial for uniform high mechanical properties combined with good machinability, because of the absence of hard spots if chromium is limited to 0.3 per cent maximum. To the author's knowledge this alloy addition is applied successfully for diesel-engine cylinder liners.

Melts 8, 9, 10 and 11 almost exactly duplicate melts 4, 5, 6 and 7, but with nickel instead of copper, the carbon equivalent being maintained constant by lowering the silicon contents.

Table 2 and Fig. 13 show the comparative results of copper and nickel additions. Copper gives distinctly better results than nickel, especially in heavy section castings where hardness is not only high but also uniform.

The comparative results of the strength properties of copper and nickel alloyed irons cast in a thin and a heavy section are shown in Fig. 14.

CONCLUSIONS

Because copper is a graphitizer at the solidification stage, and an anti-ferritizer on cooling through the transformation range, copper additions to gray cast iron prevent chilling in thin sections and free-ferrite formation in heavy sections. Thus, copper additions impart a uniform structure with neither hard nor soft spots, and at the same time remarkable overall strength and hardness properties.

This is a unique advantage in relation to machinability and wear resistance, which are both important for many of the practical applications and the engineering properties of gray iron castings.

Moreover, compared with other alloying elements whose influence is far from being so beneficial in relation to the most important engineering properties of gray cast iron, copper is a relatively inexpensive alloying element.

Therefore, it is difficult to understand why copper additions to gray cast iron are not used more extensively (this remark applies especially to Europe

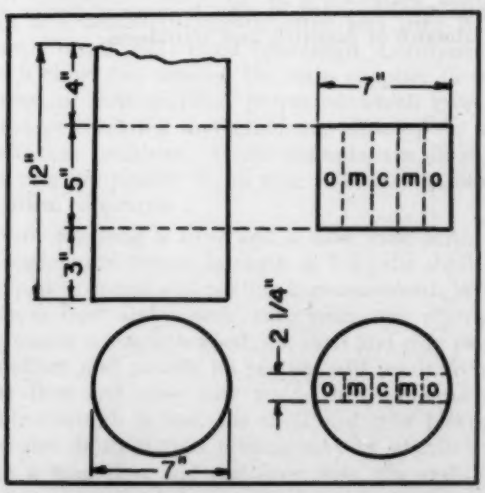


Fig. 11 — Method of extraction of test bars for tensile strength and Brinell hardness determinations from cylindrical bar 7x12 in.

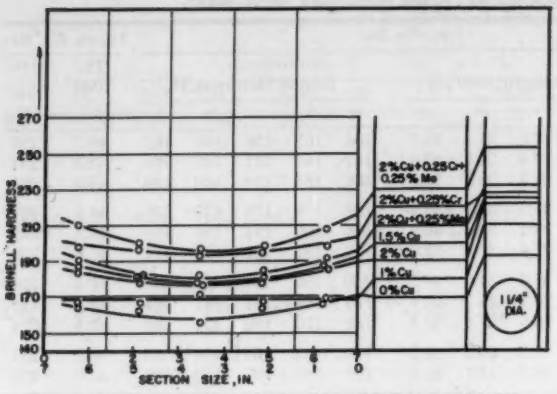


Fig. 12 — Copper effect on hardness increase and hardness uniformity in heavy section gray iron castings.

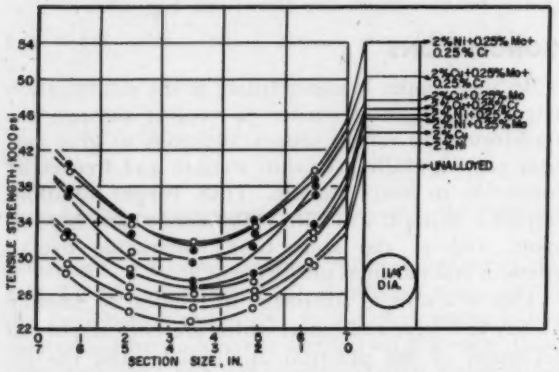


Fig. 14. — Comparative tensile strengths of copper and nickel alloyed cast irons in heavy and thin sections.

where, till recently, copper additions were practically unknown in the gray iron foundries).

At present, many foundrymen are beginning to make use of the important possibilities of quality improvement through copper additions to cast iron. Those foundries using this method which are known by the author obtain excellent results, not only from the metal quality point of view, but also from the castings quality point of view because of better density (absence of porosity) and soundness.

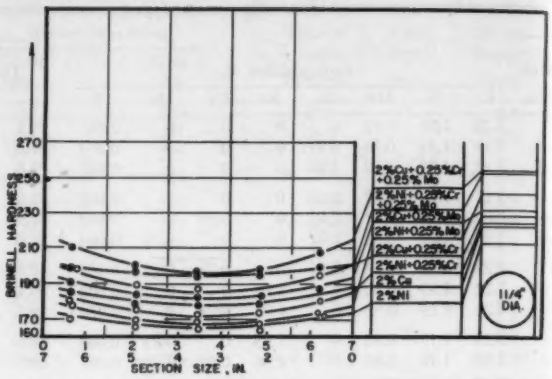


Fig. 13 — Comparative effect of copper and nickel on hardness increase and hardness uniformity in heavy section gray iron castings.

REFERENCES

1. A. Kussman and B. Scharnow, "Uber Härte und magnetische Eigenschaften von Eisen-Kupferlegierungen und ihre Beziehungen zum Zustandsdiagram," *Zeitschrift für Anorganische und Allgemeine Chemie, 178,* pp. 317-324, 1929.
2. E. C. Bain, *Function of the Alloying Elements in Steel,* A.S.M., Cleveland.
3. P. Van Bleyenberghe and R. Coussement, "Improvement of Cast Irons and Steels by the Addition of Copper," Paper 22, 25th International Foundry Congress, Liége, 1958.
4. A. de Sy, J. Vidts and J. Van Eeghem, "Le mécanisme de la ferritisation de la fonte grise et l'étude de son ralentissement par les éléments Sn et Cu," *La Fonderie Belge,* pp. 196-207, Dec. 1954.
5. A. de Sy and J. Foulon, "La suppression systématique de la ferrite de formation eutectoïde (fontes à graphite sphéroïdal et fontes à graphite lamellaire)," *La Fonderie Belge,* pp. 187-195, Dec. 1954.
6. J. A. Davis, D. E. Krause and H. W. Lownie, Jr., "Tin as an Alloy in Gray Cast Iron," AFS TRANSACTIONS, *vol. 65,* p. 590, 1957.
7. H. Hanemann and A. Schrader, "Die Gefügeumwandlungen beim Glühen von Gusseisen," *Archiv für das Eisenhüttenwesen,* H. 12, 12 jg, pp. 603-606, 1938-39.
8. C. H. Lorig and R. R. Adams, *Copper as an Alloying Element in Steel and Cast Iron,* McGraw-Hill Publishing Co. Ltd., London.
9. R. H. McCarroll and J. L. McCloud, "Ford Alloy Castings," *Metal Progress, vol. 30,* 1936.
10. J. Foulon and A. de Sy, "L'influence du cuivre sur quelques propriétés des fontes grises," *Fonderie,* 96, Jan. 1954.

NEW FOUNDRY RESINS AND APPLICATION TECHNIQUES FOR SHELL MOLDS AND SHELL CORES

By William C. Capehart

ABSTRACT

Resins used in the foundry shell mold or shell core processes are either of the pulverized or liquid type. These are applied by the dry blending and coating processes. The operational problems which are encountered when using shell molds and shell cores are peel back or drop off, poor casting finish, caking of coated sand and thermal cracking. Corrective measures can be taken to overcome these difficulties. Control tests on the resin-sand mix assure the uniformity of the mixes used in production. Tensile strength, flexural strength, melt point, cure rate and pickup test are used.

INTRODUCTION

The use of resin products in foundry technology has become an accepted casting tool. Last year U.S. foundries used an estimated 12-14 million lb of resins in the production of shell molds and shell cores. Most observers expect the 1959 total to go considerably higher. Hand in hand with this growth has come the rapid development of improved resins and improved application techniques. It seems worth while to examine some of the more recent advances to see how the ferrous and non-ferrous foundry may derive the most benefit.

RESINS

Resins are conveniently classified in two broad groups—"thermoplastic" and "thermosetting." Thermoplastics may be softened by heat and then harden into shape upon cooling. This process can be repeated almost indefinitely, as these plastics undergo physical and not chemical changes. Thermoplastic resins include such natural products as wood rosin, and synthetics such as polystyrene, polyethylene and polyvinyl chloride. The thermosetting types will harden or set into a permanent shape when heat is applied and undergo physical as well as chemical changes; thus, the process is irreversible.

Thermosetting resins include such products as phenol-formaldehyde, urea-formaldehyde, epoxies, polyester and melamine-formaldehyde. The resins used in the foundry today are combinations of chemicals reacted to form long chains of molecules linked together with little if any branching or crosslinking or partially cross-linked polymer chains, both of which form a three dimensional network design upon cure. Resins melt with heat and then cure to form a rigid bond between the sand grains.

The phenol-formaldehyde resins are used most often. Such resins are of two classes; "one stage" and "two stage." The one-stage type is made with a specifically designed phenol and formaldehyde ratio, and reacted in the presence of a catalyst. The resin as discharged from the kettle is thermosetting or heat reactive and requires only further heat to complete the reaction to an infusible, insoluble state. These resins are used for conventional core binders and some hot coating operations.

The two-stage type also is made with phenol and formaldehyde reacted in the presence of a catalyst. However, the specific ratio of ingredients are reacted such that upon discharge from the kettle they have formed a linear thermoplastic type polymer commonly called a "novolak." To produce a thermosetting resin, additional formaldehyde and heat must be added. Formaldehyde is usually furnished by the breakdown of such a material as hexa-methylenetetramine ("Hexa") when heat is applied. These resins are stable in storage even when pulverized with the hexa or placed in alcohol or solvent solutions.

Two-Stage Resin

The foundry trade is familiar with the pulverized resin compound made up of a two-stage phenolic resin and hexamethylenetetramine and used in dry powder blend shell mold operations. Continued research effort has enabled the resin supplier to select a range of resin physical properties which give consistent performance and eliminate most of the shell production problems. As an example, the flow and cure of a dry powder blend type resin can be selected as critical properties.

With too long a flow and a slow cure resin, the resin-sand mix tensile strength is low, the shell and core pickup is fast and peeling is encountered. With a medium flow and a slow cure resin, the resin-sand mix tensile strength is good, the shell and core pickup is medium and usually no peeling will result. With a short flow and slow cure resin, the resin-sand mix tensile strength is low, the shell and core pickup is slow and delamination peeling may be encountered. With a long flow and fast cure resin, the resin-sand mix tensile strength is low, the shell and core pickup is low and delamination peeling may result.

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With a medium flow and fast cure resin, the resin-sand mix tensile strength is good, the shell and core pickup is medium and some delamination peeling may be encountered. With a short flow and fast cure resin, the resin-sand mix tensile strength will be low, the shell and core pickup is slow and delamination peeling may result. These relationships formed the basis for many of the present day dry powder resins. There are exceptions to the above rules of thumb brought about primarily by the demand for higher production rates and other resin considerations.

Coating Resins

Both one-stage and two-stage phenolic resins are used for shell molding, shell cores and as bases for coating resins. There are several types of coating resins, such as alcohol solutions of novolak resins, powdered novolaks blended with hexa and wax for use with alcohol and water, crushed lump novolaks, water borne novolaks and one stage resins.

Alcohol solutions of novolak resins, usually in the 60-80 per cent solids range, are one of the most prominent coating resin forms. These resins require the addition of hexa and wax during the coating cycle. Liquid-type resins are attractive because they offer such benefits as bulk handling and metering, a lack of resin loss by dust during mixing and with a very efficient use of the resin and in handling the coated sand, the resin film has good resistance to abrasion from the sand.

The powdered novolaks containing hexa and wax are used with alcohol and water solvent systems for coating in what is commonly called alcoholic deposition. These have such benefits as controlled hexa, wax and resin ratio weighed as one unit, volumetric use of solvents and possibly 1-2 min faster coating cycle than liquids.

Most of the foundrymen are well acquainted with the types of phenolic resins just discussed. A new addition to the coating resins field, however, is the novolak or two-stage resin supplied as a rough crushed lump to be melted on hot sand. Considerable research work has been done to arrive at a resin which melts readily to give controllable mixing properties and have adequate tensile and hot casting strength.

These resins require the use of hexa and wax during the coating cycle. Lump resins have such qualities as a fast coating cycle, nonflammable solvents and little, if any, caking of the coated sand.

Water-Borne Casting Resins

Another new type of system is the recently developed water borne coating resins. These are divided into two classes. The first is the one stage resin which requires no crosslinking agents such as hexa, but react rapidly when heated. However, in some cases, hexa is used to speed up resin coating and cure. These resins are usually stored under refrigeration between production and use to extend their life.

In the coating process they are mixed to the desired melt point and tensile strength properties with heat, and the coated sand cooled rapidly for good control of these properties. Some of the benefits of the one-stage resins are; a fast cure, and a nonflammable solvent system.

The careful controls in storage, mixing and handling of the resin and coated sand made with the one-stage type resins has made the idea of a novolak resin in a stable water system attractive. These resins, as do all novolaks, require the addition of hexa, usually a powdered or dissolved form and a wax as desired. These stable water borne novolaks are desirable because of the nonflammable solvent system.

Water borne novolaks may be used with hot sand or hot air and have a fast coating cycle. The coated sand is nonblocking and stores easily, and a versatile range of resin properties is available to meet varied production requirements. There is no resin storage problem, and preliminary indications are that they are more resistant to thermal cracking than other types of coating resins.

In the use of coating resins, the relationship of various resin physical properties and the coated sand performance is similar to those of a dry blend powdered-type resin. The relationships are easily noticed in the hot coating process where the resin characteristics are changed during coating. These changes are manifest by such resin-sand mix tests as melt point, tensile strength, flexural strength, rate of cure and pickup thickness.

With two-stage resins some empirical relationships have become evident such as; when the coated sand melt point is raised, the tensile strength and flexural strength is lowered and the shell or core pickup thickness is less. At certain high melt point levels, a delamination peeling condition can be incurred. This is brought about because the resin flow is shortened and the resin cure is faster. When the coated sand melt point is lowered, the tensile strength and flexural strength is raised and the shell or core pickup thickness is increased. This is due to the resin flow being lengthened and the cure being slower.

Also, when the bake time (cure cycle) of a shell or core is lengthened, the tensile strength and flexural strength increases at a fairly rapid rate and reaches a maximum after which the resin degrades reducing the tensile strength and flexural strength. These relationships point up the importance of processing the resin to reach optimum performance characteristics for production conditions.

Cold or Warm Coating Resins

In cold or warm coating-type resins, the same relationships hold as for hot coating resins. However, since the resin characteristics are not changed during the coating process, the supplier provides resins of various properties to fit a wide range of desired shell or core operating conditions. There are exceptions to these general empirical relationships discussed, and illustrated in Fig. 1, because many different types of resins are made and new developments are anticipated in resins, coating processes and machinery for processing and shell and core production.

RESIN APPLICATION TECHNIQUES

Each foundry must select the resin that is the most economical and gives the best castings for its operation and, therefore, should study resins in light of present requirements and future outlook.

The application technique of resin in shell molding or shell cores has changed considerably in the past several years, and in coating one might say in the past few months. The dry powder blending with a powdered resin compound was the original method of dispersing the bond and still is continued profitably in many foundries. As the sand companies learned more about shell molding, they have produced special sands which have reduced the resin content needed for dry blending to a more economical level.

The wide acceptance of the dry powder blend technique is evident because no special equipment is required that is not already in the foundry, and it is fairly foolproof once a study has been made to determine the time necessary for efficient resin dispersion. Due to the simplicity of this technique, the foundry just entering the shell mold field finds that it eliminates one of the problems in consistent shell production at a time when new process problems are at their greatest.

The resin properties available cover a wide range of shell applications. Dry resins have a tendency to dust, but a number of satisfactory dust suppressants are available.

The foundryman's desires to produce more complicated castings in shell, get higher production rates, to improve his working conditions and to use blowing techniques for shells as well as cores, have given rise to the coated sand procedures. Various techniques have evolved starting with a conventional powdered shell resin and a solvent, and progressing to special resins for each coating technique.

Methods of Coating Sand

There are three broad methods of coating sand which involve the placement of resin on each sand grain. These methods are classed as cold, warm and hot. Cold coating entails the use of room temperature sand and room temperature air. Cold coating can be performed in most conventional foundry mixing equipment except for cement or cone blenders. Alcohol solutions of novolaks and powdered resin compounds used with alcohol and water will perform well with this method.

The typical coating procedure for a liquid resin calls for catalyst and lubricant placement in the mixer and dry blended 10-20 sec with the sand. Then the resin is added and the mixture blended until it is wet. At this time the room temperature air is fed in. As mixing is continued, the resin-sand mix becomes a doughy mass called a ball stage. With continued mixing and solvent evaporation this dries out and breaks down. The time to reach dryness depends upon the amount and type of resin used, the mixer, and the type of sand.

Use of a powdered resin with a solvent requires a similar operation, except that the dry blending of the resin and sand takes 30 sec to 2 min for adequate resin dispersion. Then the solvent is added and mulled until the mix is wet. From this point the operation is the same. Some of the reasons for cold coating in addition to the regular coated sand benefits are; limited capital investment and fewer mixing variables to control.

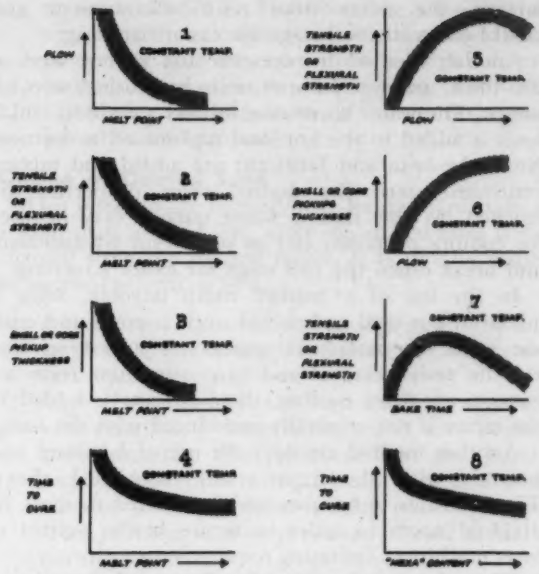


Fig. 1 — Conventional two-stage phenolic resin physical and empirical relationships.

Warm coating usually involves the same type of mixing equipment as for cold coating, but a modification has to be added to circulate warm air up to about 225 F through the muller. Air up to 450 F may be used for short periods provided the resin is not advanced appreciably. The degree of advancement or change can be measured in terms of such tests as melt point, tensile strength and flexural strength. The same types of resins and procedures used in cold coating apply to warm coating.

A short cooling cycle may be required to arrive at about room temperature coated sand before use or storage. Warm coating has the added benefits of reducing the cold coating mix cycle, increasing the coating capacity over cold coating because of less ball stage and producing less caking of the coated sand. In the cold and warm coating processes there is essentially no change in resin characteristics during the coating operation, as only the solvent has been removed leaving a resin film on the sand.

Hot Coating Sand

Hot coating involves the three basic techniques of much hotter air or hotter sand, or both. In hot coating, the resin is advanced and its characteristics are changed substantially with heat (220-500 F). A measure of the change is noticed by the melt point rise when samples are taken at various time intervals during coating. This enables the foundry to have one resin perform on many different types of molds or patterns simply by selecting the desired resin advancement to give optimum performance.

One method of hot coating employs sand at room temperature and hot air. After a short blending of the catalyst and lubricant, the liquid novolak resin or a water borne resin is added. When the resin has wet the sand, hot air is fed in until the desired resin melt point is reached at which time a water quench may be used for cooling purposes. With additional

mixing, the water stops resin advancement and breaks down the ball stage for easier handling.

Another method involves the use of hot sand at 225-400 F, and water borne resins or crushed novolak lumps. The water borne mix follows a cycle in which resin is added to the hot sand and mixed to wetness. Next, the hexa and lubricant are added and mixing continued until the desired resin properties are reached. At this time a water quench may be used for cooling purposes, and to stop resin advancement and break down the ball stage for easier screening.

In the use of a crushed lump novolak, resin is added to hot sand and mixed until it melts and coats the sand. A water-hexa quench is added to stop volatile resin removal and give minimum resin advancement. After cooling, the lubricant is added in the mixer if not originally introduced with the lump.

Another method employs the use of hot sand and hot air with the three resins mentioned above. These various techniques may be altered to meet individual needs in order to secure better control or meet machinery operating requirements.

These hot coating processes have such benefits as a fast mix cycle, little, if any, caking of the coated sand in storage, nonflammable solvents when water borne or lump novolaks are used, a hard resin film that is resistant to abrasion from the sand in handling and the opportunity to tailor the resin properties to meet a wide range of shell mold or core conditions.

Each foundry must study the various resin application techniques in light of economy, performance, as well as present and future requirements.

BOND VARIABLES

In coating sand, as in any bonding process, there are several variables which influence the bond strength. The first consideration is the sand used. The round grain sands have less surface area than subangular sands and need less resin for comparable strengths. However, round grain sands may cause more thermal cracks than sharp sands due to grain expansion rates in casting certain type metals.

The finer sand grains give more surface area for resin to cover and requires higher resin contents. Although finer sands give smoother finishes on castings, to conserve resin each foundry should use the coarsest sand possible to produce the desired finish. In some cases coarse sands may be offenders in resin abrasion from coated sand during the mixing and give erratic coating results.

Clean sand, which is low in organics and clay, require less resin. Also, organics, clays and oxides found in some sands give a cushioning effect on certain type castings to reduce thermal cracking. Therefore, a blend of two sands or the use of additives sometimes is desirable.

The bond in resin-sand mixes is developed between resin films on the sand, and, in turn, resin to sand. Thus, it is apparent that wetability of the sand grain by resin, and the bond between resin and sand, is important. Therefore, sands which have thin films of such things as clay, carbon or alkalies, may give poor strength and consequently, low strength mixes.

In the introduction of any new process, there are some variables and problems encountered which demand a solution in order to have consistent production. Efficient production depends upon the foundryman's ability to determine the source of trouble and arrive at a solution in a reasonable time. It may be of interest to define some typical problems and discuss possible causes and corrective measures.

SHELL MOLD AND SHELL CORE OPERATIONAL PROBLEMS

Peel Back or Drop Off

Peel back occurs when a shell mold or shell core pulls away from the pattern or mold face at certain sections, usually around the edge or on projections. The sections which have pulled away from the pattern lay back on the mold and remain attached to shell or core. Drop off occurs when parts or all of the shell or core completely falls from the pattern or box. This may also occur as chunk drop off. Possible causes and corrective measures are:

1) *Resin Sand Mix Improperly Made.* Check ingredient amounts and mixing procedure. If the mixer is used for other foundry cores be sure it is free of core oils. Run at least two to three cleanout batches.
2) *Wet Sand.* With the moisture above 1 per cent, sand will blend or coat poorly and retain part of the water. The cure will be slowed down so the shell will not stay on the pattern. Use dry sand below 0.5 per cent moisture. With the hot sand technique a wet sand may be used and dried as it is heated.
3) *Pattern or Core Box Temperature too High.* When the temperatures of patterns or box are too high, they cause a buildup of extra thick shells or cores which may peel or drop off because of the heavy weight. To correct this, lower the mold temperature and/or speed up the investment cycle.
4) *Pattern or Core Box Temperature too Low.* When pattern temperatures are too low, the shell or core may not be cured sufficiently to stay in place after the roll over. To overcome this condition increase the mold temperature or preheat the mold.
5) *Dwell Time too Long.* With long dwell time, the shell or core may build up too thick and peel or drop off the pattern or mold. Decreasing the dwell time gives a thinner shell or core.
6) *Sticking to Side of Dump Box.* The sides of the dump box at the pattern contact point may get hot and cause sticking and peel around the edges. Silicone rubber and/or a water jacket around this location will eliminate the problem. Also, pattern border outlines will give a clean break.
7) *Not Enough Mix in Dump Box.* Small amounts of mix in the dump box give poor compaction, and high projections will peel or drop off. A heavy-burden drop and good coverage eliminates this condition.
8) *Solvent Not Removed From Coated Sand Sufficiently.* Solvents remaining in the coated mix may give off gas which results in slow cure and

peeling or drop off. To correct this, screen and aerate more effectively or adjust the coating cycle.

9) *Pattern or Core Box Design.* Patterns with high projections which cause large temperature variations tend to cause peeling. Insert heaters or special heating relieves this. Aluminum patterns of certain design may lose heat rapidly to cause peeling. Dissimilar pattern materials may cause peeling or sticking. Consider pattern heating problems when designing patterns.

10) *Resin Content of Mix too High or too Low.* High resin content may cause too heavy a build-up on the pattern or mold and produce peeling. Too low a resin content produces weak shells or cores which peel or drop off.

Poor Casting Finish — Rough Surface

1) *Resin Sand Mix Improperly Made.* Improperly coated or blended sand may have resin rich lumps and give poor fillout of shells or cores. The fine grain sand may be lost as small lumps when the coated mix is screened, and, therefore, give rough surface finish.

2) *Sand with Large Amounts of Coarse Screen Material.* The casting finish is dependent upon the sand grain size used in the core or shell. The finer sands give smoother finishes but require more resin.

3) *Pattern or Core Box Temperatures too High.* High pattern or core box temperatures overcure the shell or core and may produce washes or sand inclusions due to a weakened shell or core. Use a constant temperature that gives consistently cured shells or cores.

4) *Too Little Resin in Mix.* Low resin content tends to cause washes, out of dimension products, cracking and metal runouts.

5) *Turbulent Metal Flow, or Lack of Effective Choke Feeding.* To produce castings with smooth surfaces, more devices must be used to assure smooth metal flow and adequate fill of the cavity. Metal pouring range and fluidity of the metal determine the shell rigging. Most metals take on a better finish if pouring is on the cool side.

6) *Poor Resin Mix Application to Pattern or Mold.* Poor fillout of cores is the result of unsuitable blowing pressure, venting and sand flowability. Poor fillout in shells comes from light drop of mix, air trapped by resin-sand mix as it falls over the pattern or slow drop of mix. Adjust blowing pressure and vents to each mold. Add more mix to the dump box, vent the pattern area and speed up box roll-over. Use a vibrator on the box or mold to increase the density and remove trapped air.

7) *Hard Pouring of the Mold.* Pour steadily enough for good fill but not too hard to cause metal turbulences, aspiration or excessive head pressure in the cavity.

8) *Poor Shell Bonding.* Bond wherever possible on the shell to prevent formation of fins, cope floating or strains. Use as many springs as possible on the glue press over the bonding area. Bond while the shell is hot. Its temperature should be at least 275-375 F. Each bonding resin has an optimum shell temperature for assembly. Bond before the glue sets up onto a thick bond, preventing mold closure and causing fins. The liquid-type bonding resins or slurries do a good job when used properly.

Caking of Coated Sand

This refers to the formation of lumps, hard balls or a solid mass a short time after the coated sand is discharged from the coating mixer. It is usually caused by residual solvent or heat.

1) *Poor Solvent Removal.* Screen and aerate the mix until no green strength is present, or adjust the mix cycle or conditions.

2) *High Humidity Day.* On high humidity days solvent removal is more difficult and may require more aeration than is needed in less humid weather. Gradual solvent reduction is recommended when powdered coating resins are used. This problem is more prevalent in cold or warm coating processes.

3) *Resin Reacted by Heat.* Long mixing cycles in warm coating may overheat the coated sand and partially cure the resin-sand mix. Warm coated sand temperatures above 170 F may cause trouble. The inclusion of a cooling cycle in mixing or proper aeration after screening will relieve this problem.

4) *Hot Storage.* Hot storage area, or failure to reduce hot coated sand below 140-160 F, can tackify the resin and induce caking.

5) *Resin Sand Mix not made Properly.* Review the coating procedure and be sure of ingredient amounts.

6) *Wet Sand or Excess Quench Water.* Excess moisture in sand is extremely hard to remove in cold or warm coating and can make the resin sticky and lumpy. If excess quench water is used in hot coating this should be reduced gradually. Moisture may be condensed on the coated sand during handling, so check at various points to determine this location.

7) *High Clay Content of Sand.* High clay content makes solvent difficult to remove and can result in solvent caking.

8) *High Sand Fines.* Sand fines form small lumps which may retain solvent and cause caking.

9) *Excessive Additives.* Excessive additives such as plasticizers or green strength additives cause caking in storage or processing.

10) *Hot Air Turned on too Soon.* In warm coating, hot air may cause encasing of the solvent by a resin film. As this solvent percolates through, it produces caking. To correct this condition, adjust the mix cycle or reduce the air temperature.

Thermal Cracking of Molds or Cores

Most shells crack at many places as the binder burns out. However, the metal has usually solidified sufficiently to prevent run-outs. Thermal cracks usually occur less than 25 sec after the metal has been poured, and may give metal run-outs or casting fins. Thermal cracks in most cases come at stress points in

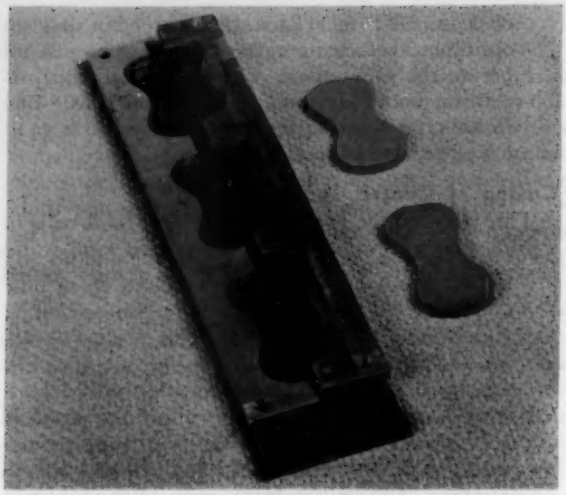


Fig. 2 — Tensile strength control test for resin-sand mixes utilizes the standard AFS recommended ¼-in. briquette. After the mix has been placed in the 400 ± 10 F mold, the strike-off is made in both directions from the center of the longitudinal axis.

the shell. Possible causes and corrective measures for this condition are:

1) *High Resin Content.* With excess resin the shell or core cures to a rigid, brittle form that is prone to crack. To eliminate this reduce the resin content or undercure the shell or core.
2) *Sand Segregation.* When coarse sands are encountered by segregation using a high resin content, the shells or cores become rigid and brittle. If there is an excess of sand fines, the mix becomes weak and cracks easily. Periodically, sample the sand and run a screen analysis. Sands with four or five screen distribution have proved more resistant to thermal cracks than two or three screen sands.
3) *Overcured Shells or Cores.* Overcured shells or cores are rigid, brittle and may be weak. Control pattern and oven temperatures to minimize the overcure.
4) *Improper Shell Assembly.* Poor placement of the shell bonding resin may produce weak bond areas and cracks. Misaligned bonding machine pins set up stress areas that are relieved with pouring. Shells may become too rigid or cold before bonding, causing stress areas or cracks and a weak bond. High air pressure on the bonder may crack or stress the shells.
5) *Warped Shells or Cores.* Warped shells or cores set up concentrated stress areas. Here again, pouring relieves this stress. Determine the cause of warpage and take corrective action.
6) *High Ferro-Static Pressure.* Hard pouring and unsupported shells may cause cracking. Pour only hard enough to get good fill. Bed the shell or give as much support as possible to prevent cracking of the shell by weight stress.
7) *Geometric Stress Cracks.* Some casting shapes are especially likely to produce stress areas in the shell. Elimination of as many sharp corners as possible helps prevent cracks. Also, notching shells in certain places forces the cracks into less critical areas. In some shells, reinforcing rods or ribs prevent or at least limit cracks. Specially designed shell flasks or frames promote even rates of expansion and eliminate cracks or reduce them to a point at which the cracks do not produce out-of-dimension castings or fins.
8) *Gold Shells.* On certain types of castings and metals cold shells may cause thermal cracks. Pouring the shells while they are warm or torching them before pouring reduces this problem.

CONTROLS FOR RESIN-SAND MIXES

A new foundry process calls for some means of control to insure uniform production. The resin supplier uses such physical properties as melt point, flow, cure, solids content and viscosity as his control criteria for each batch of resin. Rigid specifications must be met before a shipment is made. The resin suppliers are conscious of the responsibility of product uniformity to the customer.

There are several tests which can be performed on the resin-sand mixture that do not require elaborate equipment but relate to production conditions. Such tests make it possible to compare various binders, to check the uniformity of mixes and the influence of mechanical and raw material variables. Many foundries use as many as five of these control measurements. Some of these are tensile strength, flexural strength, melt point or stick point, cure rate and pickup.

Tensile Strength

This test utilizes the standard AFS recommended ¼-in. briquette (Fig. 2). After the mix has been placed in the 400 ± 10 F mold, the strike-off is made in both directions from the center of the longitudinal axis. Care should be taken to get as smooth a surface as possible in the strike-off, and also to prevent chatter which results in uneven specimen density. The specimens are cured in an oven at 580 ± 10 F for 2 min.

Cure time must be adjusted for the type and amount of resin used, and for the blend or coating process so as to develop the maximum tensile values. Coated sand usually cures faster than dry blends. The specimen can also be made by blowing techniques and cured in the same general way.

Any suitable tester may be used to break the specimens once they have cooled to room temperature. It has been found that results often vary as much as ± 15 psi between operators. Therefore, in establishing a control range for mixes this must be taken into account. The weight of the specimens indicates mix density and flowability, and relates to tensile strength. The value of the test is limited, and must be used only as a relative comparison because the values are small, because of operator variation and because of jaw failures.

Flexural Strength

The specimen mold may be fabricated from aluminum or other suitable metal, such as cold rolled steel (Fig. 3). The molding mixture may be taken

from a production batch or compounded in the laboratory in any efficient mixer. Conditions for molding are chosen to conform approximately with those of the shell or core producing machine. The practice has been to use a mold temperature of 400 ± 10 F, and a cure time of 2 min at 580 ± 10 F.

The test pieces are molded by filling the mold cavities to overflow, then striking off the excess quickly with a metal straight edge. After the pieces are cured for 2 min, its dimensions are about 5 in. x ½-in. x ½-in. One side of the specimen is rough and non-uniform. Since the flexural strength is dependent greatly upon the uniformity of the cross-sectional area, the rough side is abraded to produce a constant thickness.

This is done by passing it under an abrasive wheel with set clearance. Each piece is then cut in half to produce two bars each about 2.5 in. x 0.5 in. x 0.33 in.

The bars are tested in any suitable device for the measurement of flexural strengths. It is important that the same face (rough or smooth) be in the same position for all tests.

The test has good reproductibility.

Melt Point or Stick Point Test

This test is used for coated sand, but is not recommended for resin-sand dry mixes. Although the resins are amorphous and have no true melting point, they do soften as heat is applied. The temperature at which the resin film on the coated sand mix softens sufficiently to adhere to a plate is recorded as the melt point. A ground steel bar (11¾ x 2⅜ x 10⅛ in.) may be heated to provide a temperature gradient by means of two 50 watt, 115 V, bayonet heaters. Two holes are drilled in one end of the bar to receive the heaters. A temperature gradient of 125-250 F is obtained by incorporating a power reostat in the system (Fig. 4).

To determine the melt point or stick point, the resin coated sand is spread to a depth of about 1/16-in. After allowing 1 min to elapse starting at the cooler end the coated sand is brushed lightly off with a camel hair brush, or may be blown off with low pressure air at a constant distance of the nozzle above the bar. The temperature of the section on the bar at which the coated sand sticks is measured with a pyrometer.

This temperature is the melt or stick point. Typical satisfactory values for core mixes vary from about 160-210 F, while values for shell mold mixes vary from about 180-230 F.

This is a fast test for determining the quality of each batch before using it in production. The test is particularly valuable in hot coating. Each resin has a melt point range in which it performs best for shells or cores. Higher than normal melt points indicate resin abrasion from the sand, or too much resin advancement. Lower than normal melt points indicate excess resin, unremoved solvent, or under-advanced resin. The melt point in hot coating is a time and temperature relationship and must be adjusted with one or the other to arrive at control values.

Cure Rate or Depth of Cure Test

An empirical comparison of cure rate can be made by reducing the cure time of a standard AFS tensile briquette and breaking it while hot. The nearer the tensile strength value approaches the maximum tensile of the specimen as run on a regular test, the faster the resin cure rate. A second method consists of dumping or blowing the coated sand into a heated conical mold and base plate at 450 ± 10 F (Fig. 5). The excess is struck off, and the specimen and mold are placed in an oven at 580 ± 10 F and baked 30 sec.

The mold and specimen are brought from the oven and the specimen ejected. This specimen is cut in half, top to bottom, and the uncured mix removed from the interior leaving a cured outer shell. The cured thickness is measured to determine the rate of resin cure or depth of cure. The thicker the section, the faster the cure.

Pickup Test

Into a heated cylinder and base plate at 450 ± 10 F, the resin mix is dumped or blown (Fig. 6). The mix

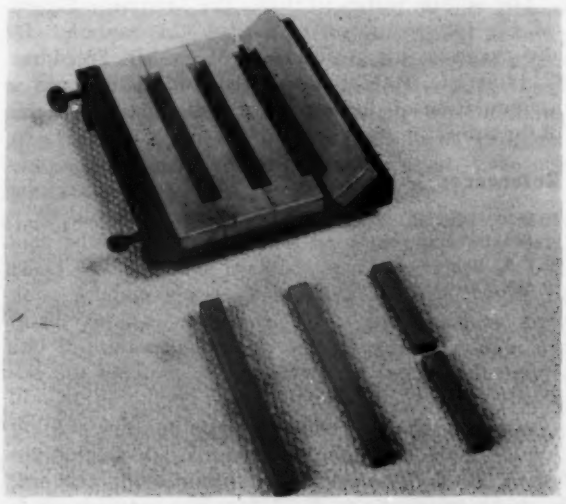


Fig. 3 — In the flexural strength test for resin-sand mixes the specimen mold may be fabricated from aluminum or other suitable metal such as cold rolled steel. Conditions for molding are chosen to conform approximately with those of the shell or core producing machine.

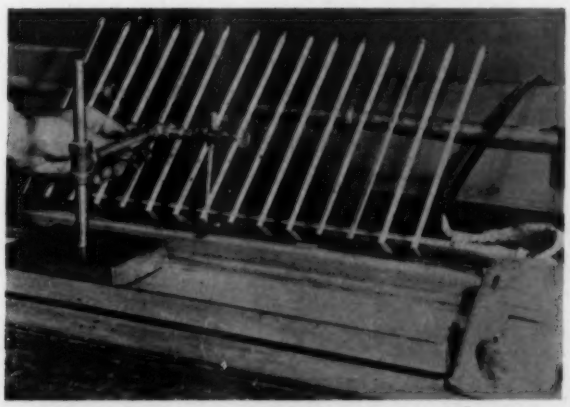


Fig. 4 — The melt point or stick point test is used for coated sand. Two holes are drilled in one end of the bar to receive heaters. The temperature gradient of 125-250 F can be obtained.

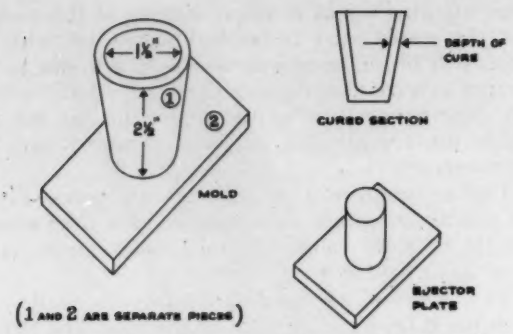


Fig. 5 — In the rate or depth of cure test illustrated the coated sand is dumped or blown into a heated conical mold and base plate at 450 ± 10 F. The excess is struck off and the specimen and mold are placed in an oven at 580 ± 10 F and baked 30 sec.

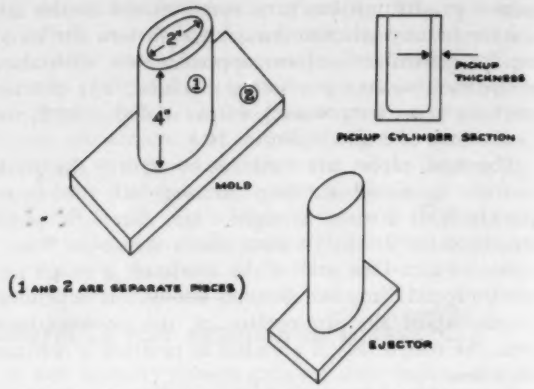


Fig. 6 — In the shell core or shell mold pickup test the resin mix is dumped or blown into a heated cylinder and base plate at 450 ± 10 F, and invested for 15 sec, and cured in an oven at 580 ± 10 F about 2 min after which the cylinder is ejected from the mold and cooled for weighing.

is invested for 15 sec and the excess is dumped. The specimen and mold are placed in an oven at 580 ± 10 F and cured about 2 min, after which the cylinder is ejected from the mold and cooled for weighing. Afterward, it may be cut in half to determine the wall thickness. This determines the investment time required, resin content or temperature needed to secure a specified wall thickness. This test also checks mix flowability and density of the mix.

There are other valuable tests such as hot deformation, expansion, peeling, hot deflection, scratch hardness, collapsibility and resin content by loss on ignition. Such tests may be of value at particular foundries where certain problems exist or a preference is indicated.

Summary

The resins used in the foundry shell mold or shell core processes are of two physical forms. These are the pulverized and liquid types. The pulverized resin forms are resin compounds and crushed lump novolaks. The liquid resin forms are alcohol solutions of novolaks, water borne solvent systems of novolaks and water borne solvent systems of one-stage resins.

The resin application techniques are dry blending and coating. The coating processes are cold, warm and hot, dependent upon the method used to place a resin film on each sand grain and degree of resin advancement. The newer coating types involve the use of hot sand or hot air or combinations of these.

Some factors that influence the resin sand bond quality are sand grain shape, grain size and contaminants.

Some typical problems in shell mold and shell core production are peel back or drop off, poor casting finish, caking of coated sand and thermal cracking. These may be encountered in using dry blends or coated sand, and corrective measures can be taken.

Control tests on the resin-sand mix such as tensile strength, flexural strength, melt point, cure rate and pickup, assure uniform mixes.

Production of satisfactory castings using shell molds and shell cores depends upon an understanding of the resins being used, the most economical resin application technique, the ability to analyze a production problem swiftly and a knowledge of proper corrective measures as well as the establishment of suitable controls to insure consistent day to day operation. Great strides have been made in all these fields, and the future holds even more rewards for the aggressively managed foundry.

ACKNOWLEDGMENT

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References

1. M. F. Drumm, "Resins and Sands for Shell Molding," AFS TRANSACTIONS, *vol. 61*, pp. 367-377 (1953).
2. J. E. Bolt, "Cold Process for Resin Coated Foundry Sands," AFS TRANSACTIONS, *vol. 64*, (1956).

THE FOUNDRY PLANT ENGINEER

a fifty year reminiscence

By James Thomson

ABSTRACT

Often, it is really worth while to take a backward look over the past to get a proper perspective of where we now stand and perhaps where we are headed. Not necessarily that we should be "proud of the good old days," but that we may have a better appreciation of progress made.

INTRODUCTION

In retrospect, let us take a brief look at the foundry industry in the early years of the present century. There were many foundries in this country—iron, steel and non-ferrous, a few large and many small, but to a great extent there was little or no mechanization. Production was mainly by manual labor. True, in some foundries there were cranes, some belt conveyors, air tools, etc., but they were more or less the exception. And this was also true in other industries. In fact, mass production, as developed by the automobile industry, had its inception in the first decade of this century.

MASS PRODUCTION ERA

For the past three or four decades American industry has been traveling along in what might be termed the "Mass Production Era." The methods, principles, organization, research, etc., developed by the automotive industry have been studied and adapted by other industries including our own, which have resulted in a tremendous increase in the productive capacity of the country. Accompanying this change in our industrial history has come a myriad of inventions, machines, equipment, materials, methods, etc. Even our educational system has been affected by it.

At the turn of the century there were few, if any, men who might be termed foundry plant engineers. In fact, I doubt if the title was even known those days. Engineers seemed to have an idea, which in many instances was true, that foundries were dirty and unpleasant places in which to work, and many did not seem to be interested. In fact, my introduction to the subject was an illustration of that attitude.

I was working on the drafting board in the design engineering department of a large machine concern. One day the Chief Engineer stopped at the board of an engineer and said, "Bill, the superintendent of our steel foundry wants a man to make a plant layout for him, and I should like you to go up and take care of it for him." Bill replied, "Chief, I would like to be excused; I don't want to work in a dirty foundry." The Chief then turned to me and asked if I had heard the conversation. I had, so he asked me if I would like to take the assignment which he guessed would take about three or four weeks.

After discussion and arrangements I told him "Yes, I should like to get some first hand knowledge of foundry work," figuring that it would be of value later in the design work we were doing.

So on the next Monday morning I arrived at the steel foundry, and the superintendent turned me over to the master mechanic who was to direct me in what was wanted.

PLANT LAYOUT

The job they wanted was a scaled layout or map of the plant buildings, and the main items of equipment therein. Some small amount of information was in the form of sketches and prints in the desk drawers and on books in the master mechanic's office which, by the way, was in the engine room of the power house, not partitioned off, just a desk and drafting board in the engine room. The remaining data had to be obtained in the field.

A couple of weeks went by and I was getting information for the map, and absorbing what I could of operations in the foundry. One day the master mechanic asked me to lay aside the layout temporarily and make some drawings for him of parts he needed to get some equipment in shape. Later there were other interferences caused by drawings needed for other things, so the three weeks job extended to three months. My chief kept asking when I would be finished at the foundry.

About this time the superintendent approached me with the proposition that he would like to have me stay with the foundry as a sort of engineering assistant to the master mechanic. He said he was convinced that the foundry needed some sort of engineering approach to some of their problems, and thought that I might fill the bill.

ATTITUDES DEVELOPED

By this time I had developed two strong attitudes, first an intense interest in the operation of that steel foundry, and second a tremendous regard for the knowledge and ability of the master mechanic. He

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was Swedish, probably thirty years my senior, a mechanic of the old school with wonderful ability and resourcefulness. I was intrigued with the idea of working with him and trying to absorb some of his knowledge and qualifications.

So I agreed to take the job if a transfer could be obtained from my chief. This then was my inception as a foundry plant engineer. The proposed original stay of three weeks lasted for ten years with that company.

As I look back over these ten years, I am somewhat amazed at the variety of experiences and opportunities which occurred. I developed into a sort of general utility man in the plant. There were, of course, the engineering jobs in collaboration with the master mechanic and the general office, such as building extensions, installation and relocation of equipment, design of special equipment, improvement of existing equipment, maintenance problems, etc. However, I sometimes got pushed into the operating departments. The superintendent would call up and say that "so and so foreman is sick (or going on vacation), will you look after his work until he returns."

Thus, I was permitted to get some acquaintance with operating problems.

PRODUCT INSPECTION

One day the plant manager called me to his office and he and the superintendent discussed with me the fact that they had never had any definite inspection of the product. What was my opinion? My reply was that I was surprised that they had not established a definite inspection program long ago. They agreed that I was right, and said that such a program was about to be started. I remarked that the inspector they appointed would have a battle on his hands, and should be responsible only to the management. They agreed and much to my surprise told me I was "it."

So the inspection department was established and, although it was a battle for a while, the quality of the finished product improved. Complaints from customers became fewer and fewer and eventually all agreed that it was a valuable addition to our operations.

One interesting problem during this period was an investigation of our power plant. In those days, central power was not as common as today, hence many plants had their own boiler and engine rooms. We had three 250 hp boilers, two engines directly connected to D.C. generators, two air compressors and various auxiliaries. Our investigation led to a vast improvement in efficiency in the power plant, and the collecting of data which proved to the company the need for additional equipment. This comprised a mixed pressure turbine-driven generator, a large compressor and auxiliary equipment.

I shall not weary you with a recital of figures about our increased efficiency, but it was outstanding enough so that our superintendent bragged about it to the other plant managers of the company. The company at that time had six plants, five of which had foundries. This led to my being loaned to these other plants for work in their power plants, and eventually to a job during my last two years with the company, in charge of power plant operations of all the plants.

Many of the problems in this work were personnel rather than equipment problems, and many incidents both sad and humorous might be told about them. For instance an old German fireman at one of the plants seemed to object to my interference with his work. He said "I can wheel more coal than any fireman in Allegheny County, and no young whippersnapper of an engineer is going to tell me how to do my job."

I replied, "Pete, the more coal you wheel to these boilers, the more ash you have to wheel out. Look at you, your face, and hands and clothes are black from so much coal and ash handling. You don't have a minute's time to rest and wash up. If I can show you how to keep up steam with less coal, wouldn't you be happy?" Pete replied, "But you can't. I've been firing these boilers for 15 years and you can't show me." Well we showed him and he was really happy, and we saved many tons of coal.

That was over 40 years ago. No doubt many changes have taken place in these plants, one of which would be a change to central station power.

In the ten years I have been talking about I received a wonderful apprenticeship in foundry plant engineering, and had come to the conclusion that there is a field of work in the foundry industry for plant engineers.

FOUNDRY ENGINEERING

One day I received a call from a man who had been manager of the above steel foundry during part of my time there, and who had left to become president of a foundry company in the middle west. He asked me to meet him in Pittsburgh. He had an interesting but discouraging tale to tell me about his plant. They could not get production problems licked because of serious mechanical and electrical breakdowns. Would I be interested in visiting the plant and considering the job of plant engineer.

To say the least, I was flattered to think that my old boss thought enough of my ability that he would call on me for help in his trouble. I therefore agreed to spend a couple of days in the plant, and would advise him later of my decision.

My visit at the plant indicated to me that the story the president had told me was undoubtedly true. The plant was far from being as well equipped as the one I was thinking about leaving. The prospect did not look inviting. So I did a lot of thinking about it. Maybe it was even worse than I had observed. Would I get proper backing from the management? Would I get along with the production departments? What kind of maintenance personnel and equipment were available? Had I the guts to fight it through against the difficulties which would arise, etc. etc?

In my dilemma I decided to talk it over with my old friend the master mechanic. After explaining the situation to him he said, "Jim, I advise you to take the job. I know it will be hard but is a great challenge and I am sure you can handle it. Here is an

opportunity to further prove your theory that plant engineers are needed in the foundry industry."

I finally decided to make the move after getting a clear understanding as to my authority, who was to be my boss, that the maintenance department was to report only to me, etc.

But what an awakening! It was far worse than had been anticipated. The plant was started in 1910, and this was 1916, and it looked like they had failed to properly maintain it. There were little or no records of equipment, practically no spare parts, no idea of preventive maintenance, few maintenance tools and machines (and these inadequately equipped) and maintenance personnel was rather poor and underpaid. As an example, they had two cranes which were identical, but instead of getting repair parts for both, one was shoved into a corner and was being robbed of parts to keep the other going.

The storeroom keeper used his ideas of what kind of materials should be bought for the maintenance department, such as oils, bearing metals and electrical materials, instead of buying what the master mechanic ordered.

What a mess I had gotten into! The master mechanic told me that when I visited the plant the manager had purposely prevented me from talking to him alone, and that if he had told me about the conditions I probably would not have come.

To say the least it was a terrible disappointment. Not only from the plant conditions, but the production department seemed to think that this "new broom" was going to quickly sweep away a lot of their troubles. However, that was impossible, and I was often the goat for things that affected production.

These experiences taught me an important lesson. Knowledge, experience and personality were not the only faculties needed. One had to add to these that important qualification—intestinal fortitude. We were in a struggle to build a better operating plant and, therefore, had to fight it through against all odds.

I might recite many incidents of this period, some almost unbelievable and some humorous, but suffice it to say that day by day and month by month things began to look up.

NEW BUILDINGS

The first world war was on and we were in it. We had to build another foundry to satisfy government demands. New buildings and equipment had to be planned and installed. These were interesting days for the plant engineering and maintenance departments. With improved and enlarged personnel and organization we were able to carry out the construction program in good shape. The demands of wartime production developed many additional problems for these departments.

Eventually the war was over and we had to resume commercial production. About this time I began to see here and there that a few foundry companies had seen the value of having a plant engineer in their organization. This made me happy, knowing that now these companies were beginning to realize the value of the engineering approach to many of their problems.

It is true that there were some engineering concerns which did some consulting work on foundry problems, but even when they were employed it became evident that when the foundry company had their own plant engineering department there existed a better means of contact with the consultants. For here were men who talked the same language and could discuss the problems from a similar background.

INDUSTRY DEVELOPMENT

The next two decades after the first world war showed a remarkable development in the foundry industry. Many new ideas showed up and much new equipment was developed. New equipment and materials companies sprung up. And may I be bold enough to assert that in the development of much of this new or improved equipment foundry plant engineers played a prominent part, even if some of the producers may not agree. Our demand for better construction, better lubrication, better dust protection, better controls and a myriad of other suggestions has made the equipment much better than it might otherwise have been.

During this period there also developed a great interest in the idea of preventive maintenance. Much has been written and discussed on this subject, and much progress has been made in some places. However, there is still much more to be accomplished.

ADDITIONAL FACILITIES

In the middle of the twenties our company was finally convinced that a machine shop for large work would be an excellent addition to the facilities. This problem was turned over to me. It was interesting not only from the standpoint of building design, machine tools and equipment, but from the fact that I had often to oppose some of the suggestions of my superiors. For instance I turned "thumbs down" on two of the locations for the new shop they suggested. I told them that they were not looking far enough into the future and that their locations prevented extensive enough enlargment.

This has been proved true as the shop has been extended until it now is over five times its original size. It is perhaps the largest miscellaneous machine shop in the middle west. Complete steel rolling mills can be produced in it.

In 1930 our company expanded by taking over two other foundry and machine companies in the East. We then had four plants. To me was assigned the job of Chief Works Engineer for all the plants. This of course added much larger responsibilities, so we organized for it by having a plant engineer for each of the divisions and located at his own plant. Thus, I had a contact man at each plant who could carry out his assignments for that plant.

Of course, you all remember the terrible days of the depression of the thirties. Most of our plans had to be laid aside. Yours truly was the complete works engineering department. He was chief engineer, blue print boy, draftsman, estimator, etc., but happy to have a job. However, things eventually got better and we gradually got back to normal.

Then we were faced with the Second World War and the foundry industry was called on to do a tremendous job. The war work assigned to our company required the expansion of our present facilities and the erection of two new plants. This was handled under the planning and superintendence of our Works Engineering Department. After the war, of course, we were faced with the many problems connected with the getting back into commercial production, and the Works Engineering Department had its share of these problems.

CONCLUSION

This then is a sort of rambling story of my connection with foundry plant engineering and I should like to conclude with the following remarks:

It has been a wonderfully satisfying and pleasing experience to watch the development of the foundry industry during these 50 years.

I am happy to note the wide acceptance among foundry companies of the idea of the "plant engineer," and to see this additional field of work which has been opened to engineers.

I am particularly pleased to note the fine work being accomplished by the Foundry Educational Foundation in interesting college students to study the cast metals industry, and, thus, provide a supply of educated men to fill important places in our industry.

Last but by no means least, I must congratulate the American Foundrymen's Society for the fine work it is doing for the industry through its Headquarters Office, Chapter Organization, officers, committee members, paper authors and others. It is constantly supplying its members and the industry with valuable information, and telling the world at large the importance of the foundry industry.

DUCTILE IRON PRODUCTION IN BASIC DIRECT-ARC FURNACE

By C. R. Isleib

ABSTRACT

Addition alloy cost savings, less reladling and a reduced tendency toward drossing are advantages gained by using a basic-lined direct-arc electric furnace for the production of ductile iron. The use of a basic desulfurizing slag during meltdown upgrades the base iron and preconditions it for the making of ductile iron at low magnesium residuals.

INTRODUCTION

Electric melting furnaces are used in many foundries today, with well-known advantages. They are becoming a part of more and more shops, either alone or as part of a duplexing operation. They steadily become more attractive costwise as electricity tends to become a comparatively cheaper source of power for melting. Atomic power generated electricity may encourage this tendency in the future.

The electric furnace can melt any kind of a scrap charge. The holding time in the furnace, and the temperatures attained, can be varied at the will of the operator. A basic-lined direct-arc electric furnace can operate with a great variety of oxidizing or reducing slags.

The production of spheroidal graphite magnesium-containing iron, commonly known as ductile iron, first of all requires almost complete desulfurization of the base iron melt. This is done either before the magnesium addition or as part of the magnesium addition. It can be done through the use of slags.

In basic electric steel practice, slags have been used for over 40 years for both oxidation and reduction of undesirable impurities. Generally, elements other than sulfur are first oxidized with a lime-iron oxide slag. Then the steel is desulfurized and deoxidized under a reducing lime-fluorspar slag.

The first published work on refining iron in a basic electric furnace, and then treating it to produce ductile iron, is that of Hignett.[1] Hignett desulfurized a molten iron bath by using a lime slag, plunged 0.2 per cent calcium metal into the desulfurized bath, and then treated the iron by adding magnesium and postinoculated it with ferrosilicon. At 0.026 per cent residual magnesium and over, a fully spheroidal structure resulted. These heats were on a small laboratory scale. These results were confirmed and commercial possibilities suggested by three tests in 1952, using a larger direct-arc furnace.

Fully spheroidal irons were produced at about 0.024 per cent magnesium. Further work was needed to define some of the reaction rates, limits of composition and properties of the product.

Since 1941 there has been much basic investigation of the principles of desulfurization of iron and steel, a great deal of it by Chipman and associates.[2] Complete and rapid desulfurization has been found to depend upon the basicity of the slag and the effectiveness of the slag-metal contact, and also upon the slag temperature and the oxygen potential of the slag in those furnaces operating under reducing conditions. Slag basicity in blast furnaces and electric furnaces is sometimes measured by adding the molar concentrations of the basic oxides, calcium oxide (CaO) and magnesium oxide (MgO), and subtracting from this the molar concentrations of the acid compounds.

A slag's oxygen potential is a measure of its iron oxide (FeO) content. Further, investigators have shown that additions of silicon dioxide (SiO_2) or manganese oxide (MnO) to slag interfere with the desulfurization reaction.

Sulfur in steels usually is not required to be brought below 0.02 per cent, but for making ductile iron lower sulfur content is desirable. A low sulfur base iron is in the best condition to react to a minimum magnesium addition.

Chipman explains that the basic desulfurizing reaction in a reducing furnace is:

CaO (in slag) + S (in metal) + C (in slag) = CaS (in slag) + CO (gas).

At equilibrium the minimum "S (in metal)" is theoretically 0.001 per cent. This served as a theoretical target for desulfurization.

EQUIPMENT AND PROCEDURE

The furnace used in this work is a basic-lined, three-phase, direct-arc furnace. It has a rammed magnesite bottom, magnesite brick sidewalls and a top of mullite shapes. It is capable of melting up to a 750-lb charge. Preheating for about 50 min by arcing onto crossed electrode ends in the bottom of the furnace preceded each heat. A few pounds of slag materials were used around the electrodes during the beginning of the heat to encourage a smooth meltdown. As the metal pool increased, more slag was added. The aim was to have a maximum proportion of lime with just enough fluorspar to make the slag workable, so that it would cover the bath surface to a depth of about 1 in.

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This encouraged a high slag basicity. A late graphite addition was made to insure a carbidic slag low in iron oxide (FeO). Silica in the form of foundry sand adhering to charged metal was avoided. The furnace lining and the charges were dry, and the charging door was kept closed as much as possible. These precautions were meant to eliminate oxygen and hydrogen as far as possible.

The calcined ("burned") lime and fluorspar used in these heats had the analyses listed in Table 1.

All furnace charges analyzed less than 0.005 per cent magnesium, and about 0.035 per cent titanium.

RESULTS

Table 2 is a summary of 11 experimental basic direct-arc furnace heats made, and the properties of the ductile irons produced.

Desulfurization in the Furnace

Desulfurization of iron in a direct-arc electric furnace was simple. It involved little more melt time or work than necessary for an ordinary meltdown at the level of 0.02-0.05 per cent sulfur being charged. Slag equal to 3 to 4 per cent of the metal weight made a ½-in. to 1 in. thick layer of fairly foamy, fluid slag at 2800 to 2900 F. A ratio of 2 lime to 1 fluorspar, all sized below ½-in. mesh, was used. Six to 8 lb of the mix were used to help at the start of the meltdown. When about 2850 F was reached, a small amount of lime or fluorspar was used to bring the slag to the proper consistency.

TABLE 1 — SLAG MATERIALS ANALYSES

Chemical Analysis, %	Burned Lime	Fluorspar	Dolomitic Burned Lime
CaO	85		57.3
MgO	13		38.3
S	0.037	0.034	0.4 (as SO_3)

Cold charges melted under this high calcium oxide lime-spar slag were desulfurized at 0.005 per cent sulfur or lower by the time they reached 2850 F. This was considered near the practical limit of desulfurization.

After the first three heats, an addition of graphite was made to the slag after meltdown. This was to insure the reducing conditions necessary for optimum desulfurization. After such an addition, quenched slag samples had a white-gray color and carbidic smell. This was used as an operating test.

Desulfurization of an already molten charge is also simple and quite fast. For example, the rate of desulfurization of a hot metal charge from about 0.024 per cent sulfur down to 0.004 per cent, is plotted in Fig. 1. A 250-lb cold charge had been melted down under 28 lb of lime-spar slag. A specimen taken for temperature and chemical analysis measured 3000 F, and analyzed 0.004 per cent sulfur. An addition of a 500 lb cupola tap measuring 2550 F, and analyzing 0.038 per cent sulfur, brought the total melt to 0.024 per cent sulfur and somewhat below 2600 F.

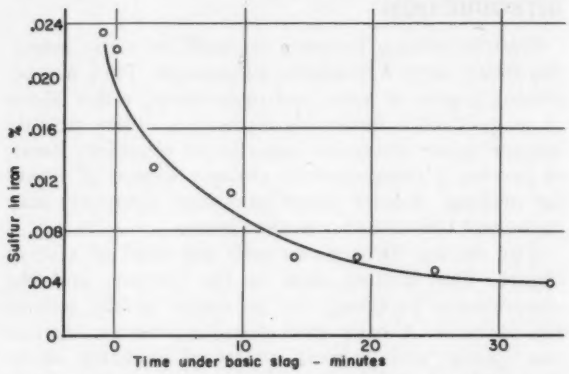


Fig. 1 — Rate of desulfurization of hot metal charge (heat 5) in the basic electric furnace.

TABLE 2 — PROPERTIES OF SOME IRONS MELTED IN THE DIRECT-ARC ELECTRIC FURNACE UNDER BASIC SLAGS

Heat No.	Pre-magnesium Treatment	Mg Treatment Alloy	Mg Treatment %	Addition Method	Chemical Content, % T.C.	Si	Mn	Ni	Mg	Postinoculation Alloy	Physical Properties TS, psi	E, %	Bhn	Graphite in 1 in. DKB Sph., %
1	None	Ni-Mg	0.38	on bath	3.6	2.0	(0.4)	(1.5)	0.023	0.5% nisiloy	105,500	9	241	90
		Ni-Mg	0.30	on bath	3.6	2.0	(0.4)	(1.5)	0.021	0.5% nisiloy	100,000	9½	237	92
		Ni-Mg	0.30	on bath	3.6	2.0	(0.4)	(1.5)	0.032	0.5% nisiloy	113,250	9	259	85
		Ni-Mg	0.30	on bath	3.6	2.0	(0.4)	(1.5)	0.016	0.5% nisiloy	68,200	2	215	80
		Ni-Mg	0.30	on bath	3.6	2.0	(0.4)	(1.5)	(0.018)	0.5% nisiloy	72,000	½	281	60
2	None	MgFeSi	3.34	on bath	(3.6)	(2.0)	(0.4)	(1.5)	0.019	0.5% nisiloy	75,000	2½	228	85
4	None	Ni-Mg	0.65	tap onto	3.6	2.11	(0.2)	(1.5)	0.029	0.75% $Fe_{85}Si$	71,900 60,750	16½ 26	181	90
		MgFeSi	1.0	tap onto	3.6	2.84	(0.2)	(1.5)	0.031	0.75% $Fe_{85}Si$	74,500 67,200	17½ 25	185	90
6	CaSi+CaC Deoxidation	MgFeSi	1.24	tap onto	3.5	2.45	(0.4)	1.18	0.013	0.6% $Fe_{85}Si$	47,400	0		75
		MgFeSi	1.61	tap onto	3.5	2.49	(0.4)	1.18	0.014	0.6% $Fe_{85}Si$	69,250	2	217	80
7	CaSi+CaC Deoxidation	Ni-Mg	1.13	on bath	3.6	2.18	(0.4)	2.31	0.049	0.6% $Fe_{85}Si$	106,500	3	277	95
		Measuring Burnout			3.6	(2.2)	(0.4)	(1.2)	0.040	0.6% $Fe_{85}Si$	93,200	5	229	90
		Measuring Burnout			3.6	(2.2)	(0.4)	(1.2)	0.023	0.6% $Fe_{85}Si$	93,400	1½	269	90
9	CaSi Deoxidation	MgFeSi	2.2	tap onto	3.56	2.45		1.01	0.051	0.5% $Fe_{85}Si$	107,700	2½	293	90
10	CaSi Deoxidation	MgFeSi	2.2	tap onto	3.56	2.60		0.26	0.058	0.5% $Fe_{85}Si$	111,250	4	293	97
11	CaSi Deoxidation	MgFeSi	2.2	tap onto	3.54	2.71		0.31	0.084	0.5% $Fe_{85}Si$	102,600	4½	256	90

By the time 2650 F was reached, desulfurization had progressed to 0.006 per cent. Since desulfurization does proceed faster at high temperatures, if the cupola metal had been hotter it is probable that the desulfurization in the electric furnace would be even quicker.

Base Irons Upgrading

Late additions of calcium silicon rabbled into the slag-covered heats, without any magnesium alloy addition, produced upgraded irons with tensile strengths up to 63,200 psi. This upgrading occurs because the late addition introduces small amounts of calcium and magnesium into the iron, partially spheroidizing the graphite. The log of one such heat is shown in Fig. 2. Late additions of calcium silicon were rabbled into the slag covered heats in all except the first four melts. These upgrading furnace treatments were done at practical working temperatures, below 2900 F.

Calcium carbide was also used as a furnace slag addition, but the evidence in Figs. 2 and 3 points to the calcium silicon as being a more effective agent in the pretreatment.

Besides upgrading the base irons, the late additions of calcium silicon to the desulfurized melts before magnesium treatment resulted in a more nearly fully spheroidal graphite ductile iron at lower magnesium residuals than without such an addition. Figure 7 shows that those heats having had a late calcium silicon addition before magnesium treatment, and then treated to between 0.01 per cent and 0.02 per cent residual magnesium, were consistently better in structure than those heats not having a calcium silicon slag addition.

Dolomitic Lime Slags Use

Iron melted under a dolomitic (high magnesium oxide) lime slag had a greater tendency toward a spheroidal graphite structure before magnesium treatment than iron melted under a high calcium oxide lime slag. Figure 3 is the log of a dolomitic lime heat and can be compared with Fig. 2, a heat under a high calcium oxide lime slag. Reports had been received of a completely spheroidal graphite iron having been made in Europe under a dolomitic lime slag simply by reducing into the iron some of the magnesium in the slag. Under the dolomitic slag more magnesium and calcium were found in the iron, and the iron contained more graphite in spheroidal form.

Magnesium Treatment Within the Furnace

After desulfurization, a small furnace addition of magnesium is sufficient to produce good quality ductile iron. Addition of ½ per cent of nickel-15 per cent magnesium alloy to a cleared spot on the bath's surface is adequate. Ferrosilicon post-inoculation is done on tapout.

Furnace addition trials were made with the alloy wrapped in nickel foil, this package being tied on a steel rod and plunged beneath the bath's surface, thrown directly on the slag blanket or thrown onto a cleared space on the metal bath surface. The third method was little trouble and most practical. Magnesium ferrosilicon gave poor magnesium recovery when used as a furnace addition (below 10 per cent in all trials).

Figure 4 shows the magnesium recovery to be expected at various addition temperatures. For consistent recovery, bath temperature must be ascertained before treatment. This can be done by an immersion couple, or an optical reading during the tapping of a small ladle of iron. A reliable optical reading cannot be obtained directly in the furnace.

After the magnesium treatment the metal is tapped at about 2750 F, and a postinoculant is added to the tapping stream. Assuming the addition was ½ per cent of nickel-15 per cent magnesium alloy at 2750 F, the ductile iron then contains about 0.025 per cent magnesium, has little tendency to form dross and is at a good casting temperature.

Magnesium burns out of the iron in the furnace at a moderate rate provided the holding temperature is not too high and no arc is used. Holding at 2700 F about 0.001 per cent magnesium burns out per min; at 2800 F about 0.002 per cent per min. Figure 5 shows an average burnout rate over a range of temperature with the power off. If treated iron has to be held in the furnace for an extended time, booster additions of the magnesium alloy can be made to make up for the predictable burnout.

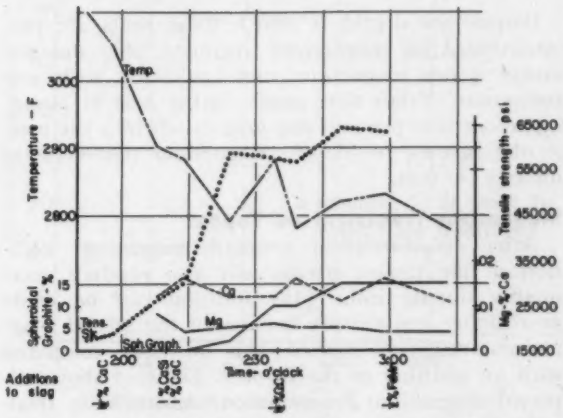


Fig. 2 — Log of heat 7, slag made using a high CaO lime.

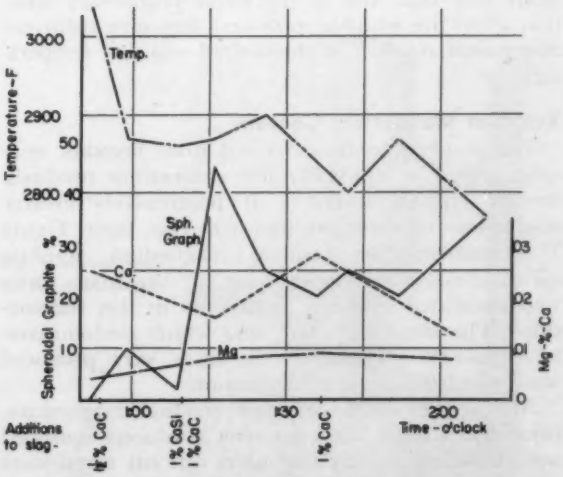


Fig. 3 — Log of heat 8, slag made using a high MgO lime.

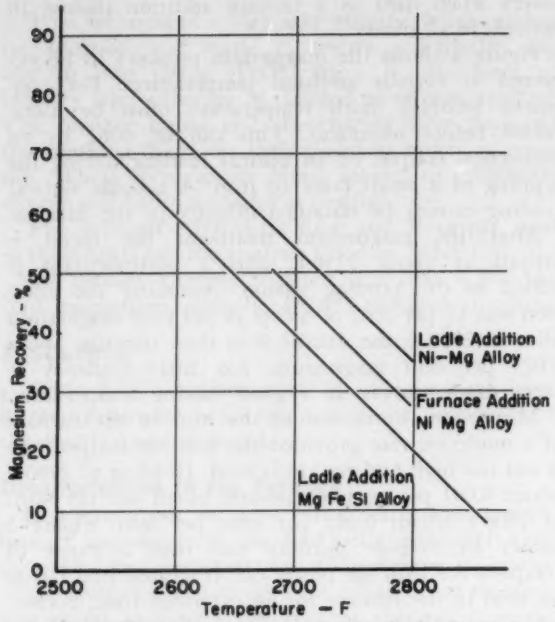


Fig. 4 — Magnesium recovery in the basic direct-arc furnace process.

Despite the degree to which these heats are preconditioned for magnesium treatment, they did not accept a raw magnesium metal addition with any composure. Trials were made during heat 8, plunging about 0.14 per cent bar magnesium into the melt in the furnace of 2750 F. Magnesium recovery was under 5 per cent.

Magnesium Treatment on Tapout

After desulfurization a small magnesium addition to the tapout stream will also produce good quality ductile iron. This addition can be made at a higher temperature because of the greater magnesium recovery effected this way, as compared with an addition to the furnace. Figure 4 shows expected magnesium recoveries at various ladle treatment temperatures. Treatment temperatures must again be ascertained by an immersion couple or a small base tap. Any of the usual proprietary addition alloys are suitable, provided they give sufficient magnesium recovery at the desired addition temperature.

Residual Magnesium Content

The graphite in these refined irons becomes well-spheroidized at unusually low magnesium residuals. Several typical structures at progressively greater magnesium residuals are shown in Figs. 6a-6c. Figure 7 summarizes the residual magnesium graphite spheroidization relationship for all the irons (only well-inoculated irons are considered in this relationship). The threshold value over which predominantly spheroidal graphite ductile irons were produced was about 0.021 per cent magnesium.

Aiming for 0.025 per cent residual magnesium, instead of a usual 0.055 per cent, producers equipped with basic electric melting units can cut magnesium alloy costs in half. Furthermore, because the producer would be treating a completely desulfurized iron, he would not need to add any magnesium alloy for desulfurization. This means further substantial savings.

Dross in Castings

The incidence of dross (cope-surface defect) is greatly lessened in low-magnesium ductile irons. The casting shown in Fig. 8 is used at the authors company's laboratory to measure dross in ductile irons. The three horizontal projecting shelves are broken off, and a $\frac{1}{20}$-in. transparent grid is placed on the fracture. The total number of grid openings through which dross can be seen is counted, and the sum of the counts for the three shelves averaged. An average dross count of 270 is read when one of these standardized 0.07 per cent magnesium castings is poured starting with an 0.04 per cent sulfur charge in a neutral electric furnace. Figure 9 shows a moderately drossy fracture.

Compared with that average dross count, basic-melted irons show less than 15 per cent as much dross at magnesium levels below 0.04 per cent. Several tests are shown in Fig. 10.

MECHANICAL PROPERTIES

Tensile Properties

The tensile properties of refined low-magnesium ductile irons in the 0.02 to 0.03 per cent magnesium range are excellent. Table 2 lists a number of fully treated bars of varying analyses. The properties of these bars, ranging from 113,250 psi ultimate tensile strength and 9 per cent elongation with a pearlitic structure to 60,750 psi ultimate tensile strength and 26 per cent elongation with a ferritic

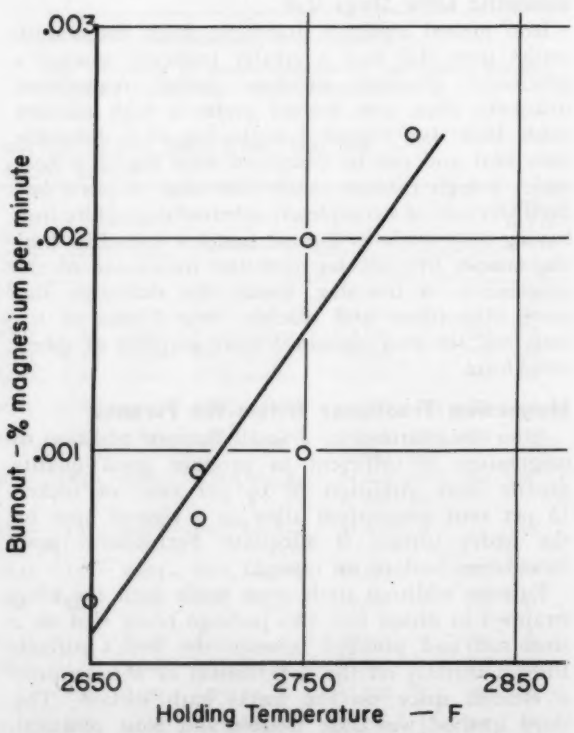


Fig. 5 — Burnout rate of magnesium in iron under a basic slag in the direct-arc electric furnace.

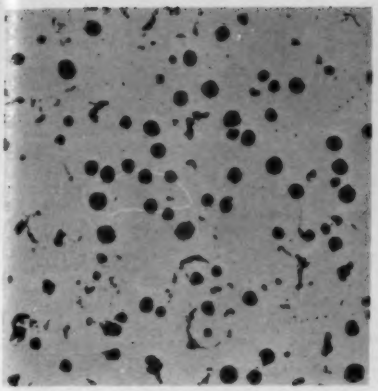


Fig. 6a — 47,000 psi tensile strength. 0 per cent elongation. 0.013 per cent magnesium.

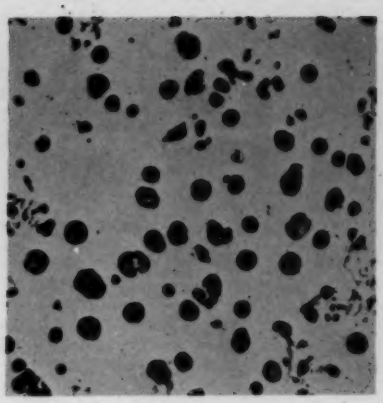


Fig. 6b — 68,200 psi tensile strength. 2 per cent elongation. 0.016 per cent magnesium.

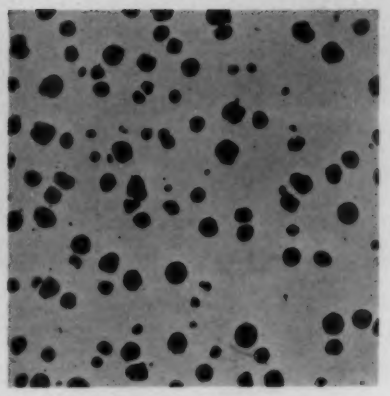


Fig. 6c — 100,000 psi ten. str. 9½ per cent elongation. 0.021 per cent magnesium.

Figs. 6a-6c — (left to right) Graphite structures of basic melted low magnesium irons as-cast 1 in. keel blocks. Unetched. 50 ×.

structure, compare favorably with average ductile iron tensile properties, Fig. 11.

Tests of several irons in the 0.01 to 0.02 per cent magnesium range are also included in Table 2. These irons contained more or less flake graphite, and their tensile properties do not match those of the fully spheroidal graphite iron with higher magnesium residuals.

The average ductile iron property range, shown in Fig. 11, has been useful in judging quality of pearlitic-ferritic ductile irons. It was developed by a statistical study of the tensile properties of 112 one-in. ductile iron keel bars containing 90 per cent or more spheroidal graphite. The range of chemical analyses included among these irons is: 3.4 to 3.9 per cent total carbon; 1.54 to 3.49 per cent silicon; 0.04 to 0.58 per cent manganese; 0.013 to 0.12 per cent phosphorus; 0.07 to 1.96 per cent nickel; and 0.021 to 0.13 per cent magnesium.

Hardness

The chilling propensity of these low-magnesium basic furnace ductile irons seems no less than that of any other ductile irons similarly inoculated. In measurements of hardness of ⅛, ¼, ½ and 1-in. sections, an approximately comparable residual magnesium and carbon equivalent were made. They fail to show the basic furnace iron to be any softer than indirect arc furnace irons melted in a neutral furnace.

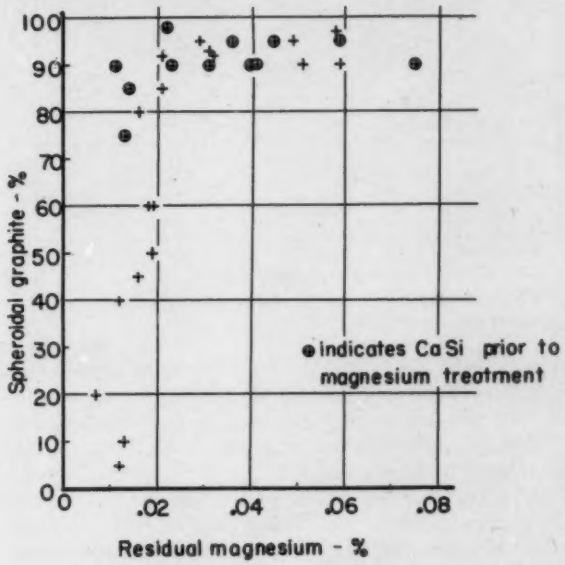


Fig. 7 — Residual magnesium vs. spheroidal graphite in 1-in. double keel block.

CONCLUSIONS

1) Ductile iron can be consistently produced at 0.02 to 0.03 per cent residual magnesium using the basic-lined direct-arc furnace for melting. Deoxidizing the base metal by a late calcium-silicon addition before magnesium treatment assures a large proportion of spheroidal graphite in the 0.01 to 0.02 per cent residual magnesium range.
2) This process offers magnesium alloy savings of 25 to 75 per cent for a ductile iron producer equipped with a direct-arc furnace.
3) Ductile irons so produced exhibit 80 to 100 per cent less drossing tendency than indirect arc furnace irons. The irons have good graphite structures and good physical properties.

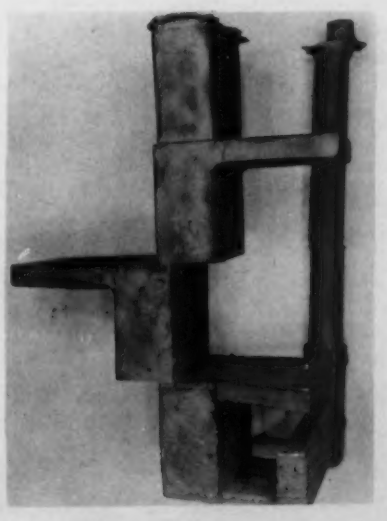


Fig. 8 — Test casting for dross measurement.



Fig. 9 — Fracture of dross test casting shelf showing moderate amount of dross. 1¼ ×.

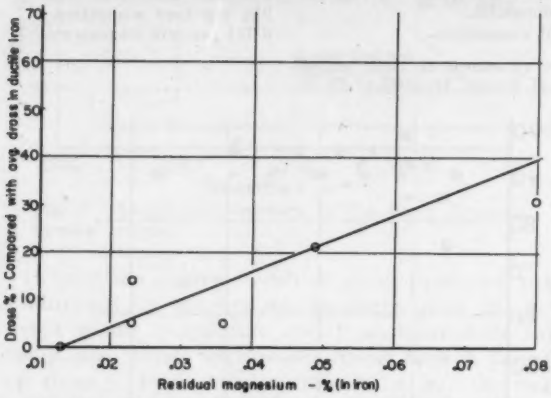


Fig. 10 — Dross in ductile iron melted in the basic direct-arc electric furnace.

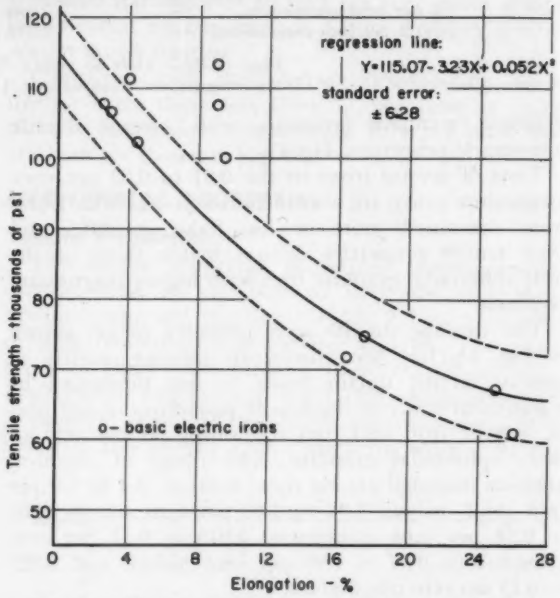


Fig. 11 — Tensile strength vs. elongation relationship of pearlitic-ferritic ductile irons.

4) Desulfurization of the melt to 0.005 per cent sulfur is accomplished under the basic slag by the time the cold charge has been brought to tapping temperature. Desulfurization of hot metal charges is also practical.
5) A 1 per cent to 3 per cent addition of calcium silicon and calcium carbide to the melt covered with a lime-fluorspar slag encourages introduction of up to 0.016 per cent calcium into the iron. This results in 15 per cent of the graphite in the base iron becoming spheroidal without further treatment. Improved physical properties accompany this change in structure.
6) Desulfurization with a dolomitic slag, followed by the late calcium silicon-calcium carbide slag addition, results in up to 0.028 per cent calcium in the iron and up to 50 per cent graphite spheroidization without further treatment.
7) Magnesium treatment of the desulfurized base iron can be accomplished entirely within the furnace. Ductile iron with 0.025 per cent residual magnesium is made by rabbling ½ per cent nickel-15 per cent magnesium alloy into this bath at 2750 F. Magnesium burnout from the treated iron in the furnace varies from about 0.001 per cent magnesium lost per min at 2710 F to about 0.002 per cent lost per min at 2780 F. Postinoculation of the treated iron can be done in the ladle during tapping.
8) Magnesium treatment of the desulfurized iron can be accomplished on tapout. Tapping onto ½ per cent nickel-15 per cent magnesium alloy at 2800 F results in an iron with about 0.025 per cent magnesium residual. Postinoculation is done on reladling.
9) A tensile strength-elongation relationship for good ductile irons was developed statistically, and was used to judge the irons produced in this investigation.

ACKNOWLEDGMENT

Thanks and acknowledgment are extended to A. S. Kightlinger and his foundrymen with whose interest and skilled help these experimental heats were made; to R. D. Schelleng whose dross casting design and system of dross measurement were used; to Dr. J. T. Eash whose trials of the basic direct-arc electric furnace production of ductile iron indicated some possibilities of the method; and to The International Nickel Co., Inc., for permission to publish this paper.

REFERENCES

1. Hignett, H. W. G., U.S. Patent 2,652,324.
2. Chipman, J., "Chemical Behavior of Sulfur in Iron and Steelmaking," *Metals Progress*, pp. 97-107, Dec. 1952.

A HYPEREUTECTIC ALUMINUM-SILICON ALLOY

By Rolf Kissling and Oldrich Tichy

ABSTRACT

Hypereutectic aluminum-silicon alloys have been improved through intensive research. Because of their low coefficient of thermal expansion, high hardness and excellent wear resistance, their potential was recognized early in their development.

One of the commercial hypereutectic aluminum-silicon alloys is analyzed, giving its structure and various properties, as well as foundry characteristics.

Mechanical properties, physical properties (including thermal conductivity, thermal expansion and dimensional stability) and some of the applications of this alloy are given.

DEVELOPMENT

Aluminum-silicon alloys gained little industrial importance until Pacz in 1921 patented the first process to modify the coarse eutectic silicon crystals, thereby improving the mechanical properties considerably. But the sodium modification of Pacz and his successors does not work with hypereutectic alloys, i.e., does not modify the coarse primary silicon crystals, the main disadvantage of these alloys. During the 1920's research therefore was concentrated mainly on eutectic and hypoeutectic alloys with continuing improvement in properties through alloying additions.

Nevertheless, based on their low coefficient of thermal expansion, their high hardness and excellent wear resistance, the potentials of the hypereutectic alloys were recognized and a number of patents appeared, each one with additions intended to refine the primary silicon and improve the properties. But additions such as copper, cobalt, antimony, chromium, nickel, titanium and vanadium merely strengthened the matrix, or acted as grain refiners at the most, without affecting the size and number of the primary silicon crystals. Machining still was extremely difficult, and interest in these alloys therefore faded while the foundry industry concentrated on the heat-treatable near-eutectic alloys.

During World War II, with its shortage of certain alloying elements, interest in the hypereutectic aluminum-silicon compositions again developed, particularly in Europe. Renewed research led to improved alloys, especially in the silicon range 18-35 per cent with a preferred content in the neighborhood of 21 per cent silicon. The researchers found limited use for pistons and small internal combustion engines, but machinability even with the newly developed carbide tools still left much to be desired.

Although the addition of phosphorus to hypereutectic aluminum-silicon alloys had been patented as far back as 1933,[1] it was not until the early 1950's that European workers[2,3,4,5,6,7,8] developed an effective method of refining the hypereutectic silicon with phosphorus to improve machinability, and thereby opened up a field of applications for these alloys.

STRUCTURE

One hypereutectic alloy with a silicon content of about 21 per cent, is a commercial alloy (U.S. Patents 2357 449, 2357 450, 2357 451, 2357 452) of the following nominal composition:

Mg %	Mn %	Si %	Cu %	Ni %
0.7	0.5	21	1.8	0.4

The structure of this alloy shows primary silicon crystals surrounded by a complex, nonmodified eutectic. Magnesium and copper, through the formation of magnesium silicide and aluminum-copper-magnesium compounds, provide the desired heat treatability. Silicon, although soluble to the extent of about 1.6 per cent at the eutectic temperature and less than 0.1 per cent at room temperature, does not contribute noticeably to the heat treatment response. Copper in solid solution improves the dynamic properties, while manganese and nickel contribute to better strength at elevated temperatures.

The iron content is controlled by the use of special high-grade silicon with low iron developed in the authors' company. Since the liquidus temperature of a 21 per cent silicon alloy is appreciably higher than that of most commercial alloys, and the high latent heat of silicon also tends to slow down the rate of solidification, grain refiners are added to the alloy.

REFINEMENT

The main difficulty with the early hypereutectic alloys consisted in their coarse primary silicon crystals which, due to their lower specific gravity tended to segregate and their large size ruined the machining tools. Various modifying compounds were known to influence the silicon, and in the early 1930's it was found in eutectic[9] and hypereutectic alloys[1] that small amounts of phosphorus break up and transform the lamellar silicon into a polyhedral form. The effect probably is due to the presence of nuclei

R. KISSLING is Met. and O. TICHY is Chief Met., Apex Smelting Co., Cleveland.

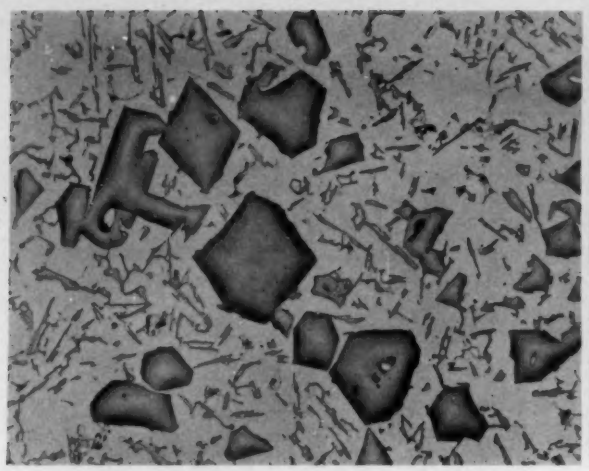


Fig. 1 — Microstructure of a hypereutectic alloy containing 21 per cent silicon and 0 per cent phosphorus. 100 ×.

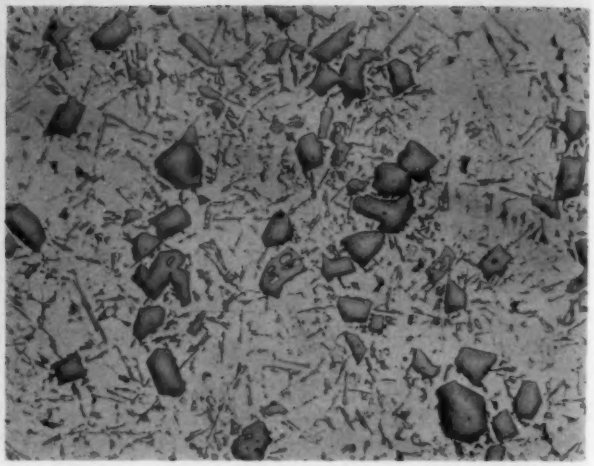


Fig. 2 — Microstructure of a hypereutectic alloy containing 21 per cent silicon and 0.01 per cent phosphorus. 100 ×.

and, like the sodium modification mechanism, is controversial.

The most important feature of the modified alloys is the marked increase in the dispersion of silicon. The phosphorus probably has a twofold effect:

1) It changes the crystal habit of the silicon.
2) It acts as a nucleating agent for the primary silicon.

But it seems that consistent results were difficult to obtain in commercial practice until it was found that the size[8] of the refining salts apparently plays an important role, and that sodium on the other hand tends to counteract the phosphorus effect.[5,6]

Care should be taken, therefore, that phosphorus-modified alloys are not contaminated with sodium in the foundry, for instance through sodium-containing fluxes of fluorides or carbonates. Unlike the sodium modification which disappears rapidly, the phosphorus refinement carries over into the remelts, possibly as much as three or four times, which is of great advantage to the foundryman. Chlorine can be used on the melts for cleansing without detrimental effects on the phosphorus modification; in fact chlorine is recommended for better results.[7,8]

Figure 1 shows a hypereutectic alloy containing 21 per cent silicon without phosphorus. The primary silicon is in the characteristic large form. An addition of 0.01 per cent phosphorus (Fig. 2) breaks up the coarse particles into smaller more rounded ones without affecting the eutectic structure.

FOUNDRY CHARACTERISTICS

The fluidity of aluminum-silicon alloys increases with increasing silicon content. At 20 per cent silicon the curve starts to flatten out and seems to reach a maximum around 21 per cent silicon.[11] Since the cooling curve shows a primary arrest at approximately 1280 F which is unaffected by the phosphorus addition, the pouring temperature therefore must be higher than for most commercial alloys—1400-1450 F for sand and permanent molds, and 1300-1350 F for diecasting.

Hypereutectic alloys will have a slight tendency to hot-shortness of the same order as the commercial hypoeutectic alloys, hot-shortness being a function of the solidification interval. Feeding characteristics on the other hand are poor, due to the wide solidification interval and the corresponding mushy condition. Like any other alloy with a solidification interval and a primary crystallization, the hypereutectic alloys exhibit a certain amount of microshrinkage.

Therefore, feeding distances must be kept short and wide, but short risers should be used instead of the narrow and long risers used with eutectic aluminum-silicon alloys. Linear shrinkage is smaller than for most commercial aluminum alloys, of the order of 1.1 per cent.

MECHANICAL PROPERTIES

The mechanical properties of the hypereutectic aluminum-silicon alloys depend strongly on the silicon content. Figure 3 shows the decrease in tensile strength with increasing silicon content. The nominal composition of the previously mentioned alloy given earlier is a compromise between low coefficient of expansion and high mechanical properties. For the intended special applications, the coefficient of thermal expansion and the wear resistance are of more importance than high mechanical properties. Table 1 gives the results obtained with separately cast test bars of a commercial heat of the composition:

Mg %	Mn %	Si %	Cu %	Fe %	Ni %	P %
0.77	0.58	20.30	1.66	0.65	0.37	0.01 (added)

Due to the slow cooling in sand castings, the hardening constituents precipitate in relatively large particles with only a small amount retained in solid solution, and therefore the response of mechanical properties to an aging heat treatment is small. Permanent mold and die castings on the other hand show a definite increase. A solution heat treatment followed by a hot water quench and subsequent aging will give further improvement (Fig. 3).

Up to 400 F (Fig. 4) the influence of temperature is small, and at about 500 F the strength of the hyper-

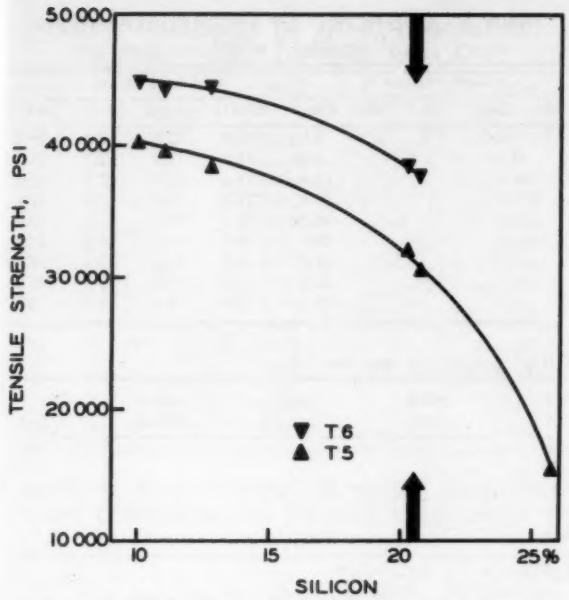


Fig. 3 — Silicon effect on tensile strength of aluminum-silicon alloys. Chill cast.

eutectic alloy compares favorably with that of forged A.S.T.M. CN 42 A (27,500 psi)[22] and chill cast A.S.T.M. SC 103A (23,000 psi)[23] alloy. The testing of mechanical properties after 1 hr at temperature seems to fairly measure the temperature effect. Longer times at temperature had almost no further effect on these properties, as illustrated in Fig. 5.

Short time elevated temperature properties are given in Table 2.

PHYSICAL PROPERTIES

Thermal Conductivity

The thermal conductivity of aluminum increases with increasing temperature. This is in contrast with ferrous metals where increasing temperature reduces the thermal conductivity. Alloying elements which form solid solutions with aluminum reduce its thermal conductivity. Silicon being only slightly soluble does not markedly affect the conductivity of the matrix. Since the thermal conductivity of this matrix in a solution heat treated condition is noticeably lower because of the larger amounts of dissolved elements, a heterogenous, i.e., completely precipitated structure is preferable from the viewpoint of thermal conductivity.

A chill cast specimen has a higher thermal conductivity than a sand cast one due to the finer and denser structure, and the modified alloy shows a slight advantage over the nonmodified. In the modified state the casting method, sand or chill cast, has little effect on thermal conductivity.[12] Typical results reported from various sources are shown in Table 3.

Thermal Expansion

The hypereutectic aluminum-silicon alloys have the lowest coefficient of thermal expansion of known aluminum alloys, which makes them ideally suited for

TABLE 1 — MECHANICAL PROPERTIES AT ROOM TEMPERATURE* (AVE. OF 12 BARS TESTED)

Item	Temper	Tensile Strength, psi	Brinell Hardness, HB
Sand Cast —F	As-Cast	21,700	76
—T5	400 F for 8 hr	22,600	90
Chill Cast —F	As-Cast	25,100	87
—T5	400 F for 8 hr	32,000	102
—T7	960 F for 6 hr — Hot Water Quench—400 F for 8 hr	38,200	110
Die Cast	As-Cast	35,200	115

*Elongations 0-1 per cent.

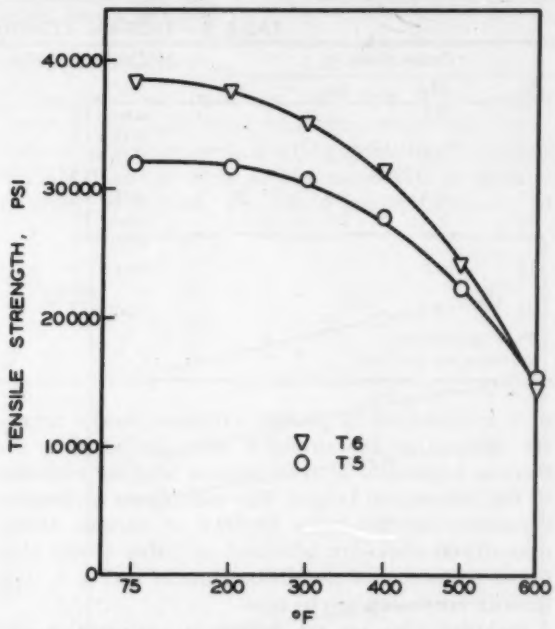


Fig. 4 — Elevated temperature effect on tensile strength. Mechanical properties tested after 1 hr at temperature. Chill cast.

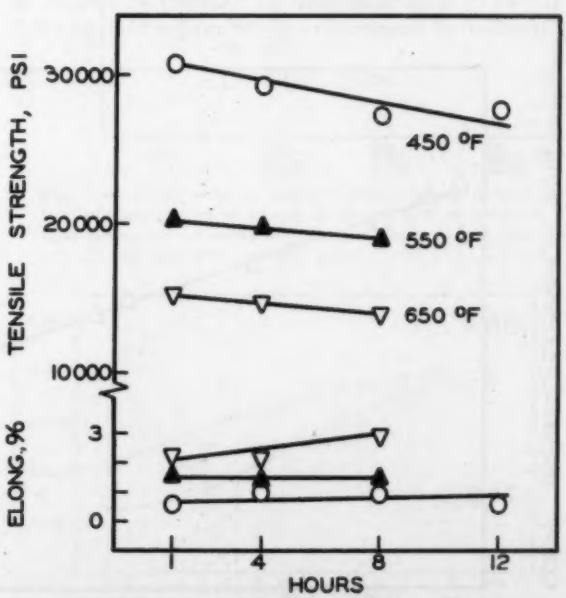


Fig. 5 — Length of time at elevated temperatures effect on elongation and tensile strength. Chill cast.

TABLE 2 — SHORT TIME ELEVATED TEMPERATURE PROPERTIES
(Chill Cast: 1 hr at temperature. Ave. of six bars tested)

Temp., F	—T5 400 F/8 hr Tensile Strength, psi	Elong., %	—T7 960 F/6H—HWQ—400 F/8 hr Tensile Strength, psi	Elong., %
75	32,000	0-1	38,200	0-1
200	31,700	0-1	37,800	0-1
300	30,900	1.0	35,200	0-1
400	27,800	1.0	31,600	0-1
500	22,300	1.4	24,200	1.0
600	15,300	1.9	14,500	2.6

TABLE 4 — COEFFICIENT OF THERMAL EXPANSION: in./in./°F x 10^{-6}

Composition, %					Temperature, F			
Cu	Mg	Ni	Mn	Si	68-212	68-392	68-572	68-752
1.6	0.5	0.4	0.5	8.1	11.9	12.3	12.9	13.2
				9.8	11.6	12.2	12.7	13.0
				10.8	11.4	12.0	12.5	12.9
				12.3	11.2	11.6	12.1	12.4
				16.2	10.3	11.0	11.6	11.9
				20.1	9.9	10.4	10.8	10.9
				21.1	9.8	10.2	10.7	11.0
				25.6	9.1	9.5	9.8	10.0
				26.2	9.1	9.4	9.7	9.9

TABLE 3 — THERMAL CONDUCTIVITIES: cal/sq cm/sec/cm/°C

Composition, %					75 F (24 C)	200 F (93 C)	400 F (204 C)	600 F (318 C)	Ref.
Si	Mg	Cu	Ni						
7	0.3			sand—T6	0.36				14
				chill—T6	0.38				14
8	0.7	1.5		chill—F	0.25				18
				—T5	0.27				18
12	0.8	1.0	2.5	chill—T55	0.28				14
20	1.0	2.0		sand—T5	0.20				12
				chill—T5	0.22				
12	0.5	1.2		chill—T4			0.337	0.350	13
				—T7			0.378	0.379	13
12	0.3			chill—T4			0.354	0.365	13
Pure aluminum							0.540	0.538	14
Automotive cast iron					0.112	0.110	0.107	0.105	15

such applications as pistons, cylinders, brake drums, etc. Silicon has the strongest effect in reducing the thermal expansion of the common alloying elements in the commercial ranges. The coefficients of thermal expansion for the range 68-392 F of various aluminum-silicon alloys are tabulated in Table 4, and plotted as a function of the silicon content in Fig. 6. The linear relationship can be seen.

Included also are the values for automotive cast iron (7.1 x 10^{-6} in./in./°F)[15], and the formerly widely used A.S.T.M. CN 42 A alloy (12.6 x 10^{-6} in./in./°F).[16] The coefficient of thermal expansion as a function of temperature in the ranges from 68-750 F of aluminum-silicon alloys containing 12 per cent, 21 per cent and 25 per cent silicon, are shown in Fig. 7. A similar curve for cast iron[15] is included for comparison.

Dimensional Stability

Aluminum alloys, like most metallic alloys, have slightly different densities depending on the temper. The corresponding dimensional variations are commonly called permanent growth, or simply growth. In aluminum alloys, growth is generally considered to be the result of a precipitation of alloying constituents from a supersaturated solution such as pro-

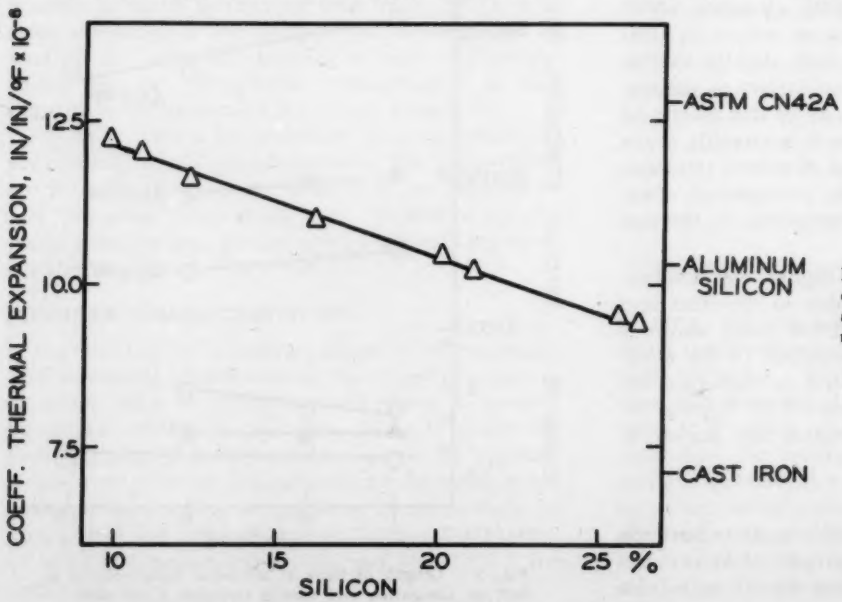


Fig. 6 — Silicon effect on the coefficients of thermal expansion of aluminum-silicon alloys in the range of 68 — 392 F. Chill cast.

TABLE 5 — INFLUENCE OF AGING TEMPERATURE ON DIMENSIONAL CHANGES, CHILL CAST

Time, hr	400 F in./in. × 10⁻³	450 F in./in. × 10⁻³
2	0.137	0.211
4	0.146	0.314
6	0.187	0.309
8	0.217	0.314
10	0.244	0.330
12	0.261	0.359
14	0.277	0.349
16	0.291	0.361
21	0.309	0.364
30	0.350	0.377
36	0.366	0.397
40	0.361	0.396
44	0.367	0.384
54	0.369	0.369
72	0.381	0.384

TABLE 6 — TYPICAL VALUES OF TENSILE STRENGTH AND CORRESPONDING REMOVAL OF GROWTH, CHILL CAST

	Tensile Strength, psi	Growth Removed, %
As-Cast	26,000	
960 F/6 hr — Hot Water Quench —		
300 F for 3 hr	35,000	
355 F for 12 hr	37,000	
400 F for 8 hr	38,000	
355 F for 12 hr	33,000	
400 F for 8 hr	30,000	55
450 F for 8 hr	28,000	85

duced by a solution heat treatment and subsequent quenching. It should vary, therefore, with the temperature of reheating in the same manner as the solid solubility of the alloying elements concerned.

The time to maximum growth is a logarithmic function of temperature, but does not coincide with that for maximum hardness. At slightly elevated temperatures growth takes place so slowly as to be of no practical importance, while the upper limit of the reheating temperature is fixed by the necessity of getting a high combination of mechanical properties. Since growth take place later in time than hardening, maximum mechanical properties and complete freedom of growth cannot be obtained simultaneously. The heat treatment, therefore, is necessarily a compromise between an aging treatment and a degrowthing treatment.

However, it must be noted, that dimensional changes are not necessarily due to growth. Machining, for example, can release internal stresses with an increase in dimensions. An internal or service stress of 15,000 psi for instance, produces a dimensional increase equivalent to the maximum growth of most commercial alloys. The same holds true for a thermal expansion due to a temperature rise of only 100 F.[17]

The hypereutectic aluminum-silicon alloy mentioned previously under "Mechanical Properties" exhibits a freedom from growth unique among the usual piston alloys and other aluminum alloys of the copper-silicon type. Typical values are shown in Table 5.

The permanent growth of this alloy is plotted against time at temperature (logarithmic scale) in Fig. 8. The effect of several heat treatments on the tensile strength and growth is illustrated in Table 6. It can be seen that with a simple aging treatment the peak values probably should be obtained with 12 hr at 355 F in commercial practice, while the commonly used heat treatment of 8 hr at 400 F or 450 F leads to a slight over-aging.

However, considerably more stress relief and degrowthing takes place at these higher temperatures, which in many cases is of more importance than a slight loss in properties.

APPLICATIONS

The combination of low coefficient of thermal expansion and high wear resistance makes the hypereutectic aluminum-silicon alloys particularly suitable for applications such as pistons, brakes,[19] internal combustion engines, etc. Their use in small gas en-

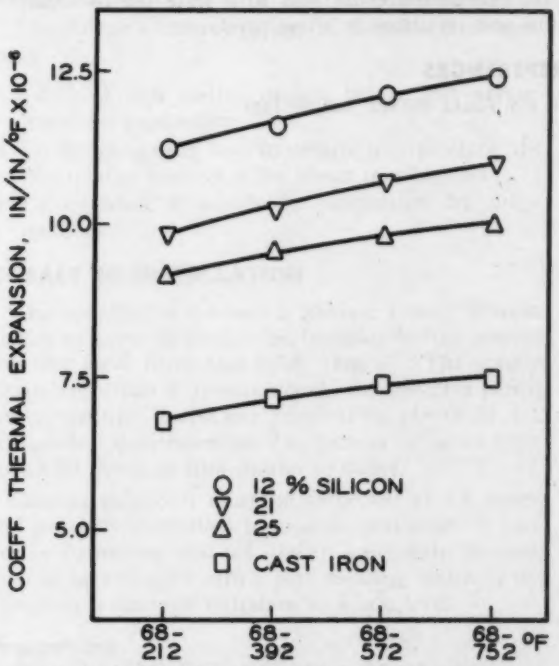


Fig. 7 — Coefficients of thermal expansion as a function of temperature in ranges of 68 — 750 F of alumninum-silicon alloys containing 12 per cent, 21 per cent and 25 per cent silicon. Chill cast.

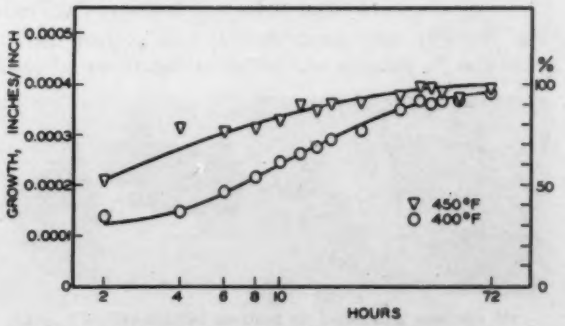


Fig. 8—Aging temperature effect on permanent growth of tested alloy plotted on logarithmic scale.

gines was started some time ago, and the results point to excellent possibilities for automotive engines where early research and development work led to sand, semi-permanent mold and diecast engine blocks with wet or dry aluminum and cast-iron liners.[20]

The perfecting of the hypereutectic aluminum-silicon alloys makes an all-aluminum engine look attractive. The entire block can be cast in the previously mentioned alloy thereby eliminating the need for liners, or an aluminum block can be fitted with liners of this alloy. These liners will be either cast or extruded and forged. The use of the same material for the pistons is a natural consequence, since the necessary clearances can be greatly reduced.

The early difficulties with valve seats (experienced for example in aluminum heads for aircraft engines) have successfully been overcome in automotive blocks by the use of pressed-in seat rings made of special cast iron.[21] Work is under way between foundries and automotive manufacturers, and the mutual efforts will undoubtedly lead to a wide use of aluminum and its alloys in these products.

REFERENCES

1. U.S. Patent 1940922, Dec. 26, 1933.
2. M. L. Grand, *Fonderie, 2625,* 1951.
3. K. Lohberg and E. Schulz, *Z. Metallkde,* 43, *50,* 1952.
4. E. Ontisch-Modl, *Proc. First Met. World Congress,* A.S.M., *325,* 1952.
5. G. Guertler. *Z. Metallkde, 44,* 503, 1953.
6. C. Mascré, *Fonderie, 91, 3556,* 1953; *Foundry Trade Journal, 725,* 1953.
7. W. Thury and H. Kessler, *Z. Metallkde, 46,* 846, 1955; *Light Metals, 19,* 225, 1956.
8. H. Kessler and H. Winterstein, *Z. Metallkde, 47,* 97, 1956; *Light Metals, 19,* 294, 1956.
9. E. Scheuer and E. Schulz, *Bericht des Metallaboratoriums der Metallgesellschaft A. G.,* No. 1433, Feb. 1934.
10. J. C. Wagner, *Metal Progress, 69,* 334, 1956.
11. R. Kuemmerle, *Metall,* 848, 1957.
12. R. M. Smith, *S.A.E. Journal, 48,* Sept. 1958.
13. Euringer and Hanemann, *Metallwirtschaft,* 14, 389, 1935.
14. *Metals Handbook,* A.S.M., 1948.
15. Cast Metals Handbook, American Foundrymen's Society, 1957.
16. C. R. Barber, *Metallurgia,* 41, 20, 1949.
17. L. W. Kempf and H. L. Hopkins, *Trans. A.I.M.E., 122,* 266, 1936.
18. Apex Research Laboratory files (unpublished).
19. R. L. Atkins and F. J. Weber, S.A.E. Preprint 14R, Jan. 1959.
20. J. M. Smith and R. M. Smith, S.A.E. Preprint 68A, June 1958.
21. E. Hundt, S.A.E. Preprint 68B, June 1958.
22. R. L. Templin and D. A. Paul, *Symp. on Effect of Temperature on the Properties of Metals,* A.S.M.E.-A.S.T.M., 198, 1931.

PROPER GATING THROUGH USE OF COBALT 60

By A. J. Karam

ABSTRACT

Nondestructive testing by use of cobalt-60 radiography has helped the author's company improve the quality of their castings. The company makes a threefold use of this type of inspection:

1) **To check new casting design before production begins.**
2) **To design gating layouts to produce highest quality castings at lowest cost.**
3) **To establish inspection standards.**

Examples of how cobalt-60 (gamma) radiological inspection has helped various castings produced in a proprietary pearlitic malleable iron are given.

INTRODUCTION

During recent years, the foundry industry has witnessed advancements which require improved casting techniques and better inspection media. One inspection medium, Cobalt-60 radiography, enables the author's company to provide customers with quality castings at reduced cost, and in a minimum of time.

A proprietary pearlitic malleable iron makes castings suitable for many applications which were formerly forgings, stampings and fabrications. When castings replace other materials in applications where soundness is a major consideration for satisfactory performance, it is especially important to utilize inspection methods which assure castings free from internal defects of any kind.

The traditional method of inspecting castings for internal defects in the foundry was cutting, breaking (Fig. 1) and etching. This procedure not only destroyed the casting, but was time consuming and costly, without a guarantee that the presence of a defect would always be located.

Shrinkage voids usually constitute the largest portion of sub-surface defects in castings. Shrinkage in a casting is caused by the lack of sufficient liquid feed metal to make up for the contraction, which occurs during solidification from liquid to a solid form.

FEEDER SIZE EFFECT

The size of the reservoir or feeder containing the liquid feed metal affects the internal soundness of a casting. In Fig. 2 are three sectioned castings made with varying heights of feeders. At left is a sound, shrink-free casting. The center casting has an extensive area of light internal shrinkage or porosity as indicated by the dark etch stained area on the casting. Casting on the right has a severe internal shrink void at the top with a lighter area of porosity below it.

Figure 3 shows the same group of castings as they appear on a radiograph. On the left is a sound casting, while the middle casting is an example of slight internal shrinkage as indicated by the dark spots and the right-hand casting shows severe internal shrinkage.

The author's company's use of Cobalt 60 is three fold:

1) To check new casting designs before they are released for production.
2) To design gating layouts which will produce the best quality castings at the lowest possible cost.
3) To establish standards for inspection by other methods.

COBALT 60 INSTALLATION

The installation contains a 30-curie Cobalt-60 unit, housed in a special windowless building having poured concrete walls three feet thick (Fig. 4). The cobalt-60 source, which is approximately the size of a pencil eraser, has the equivalent penetrating power of a 2 million volt x-ray machine. The gamma radiation from cobalt 60 reacts on film similar to x-rays.

Gamma radiation must be respected at all times, and properly controlled by trained personnel. It cannot be turned on and off like an x-ray unit. Instead, it must be retracted into a lead housing which is designed to reduce the radiation to a safe level.

Precautions

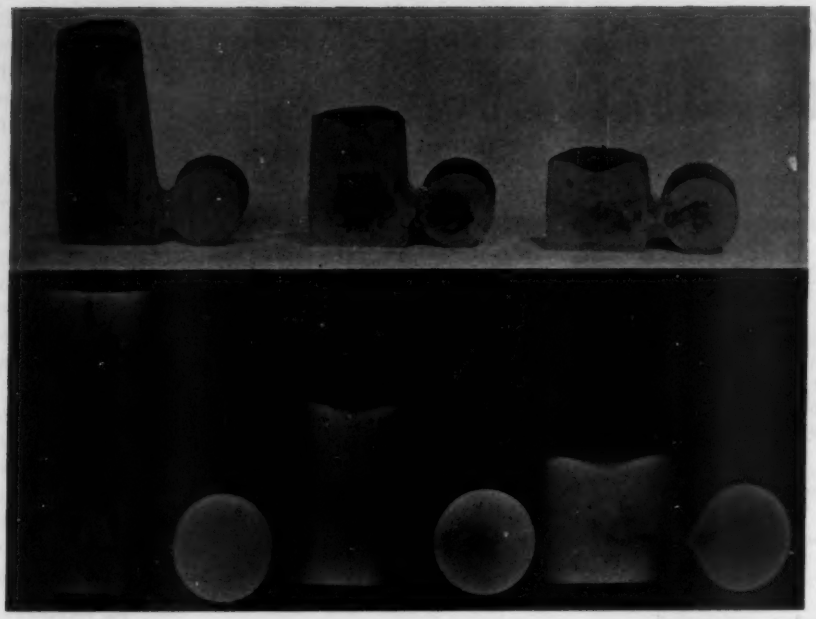
The control panel is located outside the exposure room behind an electrically interlocked door, and in conjunction with signal warning lights. A gamma survey meter (Fig. 5) is used to monitor the room after each exposure as a safety measure.

Film badges and pocket dosimeters (Fig. 6) are worn by personnel to detect the amount of radiation



Fig. 1 — Traditional method of inspecting castings for internal defects was by cutting (left), breaking (right) and etching.

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Figs. 2 and 3 — Three sectioned castings made with varying heights of feeders (top), and as they appear on a radiograph (bottom). At left is a sound, shrink-free casting. The center casting has an extensive area of light internal shrinkage or porosity (dark areas). The casting on the right has a severe internal shrink void at the top with a lighter area of porosity below it.

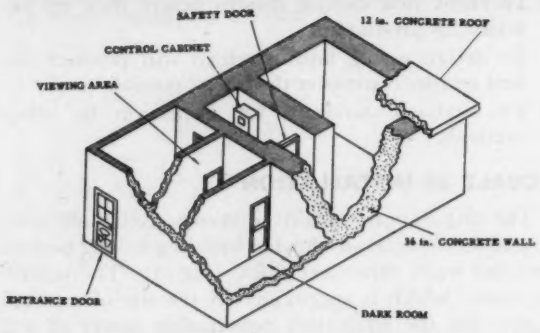


Fig. 4 — Special housing for cobalt-60 unit, showing main features of installation.

exposure. On the right of Fig. 6 is shown a film badge which creates a permanent record of radiation dose (milliroentgens) received during a period of time, usually one week (A milliroentgen is a unit of radiation. It is a measure of the film darkening effect or possible damage to animal tissue). On the left of Fig. 6 is a pocket dosimeter which records accumulated dose in mr (milliroentgens) of radiation. This instrument can be read at any time without waiting for a film badge report.



Fig. 5 — A gamma survey meter is used to monitor the room after each exposure as a safety measure.

The cobalt source is housed in a small stainless steel capsule in the center of a 1000 lb lead container (Fig. 7). It is moved to the exposure position by remote control. Exposure time is decided by the strength and age of the cobalt source, type of film used, casting section size and distance from source to film. All developed radiographs are read on a high intensity viewer by personnel trained to interpret the film accurately.

DESIGN CHANGES

Figure 8 shows how radiography was used in work on the elimination of one gate and feeder on a differential carrier. The first samples were made using three gates and feeders, two on the flanges and one on the tube as indicated. This tube gate is not a part of the mold, but is formed with a separate gate core.

Figure 9 shows a cutaway view of the position of the gate core in relationship to the casting. This adds cost and extra operations to the making of the casting, which makes it desirable to eliminate the tube gate. In Fig. 10 is the radiograph of the casting made in this manner, showing slight shrinkage (circled area) in the tube gate area. This indicated that this feeder was not providing adequate feed. Many variations with two flange gates were tried until satisfactory castings were made.

This was accomplished by a redesign of the part (Fig. 11), which added five ounces of metal to a thin section near one gate for effective feeding and the use of one chill in the tube area. This chill enabled the author's company to equalize the cooling rate of this heavy, isolated tube section without the use of an additional feeder. These changes resulted in a sound quality casting of improved design, as shown in Fig. 12.

Cobalt-60 radiography is also being used to lower

costs by reducing the amount of waste metal in feeders and runners to produce a given casting. Uniform sections and good design enable the foundryman to get the maximum number of good castings per mold with minimum amount of gating.

Quality castings at lower cost is the ultimate goal of the foundry engineer, and is accomplished only by a series of methodical steps on each job. Development of a better gating layout is done step by step through a series of changes, and radiography shows the results brought about by each change.

A typical example of improving foundry practice through the use of cobalt-60 is a steering gear housing casting which was gated in the same manner as many of the housings of this type. Figure 13 a radiograph of this casting with its attached feeders. Note that the feeders marked 1, 2, 3 and 4 showed little internal shrinkage. This is a sign of excess feed metal (Note slag catcher marks on radiograph).

It was decided to reduce the height of feeders 1, 2 and 3 by 1½ in., and through a design change to eliminate feeder 4. A small amount of metal was added to the wall of the tube section to permit feeders 1, 2 and 3 to feed this section through progressive solidification.

These changes were too drastic, since shrinkage voids appeared in the vicinity of feeders 2 and 3, as marked (circled areas) in Fig. 14. Note that no shrinkage voids appeared in the vicinity of feeder 1, indicating sufficient feed metal at this point.

The volume of metal in feeders 2 and 3 was increased by raising the feeders ½-in. in height (Fig. 15). This additional metal was sufficient to eliminate the shrinkage voids, and resulted in the sound casting. This is an excellent example of how a small engineering change supported by radiography eliminated the need for a feeder.

Figure 16 shows a universal joint yoke that is made in the proprietary pearlitic malleable iron, with a single gate and feeder on the tube of the casting. An engineering change request by the customer reduced the wall thickness along the entire tube 1/16-in. After this pattern change was completed, sample castings

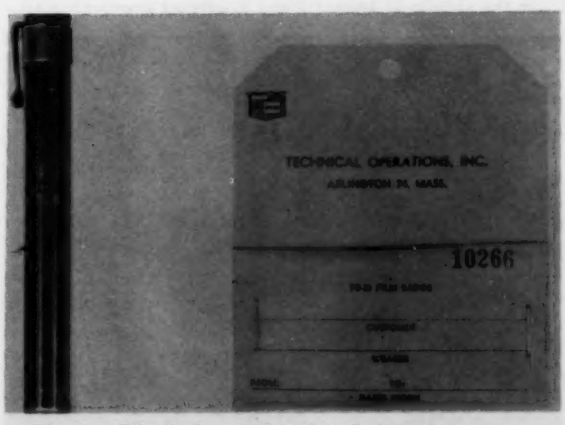


Fig. 6 — Film badges and pocket dosimeters are worn by personnel to detect the amount of radiation exposure. The pocket dosimeter (left) records accumulated dose of radiation in milliroentgens. The film badge (right) creates a permanent record of radiation dose in milliroentgens received during a period of time (usually a week).



Fig. 7 — The cobalt source is housed in a small stainless steel capsule in the center of a 1000 lb lead container (left). It is moved to the exposure position by remote control.

(Fig. 17) were checked radiographically, and shrinkage voids were found in the yoke arm away from the gate (circled areas). The feeder was radiographed, and found to have sufficient feed metal. This indicated that the decreased section thickness of the tube was causing the shrinkage voids. The section thickness of the casting in this area was increased to permit adequate feeding (Fig. 18). This was done to the inside diameter of the tube from the gate area to the yoke arm, and sound castings were produced using this gating.

Figure 19 is a small gasoline engine crankshaft made in the pearlitic malleable iron with two gates and

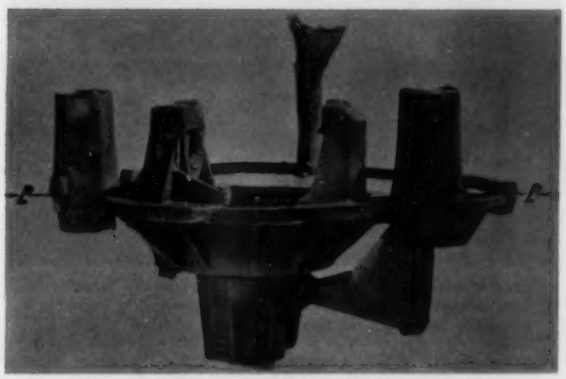


Fig. 8 — Radiography was used in work on the elimination of one gate from this differential carrier.

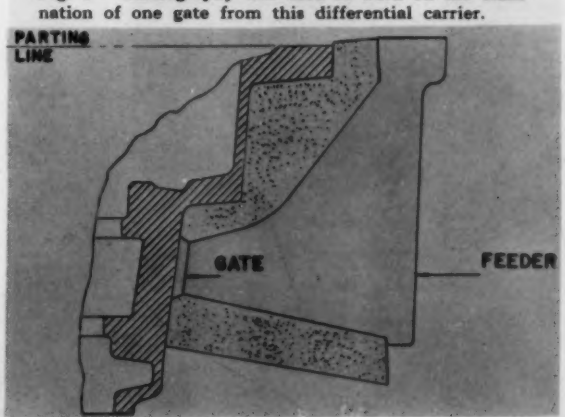


Fig. 9 — Cutaway view of differential carrier shown in Fig. 8, indicating position of gate core in relationship to the casting.

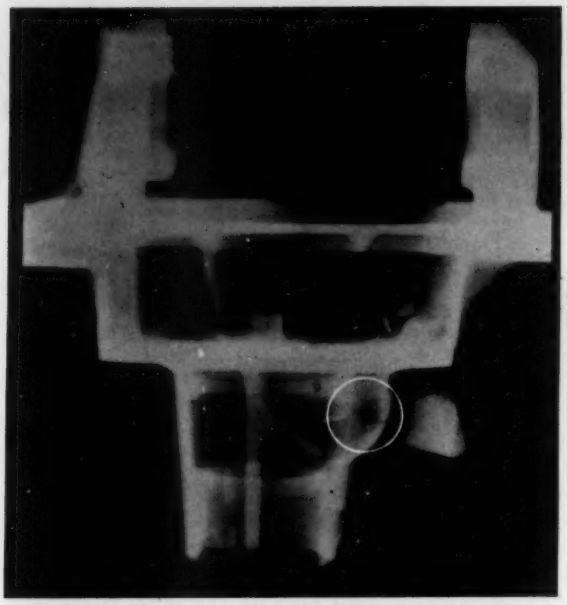


Fig. 10 — Radiograph of casting (Fig. 8) with tube gate, showing (circled area) slight shrinkage in area of tube gate. This indicated that the feeder was not providing adequate feed to the casting.

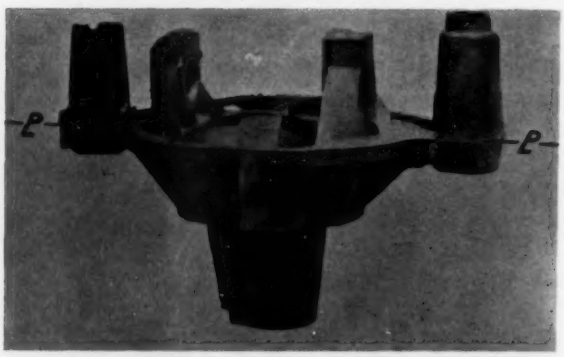


Fig. 11 — Redesign of casting in Fig. 8 which added five ounces of metal to thin section near one gate for effective feeding and the use of one chill in the tube area.

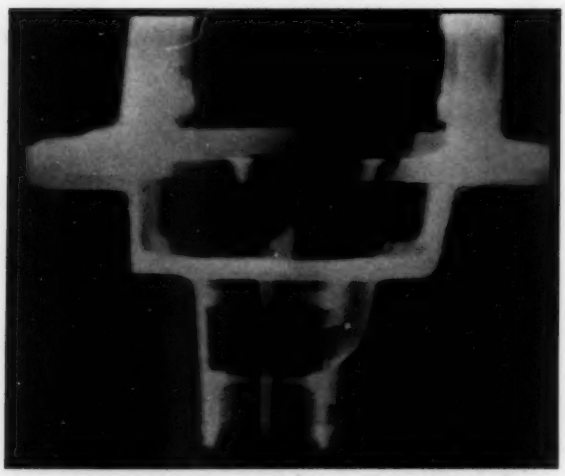


Fig. 12 — Radiograph of redesigned casting in Fig. 11, showing sound quality.

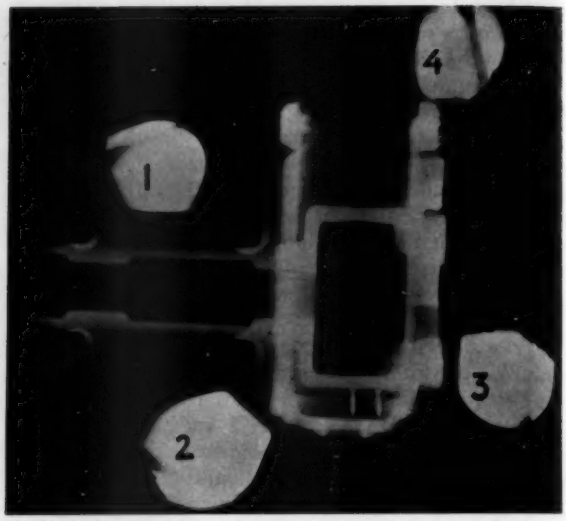


Fig. 13 — Radiograph of steering gear housing casting with its attached feeders. Little internal shrinkage is shown in feeders 1, 2, 3 and 4. This is a sign of excess feed metal. Note slag catcher marks on radiograph.

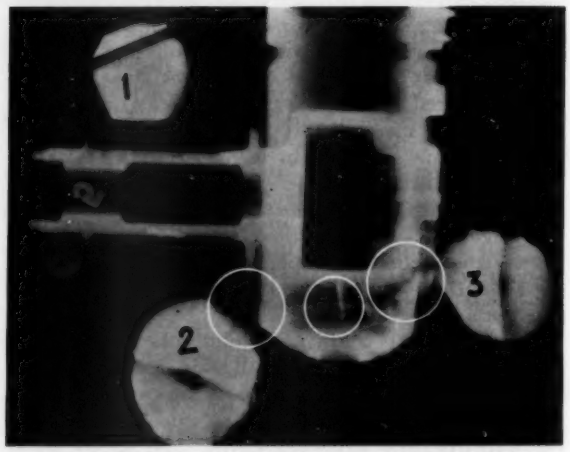


Fig. 14 — Elimination of feeder 4 proved too drastic, as shown in radiograph, since shrinkage voids appeared in the vicinity of feeders 2 and 3 (circled areas).

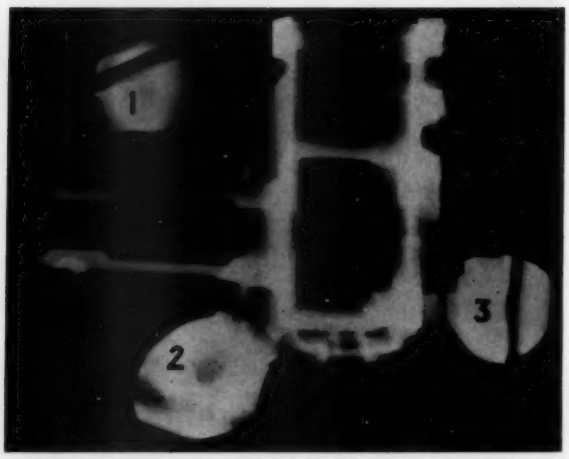


Fig. 15 — Volume of metal in feeders 2 and 3 was increased by raising height ½-in. The additional metal was sufficient to eliminate shrinkage voids, and resulted in the sound casting.

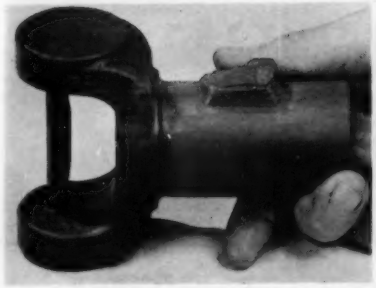


Fig. 16—Universal joint yoke with single gate and feeder on the tube of casting.

feeders. This gating was considered necessary to produce a sound casting because of the geometry of the part (Fig. 20). Also, a small tie bar was required between the two counterweights to prevent distortion during annealing. Because the gates were on a contoured surface removal by grinding was a costly time-consuming operation.

These gates could not be ground to an exact contour by hand, which presented the customer with problems in his machine shop by causing an intermittent cut during the turning operation on the shaft. Also, the gates made the shafts difficult to straighten by pressing because of irregular die contact in processing at the plant. An engineering change (Fig. 21) was made to determine the possibility of feeding the castings by combining the two gates in the form of an enlarged tie bar, which could be removed with a milling operation.

Experimental castings made in this manner were satisfactory (Fig. 22), and a pattern change was made to incorporate this into the production equipment. These castings were processed by the customer without the previous machining difficulties, and the author's company was able to supply straighter crankshafts because the elimination of the gates on the shaft diameters gave better contact in the straightening fixtures. These changes resulted in a cost reduction of 2.3 per cent.

The proprietary pearlitic malleable iron is being used as a replacement material in many instances where forgings had been used previously. One of these is an automotive connecting rod (Fig. 23) which was made experimentally using four gates and feeders to produce a sound casting. However, because of gate removal problems and costs, it was necessary to improve the gating. Through the use of cobalt 60, these castings could be made with two gates and feeders (Fig. 24).

The design of the casting in the bearing cap area was altered by increasing the section thickness between the balancing lug and the bolt bosses. This made it possible to obtain sufficient feed for the bolt boss areas with one feeder instead of the three previously used.

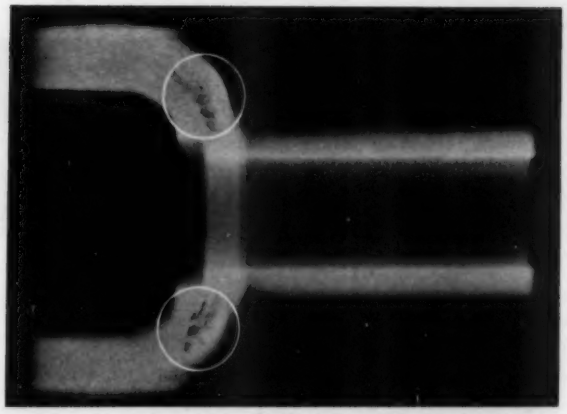


Fig. 17 — After a change in design which reduced wall thickness $\frac{1}{16}$-in., sample casting showed shrinkage voids on radiograph in yoke arm away from gate (circled areas).

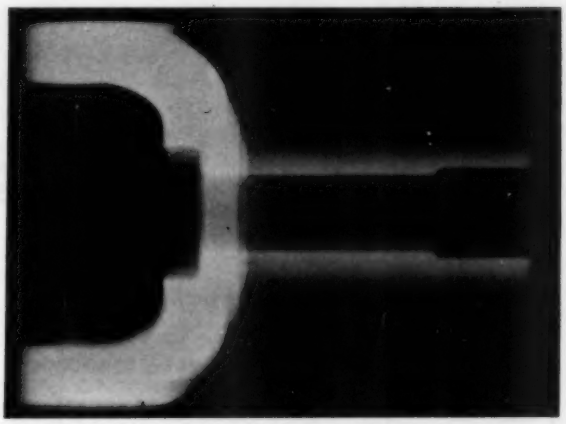


Fig. 18 — The section thickness of the casting (Figs. 16, 17) was increased in this area to permit adequate feeding. This was done to the inside diameter of the tube from the gate area to the yoke arm, and sound castings were produced.

This small addition to metal to the casting gave a radiographically sound part (Fig. 25). These shell molded connecting rods have stood up under many tough fatigue tests, and are being tested in automobiles today.

Cobalt 60 is also an aid in setting up and controlling other forms of casting inspection. It is a rapid method of determining the seriousness of a defect, and helps set up sound inspection standards without destroying the part. For example, sonic frequencies are verified for a set control point by radiographing the castings and establishing standards for the process.

Another example of how this radiography is used at the author's company is shown in Fig. 26, which shows a group of cast crankshafts being set up in one of the



Fig. 19 — Small gasoline engine crankshaft made with two gates and feeders of pearlitic malleable iron.



Fig. 20 — A small tie bar was required between the two counterweights to prevent distortion during annealing on crankshaft shown in Fig. 19.



Fig. 21 — An engineering change in casting (Fig. 19) was made in order to determine the possibility of feeding the casting by combining the two gates in the form of an enlarged tie bar.

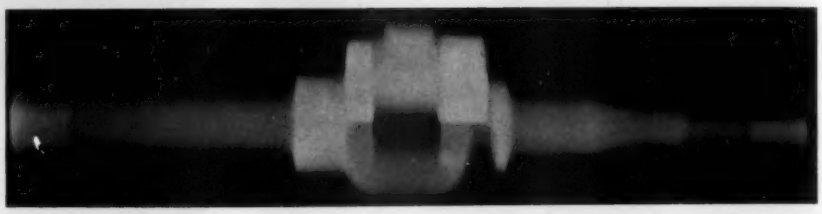


Fig. 22 — Experimental casting made with enlarged tie bar (Fig. 21) was satisfactory.

facilities. This is only one of the many products which are inspected by the means of cobalt-60 radiography.

In summarizing, the major benefits of cobalt 60 have been found to be:

1) The assurance of internal soundness provides the correct safety factor in gating.
2) The time element in development of a part is greatly reduced.
3) Radiography enables sound casting techniques with a minimum of pilot casting trials.
4) Also, the process broadens sales opportunities by aiding in enlarging the scope of casting application in industry.

The potential of this process is virtually unlimited. Cobalt-60 radiography substantially aids in the production of castings of the highest quality at the lowest possible costs for the author's company's customers.



Fig. 23—An automotive connecting rod made experimentally using four gates and feeders to produce a sound casting.



Fig. 24 — Through use of cobalt 60, these castings (Fig. 23) could be made with two gates and feeders, through altering the design in the bearing cap area.

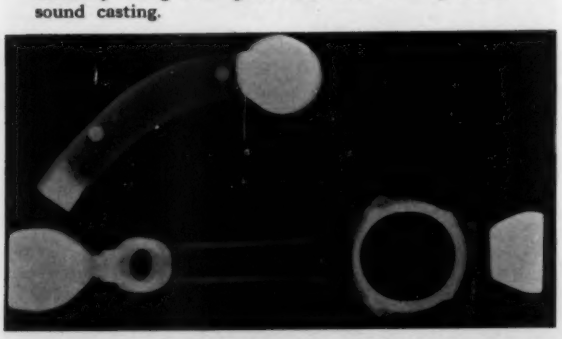


Fig. 25 — The additional metal added by increasing section thickness produced a radiographically sound part.

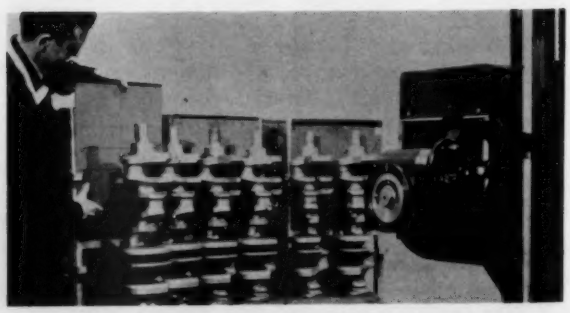


Fig. 26 — Group of crankshafts being set up for radiography.

STEEL CASTINGS FOR THE AIRCRAFT INDUSTRY

quality requirements

By Y. J. Elizondo

ABSTRACT

The need for quality steel castings in the aircraft industry is readily understood when the aircraft and missile design requirements are reviewed. The experiences of one aircraft company over the past three years are summarized to illustrate various reasons for success or failure of several steel casting designs.

Most of the problems and troubles which exist today are not due to technical reasons, but are the direct result of misunderstandings existing between the two industries. Recommendations are made for eliminating or at least reducing this source of problems.

In order to achieve optimum utilization of steel castings in the aircraft industry, which is required to keep up with the state of the art, several improvements must be made in the foundry practices. In addition, technological advances will also be required to keep up with the modern design requirements. Recommendations for achieving these goals are presented.

INTRODUCTION

For the past several years engineers at the author's company have tried to incorporate precision steel castings in both aircraft and missile designs. This effort, although initiated for economic reasons on the individual designs, was also supported by the knowledge that steel castings would become essential for the support of future designs. This trend is readily evident from data in Fig. 1, which presents the number of steel castings used by the company on two general types of designs. Although the actual number of castings will be a function of the design, the trends indicated are representative. It is worth noting that the discontinuity in the upper curve coincides with the above mentioned effort.

It should also be noted that Fig. 1 represents only a portion of the total number of potential steel casting applications. Many potential applications reverted to other forms of fabrication when the steel casting supplier was unable to produce a part with the required performance. It will be the purpose of the following discussion to clarify some of the requirements which prevented the utilization of all the casting designs.

In order to proceed without any confusion a definition of structural aircraft castings is needed. If an aircraft or missile were to have the skin, hydraulic system, electrical system, fuel system, electronic equipment, instruments, and the like removed there remains a skeleton which can be referred to as the primary structure. The subject castings would be those which form a part of this basic skeleton. They are by no means all the castings used in an aircraft or missile, but they do represent those which are causing the greatest amount of trouble to both the foundries and the aircraft designer. In addition they represent the largest potential market to the casting vendor.

Figure 2 illustrates several typical examples of this type of casting. For example, the 95 structural castings used in one design represent approximately 176 lb or only about 2 per cent of the airframe weight. In this same design there were four additional castings proposed which were not used due to foundry problems, design details and schedule difficulties which would approximately double the weight of the castings used on this design. In retrospect, several

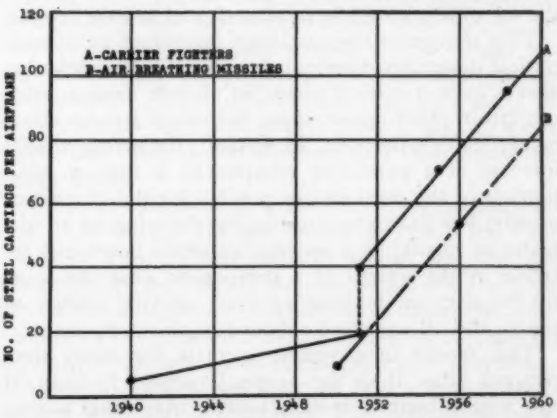


Fig. 1 — Steel castings used on aircraft designs at the author's company.

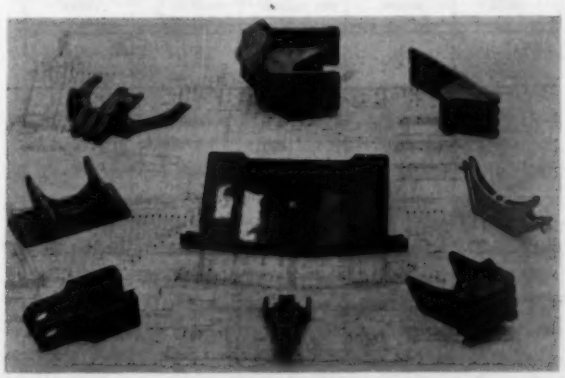


Fig. 2 — Typical structural aircraft castings.

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other casting applications have been discovered, and at the present time three of these casting applications are being cast as part of a steel casting program sponsored by the Air Material Command, and on which the author's company is the prime contractor.

These three castings would increase the weight of the steel castings used on this design by approximately 140 lb. It is easy to see that the trend in Fig. 1 could easily be doubled or even tripled if a better understanding existed between the foundries and the aircraft designers.

AIRCRAFT AND MISSILE DESIGN REQUIREMENTS

To better understand the requirements for the steel casting, a brief knowledge of the design requirements being imposed on the aircraft designer is necessary. With the advent of supersonic speeds the problem of aerodynamic heating becomes a significant factor. Almost any article today will present a picture of the temperatures that can be expected at various altitudes and speeds. These temperatures for existing aircraft are already of the order of 300 F and 600 F. With the advent of intercontinental missiles and space and satellite programs the temperature encountered will become even higher.

Designs already exist where temperatures in the order of 1000 F to 2000 F will be attained, and there are other experimental designs where even higher temperatures are expected. The first visible result of this temperature problem was the gradual increase in the use of ferrous alloys in aircraft and missile designs.

This change to ferrous alloys magnified an already critical design requirement—weight. While weight has always been a critical factor in aircraft design, more so than in other applications, the use of ferrous alloys magnifies this problem. In general, the actual weight increase of a particular component is only a small portion of the total weight penalty for the aircraft or missile. For example, considering the wing of an airplane or missile as a simple cantilever beam any increase in the weight of a component near the wing tip requires an increase in load carrying ability of practically all structure inboard of the component.

This results in a weight increase for many components other than the original culprit. In turn, if this weight increase is significant it may cause a drop in performance, which can only be overcome by other means such as increased engine output, meaning a heavier engine. This in turn may require an increase in weight of the structure supporting the engine, in order to carry the heavier engine. It is not unusual, therefore, to experience an overall weight increase of seven to 15 times the actual weight increase of the original component.

SHAPE	SIZE INCHES	AREA SQ. INCHES	SECTION MODULUS $Z=I/C$, IN^3	STRUCTURAL INDEX Z/A
○	1	0.785	.098	.125
□	1 x 1	1.000	.167	.167
L	2 1/2 x 1 1/2 x 1/4	1.000	.266	.266
◎	2 1/2 x 1/8	0.938	.406	.476
C	3 x 1 3/4 x 1/8	0.781	.750	.963
I	3 x 2 x 1/8	0.844	.840	.994

Fig. 3 — Structural efficiency index, bending.

Closely associated with weight in the designs of today is available space. In the modern aircraft or missile there is practically no unused space within the aero-dynamic shape or shell. Any significant change in shape of a component usually results in a change in external shape, elimination of some other component, or compromise in function. This means that even in those cases where the weight increase may not be objectional, loss in aerodynamic efficiency or performance of some other part will prohibit a change in size or shape of a component.

This factor is more important in proposed design changes than in the original design, since it is assumed that the original design encompassed the above considerations. However, it will still influence to a certain degree the design freedom in the original concept.

Design Requirements

To meet the weight and space requirements the aircraft designer must utilize more complex shapes and maintain closer adherence to optimum structural shapes than is generally necessary in other industries. This is reflected in two ways, 1) the designs are usually based on ultimate strength not proportional limit or yield strength; and 2) the margins of safety rather than factors of safety are used. This means that many short cuts, compromises and tolerances which may be common in a majority of the commercial applications will be unacceptable to the aircraft designer. It also requires the use of different design parameters.

For example, if a member to carry bending is desired, an analysis of the many possible shapes will be made. A summary of a few of these is illustrated in Fig. 3. Where a general designer might be able to use the third column as a guide, since this number is a measure of the load carrying ability, the aircraft designer must be guided by the fourth column which is a measure of the strength-weight ratio. This figure is by no means complete nor does it consider all possible modes of failure, it assumes failure will occur in tension at the outermost fiber.

To obtain a complete evaluation such items as complex loading or bi-axial stresses, buckling and crippling must be taken into account. The aircraft designer must go through all of these factors, as well as repeating the same type of analysis for the other major types of loadings, in order to arrive at the optimum design.

The loading breakdown of a typical aircraft design must be known in order to realize the relative importance of the various mechanical properties which govern the design of a component. In a typical design the breakdown may be as follows—shear 37 per cent; bending 23 per cent; compression 21 per cent; and

tension 19 per cent. The actual percentages might vary significantly from one design to the next, but the general order of magnitude will probably remain fairly constant. It should also be noted that this breakdown is based on the critical loading for each component, and many of them are subjected to more than one type of loading at any given time.

In most cases fatigue has not been a major factor, but this is closely associated with the design function of the aircraft. For example a transport or passenger aircraft would be much more prone to be critical in fatigue than say a military fighter, and a ballistic missile would probably have no fatigue problems. Wherever fatigue is a factor, the major consideration is that endurance limit is not the criteria to use. What will be important is the life of a part at high stress levels, that is in the finite life region rather than infinite life.

In summary, the aircraft designer will be searching for a material and a process that will meet his design objectives. In order to do this he will require a part which will meet certain minimum strength requirements, with closer tolerances than are generally needed in other industries. He will not be as willing to compromise as his counter part in other industries, but he will be willing to pay a greater premium for the attainment of his objectives than his counter part.

CASTING REQUIREMENTS

The first item that comes to mind as a requirement for aircraft steel castings is high strength, or the ability to retain its strength at elevated temperature. Although this will be essential to the efficient utilization of steel castings by the aircraft industry the primary requirement is more basic. Before the above can even be evaluated properly the problem of reproducibility must be resolved. As pointed out earlier the aircraft applications do not permit the use of large factors of safety to allow for any significant variation in properties.

There are several instances today where castings were originally designed into the structure but eventually were not used because of the inability of the foundries to produce a uniform product from heat to heat. Usually by the time sufficient design factors were incorporated in the design to account for the observed variations, the part had become too large to fit into its proper place or so heavy that a non-ferrous material could be used with improved strength-weight ratio.

Over the past year a statistical sampling plan has been in operation at the author's company to evaluate the mechanical properties of production heats of castings. The normal procedure followed to place a production order was to release the design for procurement of a qualification heat of castings. This order required the castings to be of a certain alloy and have minimum strength requirements in addition to dimensional requirements. In several cases negotiation with the foundries produced several changes in the original requirements. Once agreement was reached, the first lot of castings was produced and delivered.

This lot was dimensionally inspected, and then tested in a jig simulating the actual airplane structure and loadings, as well as by conventional tensile test coupons

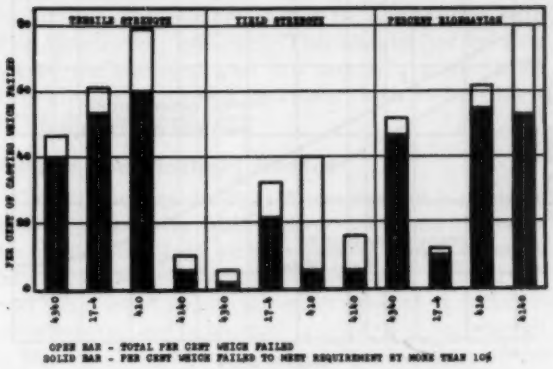


Fig. 4 — Per cent of castings which failed to meet minimum strength requirements.

cut from the castings. The jig test was then corrected to the minimum strength level and compared to the design requirements. In several instances this procedure was repeated since the above evaluation disclosed design errors as well as strength deficiencies in the casting. Once the casting passed the above qualification, production orders were placed. The sampling plan was then used on castings procured in this manner. The results are shown in Fig. 4.

The picture that evolves is discouraging to say the least, particularly when follow-up action has disclosed the reasons for these failures. In some cases melt practice was changed, heat treatment procedures were varied, raw material sources were changed, gating and risering was changed, mold composition was changed and pouring and tap temperatures were varied. In most cases the changes were made with a sincere desire to improve the product either economically or strength wise. The fact remains, that the effects of these changes were not always known, and in cases where they were it was assumed that they would be insignificant.

If the aircraft designs were to have significant factors of safety the changes would probably have resulted in a part which was still usable, but this was not generally the case. In striving to achieve the minimums that the aircraft designer set, considerable time and effort was expended by the foundries in developing improved alloys. Extensive metallurgical and chemical analysis studies, and involved heat treatment procedures were also investigated. This without a question has helped and will be required for the eventual optimization of the castings, but it does not solve the basic problem. Neither will it be possible to properly evaluate these other factors until a uniformity of foundry procedures are achieved.

Guaranteed Minimum Strength

Returning now to the requirement of high strength, it must be realized that this to the aircraft designer means high guaranteed minimum strength not nearly average strengths. Furthermore, this strength must be achieved on the castings, not on separately or integrally cast test bars. This is not to say that there is nothing to be gained from test bar results, since they must be used effectively for foundry control, but to the casting user they are useless unless the designer wishes to fly test bars. The aircraft designer recognizes that

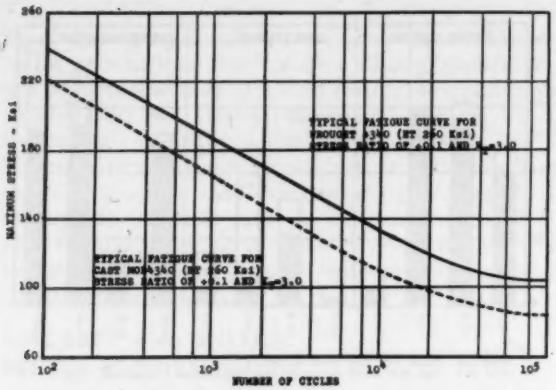


Fig. 5 — Axial load fatigue strength for S.A.E. 4340 steel. The only significant difference in alloys was the microscopic gas porosity and inclusions. Both of the alloys met existing inspection criteria.

the problem of establishing minimum strength levels in a complex casting is not simple, but then neither is his problem of guaranteeing the strength level of the aircraft.

Strengths Desired

Without having minimums for the individual components, it is almost impossible to guarantee the strength level of the aircraft. In the past he has worked with other industries and has obtained minimum guarantees from them, which have been met consistantly, therefore, he is expecting the foundries to do the same thing.

As pointed out earlier the strength desired by the aircraft designer is not always tensile. In most cases the criteria will be shear, bending or compression, and unless these other properties have been proven to have a fixed relationship to the tensile properties they also must be evaluated. In many cases castings are rejected because the tensile properties are a few per cent below the requirement. To the foundry this appears unreasonable since the deficiency is practically insignificant, and if the casting were critical in tension only, this might be true. However, unfortunately few parts will remain critical in the manner designed if other properties are reduced.

A simple illustration of this would be a case where a casting was rejected due to the elongation being five per cent rather than six per cent as required. If the part were critical in tension only it is highly doubtful that this reduction would make the difference between a good and a bad part. Assume, however, that this reduction was due to small inclusions or some other similar defect too small to be picked up by x-ray or other conventional inspection procedures. What would this do to the fatigue strength? A good example of this is illustrated in Fig. 5. This means the aircraft designer is faced with a decision not of a few per cent deficiency in elongation but a reduction in fatigue strength many times greater.

The need for elevated temperature strength is fairly obvious when the aircraft requirements discussed previously are considered. Steel castings capable of being used at temperatures up to 1000 F must become common place, and casting alloys for use at 1000 to 2000 F must be developed. The next question might be what strength levels are required? The only possible answer at the present is that they must be as high as possible. When the present missile designs are examined it will be discovered that the pay load is usually five per cent or less of the total weight.

Any increase in strength realized in the structural components will permit a significant increase in this pay load by reducing the weight required for these structural components. This permits heavier pay loads to be carried without requiring a new family of missiles to accomplish the job, and allows higher performance with the lower pay loads.

Two Important Factors

Two factors must be kept in mind when attempting to get these high strength levels at the high temperatures. First, the other properties such as compressive strength and fatigue strength must be kept at approximately the same relative position to tensile strength as they are at room temperature. This is necessary in order to avoid any major changes in design concepts. If it is not possible to do this, then the aircraft designer must be informed so that the basic design concepts can be modified accordingly.

Second, the properties at room temperature must be kept at reasonable levels. This latter item is mentioned since there have been several instances where desirable elevated temperature strengths have been obtained, but the room temperature ductility has been reduced to practically zero. Since the aircraft or missile must at least start its flight at relatively low temperatures and carry significant loads at the same time, it must maintain some reasonable load carrying ability at the lower temperatures.

Associated with the problem of strength at temperature is that of time at temperature. In a missile where the total life may be measured in minutes, the desired criteria would be completely different than for a transport or passenger airplane where the life desired may be several thousand hours. This means there will be widely varying needs by the aircraft designer depending on the type of design on which he is working. Most of the work done up to the present deals with properties associated with long exposure time.

The properties most urgently needed are those for short exposure times, from a few seconds to some one or ten hours. Creep properties for total deformations of less than one per cent are also needed. In addition, fatigue strengths at temperature using practical stress raisers for life on the order of one to 100,000 cycles are needed since most published data deals with the longer life region.

Loading Importance

One other factor which has assumed an important roll is rate of loading. There are many applications where a particular casting was not considered due to its reportedly low strength. Closer examination revealed that the strengths reported were based on relatively slow loading rates. When the material was tested at loading rates simulating the design load rates it proved to be satisfactory. Since these loading rates will vary quite widely depending on the design it is almost impossible to predict what value should be used.

In general, the rates used should be increased as

much as possible to obtain data at faster load rates than are commonly used today. The important item is to record loading rates not strain rates, and include this information with the reported strength data. In any case, more information is needed on the effect of loading rates on the strength of the castings and casting alloys to enable the aircraft designer to use them.

In establishing tolerance requirements it is not just a matter of making parts fit together and within a given envelope, but also a matter of weight. Studies conducted on existing aircraft have disclosed that due to tolerance allowables the average weight of the airframe has increased over the nominal by as much as ten per cent. This is readily understood when it is recognized that each process operation on a part is going to be established on the high side of the tolerance range. For this reason the aircraft designer is going to be striving for the closest tolerance ranges possible.

Optimum

The optimum is going to be a balance between what is possible by the casting process and what the aircraft designer is willing to pay. What makes this problem of compromise difficult is to have the foundry demand its usual tolerance of $\pm$ 0.005 in./in., and then produce a part with less than one half of this spread and refuse to back down on the original tolerance spread on follow-up orders. For example, a part with a 15 in. dimension would have a tolerance of $\pm$ 0.075, yet when produced by the foundry the spread from one casting to the next over 15 heats was on the order of $\pm$ 0.025.

Perhaps the solution would be to recognize that the tolerances requested by the foundries at the present cover two items, 1) the possible variation from one part to the next due to the normal process variations, and 2) the error involved in estimating shrinkage in constructing the pattern. If this were made clear to the aircraft designer, then a compromise could be reached on how much he is willing to pay for pattern rework, to reduce or eliminate the latter of the two sources of tolerance build up.

Finish requirements are going to vary widely depending on the design function of the part. In general, due to the nature of the casting applications, the requirements are going to be as rigid if not higher than for a majority of the other structure. Basically, a finish from 125 to 250 rms will be sufficient. Where better finish is required provisions should be made for machining or polishing, unless the casting process will normally produce a better finish. The finish is a source of many arguments as pertains to fatigue life.

If it is considered that all components must be joined to each other, and that this junction is usually formed by putting holes in the part and using fasteners, the stress concentrations due to surface finish become less important. It has also been found that for most alloys presently used, the effect of nominal changes in surface finish do not significantly affect the fatigue life of the parts at high stress levels, rather they affect the fatigue life at low stress levels. Since the aircraft designer is primarily concerned with high-stress low-life fatigue, the surface finish, within limits, assumes a secondary role.

In any case, if the surface finish of 125 rms could be maintained, the castings would be competitive with other fabrication processes. This should not be considered the ultimate goal but merely a guide, since there already are casting processes which will give better finish than this.

Uniform Specifications Needed

Finally, the most important requirement is going to be a set of uniform and rigid specifications to govern the production of the aircraft castings. Based on experiences in other industries it is going to be necessary to establish a set of specifications for general or commercial products, and a separate specification for aircraft or high quality products. This has been done in the other industries but not in the casting industry. Here the aircraft industry is being asked to buy castings that do not meet any basic minimums. This is not possible to do, and utilize castings as completely as the aircraft industry would like to do.

Once an alloy having desirable mechanical properties and acceptable casting qualities has been found a specification for it should be written. This is not always the same composition or necessarily the same properties as the wrought alloy. This specification should govern such things as compositional limits, strength limits and those process variables which affect the strength levels as well as establish sampling procedures for evaluating these properties.

Since the foundry industry has not produced these necessary specifications many aircraft companies have a set of casting specifications, supposedly for the same alloys. This produces an even greater amount of confusion than is already present. In each case, based on the experiences with the individual design involved, the specifications are modified to produce the optimum for that design. This means that each specification tailors the alloys and casting to meet some specific design requirement.

Unless by some chance the original alloy and configuration were optimum from all points of view this procedure will eventually lead to as many specifications for a given alloy as there are design applications. This is not necessary, nor is it desirable from the viewpoint of the aircraft designer or the foundry. All that is required to reduce or eliminate these various specifications, many of which conflict with each other, would be a set of specifications from the foundry industry guaranteeing minimum properties for the steel castings.

Setting Up Specifications

In setting up these specifications it must be recognized that the existing inspection criteria are not adequate for evaluating the aircraft steel castings. Essentially they will reject those castings having obvious defects, but those which pass are not necessarily good. In a test program conducted by the author's company a total of 114 specimens from 14 heats produced at 3 foundries using a modified S.A.E. 4340 alloy were inspected and tested. Basically, the specimens consisted of standard 0.505 in. round tensile specimens machined from keel blocks. A fairly simple form to pour, and certainly an easier part to inspect than the average casting.

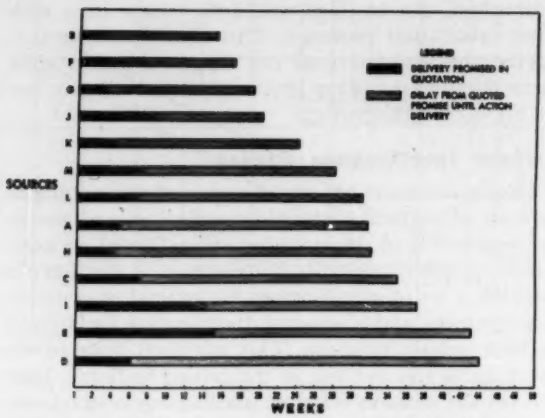


Fig. 6 — Foundry performance showing quoted delivery vs. actual delivery of castings.

These specimens were subjected to the normal inspection criteria required by existing military specifications as well as other applicable casting specifications by both the foundry quality control personnel as well as company engineering personnel. These bars met all the requirements such as x-ray, dye check, fluorescent penetrant, Rockwell, Brinell, etc. However, 15 per cent of them failed at loads significantly below the general level and sometimes at less than 30 per cent of the average. If the specification criteria were disregarded and all bars showing some irregularity, particularly in x-ray, were thrown out, half of the low values would be eliminated.

This still leaves approximately seven to eight per cent which cannot be detected. If actual castings were used instead of test bars, than the numbers would probably be more like those indicated in Fig. 4. The above study was made specifically to check the existing specification criteria without introducing shape or complexity as a factor. Granted the data is far too limited to draw firm conclusions, but certainly the indications are strong that a re-evaluation of existing specification criteria is necessary for aircraft quality castings.

FOUNDRY REQUIREMENTS

To achieve these casting requirements several factors involving foundry practices and procedures must be recognized. The first of these is process control. It is logical to expect that an improvement in quality of the product can only be achieved by improvement in process controls. Since the aircraft designer is asking for a higher quality casting than the general commercial casting, it follows that the process controls utilized for a commercial casting must be improved, when used on aircraft castings. In many instances this merely means closing up of the tolerances allowed. In other cases it may require a whole set of new controls.

An example of this encountered by the author's company was a program conducted to produce a high alloy steel casting. The original efforts by three foundries resulted in parts for which the strength levels varied from 160,000 to 230,000 lb/sq in. The first efforts by the foundries to improve this situation was by variation of composition, improved heat treatment procedures and the like These efforts raised the average result but did not change the relative magnitude of the scatter.

When, at the request of the company, the process variables such as melt practice, pouring temperature and pouring time were controlled as much as was practically possible, the range was reduced to 180,000 to 200,000 lb/sq in. for the original alloy. In other words, mechanizing the process as much as possible increased the minimum strength by a greater amount than was possible with the metallurgical, chemical and heat treatment variations.

When the latter improvements were then added to the improved process the final result was an alloy with a minimum strength of 210,000 lb/sq in. and a scatter of only 20,000 lb/sq in. The interesting thing is that this was accomplished at no significant increase in the cost of the casting.

The second factor which must be considered is time. In order to keep up with the advances in the aircraft industry, and in the military field to maintain the national defense status, the time required to produce an aircraft or missile has assumed a major role. All operations involved in producing the final product must be accelerated. In order to meet these accelerated demands all processes must be expedited. In this respect, the experiences of the aircraft industry with the foundry industry have been poor. Figure 6 presents a summary of the performance experienced by the author's company over a period of time.

Two things must be improved, 1) the time lapse between promised delivery and actual delivery must be reduced or eliminated, and 2) the time required for actual delivery must be reduced. This need becomes evident when it is realized that there are aircraft designs which must be completed within a time span comparable to the times required by the foundries to produce one component of the design. The greater the time required by the foundries to produce a casting, the less chance there will be of using the casting. It has been estimated that if the times required by the foundries could be reduced to half that shown in Fig. 6, at least three times as many castings would have been used in the same time interval.

Cost Performance

Closely associated with the time element is the cost performance. Essentially the cost performance has been almost an exact duplicate of the time performance indicated above. Basically, then, a need exists for accurate cost and scheduling procedures as well as an accelerated schedule in most cases. The major fault to date, appears to be the lack of distinction between the commercial and the aircraft castings by the foundry. The two price structures are bound by necessity to be different.

A third factor that must be considered is not too easy to define. For lack of a better name it can be referred to as progressive thinking. The best example of this is a casting design recently attempted by a foundry. The foundry was given a drawing showing the nominal dimensions required for the final part and asked to cast the part as close to this as they could. The idea was to obtain a final part requiring a minimum of machining. The foundry would also be permitted to make any

changes required as they progressed to achieve this goal. The nominal weight of the part was 110 lb.

The resulting casting produced by the foundry weighed 213 lb, or it means that almost as much material would have to be machined away as the total part weighs. The argument used was that they first wanted to produce a sound casting and then gradually remove material where they could. At this rate it is easy to see that economic considerations would probably stop this procedure long before the optimum were reached. At the insistence of author's company a second pattern was made to the nominal dimensions and castings poured.

This part weighed 123 lb, due to the pattern being at the upper side of the tolerances given to the pattern maker and one or two changes that were obviously needed. In a like number of trials, and at approximately the same cost as before, a final part has been obtained that weighs 133 lb.

When it is considered that most of the castings required by the aircraft designer are going to be of alloys which are difficult and expensive to machine, the importance of keeping the part to a minimum cannot be over emphasized. It seems a shame to produce a casting which will cost more to machine than to cast. Consider also that the casting is competing not against a forging or an extrusion but against these forms plus machining. Any machining eliminated is, therefore, just that much more money that can be directed to obtaining a sound casting. Every effort, within reason, must be made to cast the final shape desired and avoid the middle man, or machining, as much as possible.

The final factor that must be considered is that the aircraft industry will never become a large tonnage customer to the foundry industry. However, it can and will become a large dollar customer. The aircraft designer will be willing to pay for the part and quality that is required, not just purchase steel at so much a pound. As the foundries gain confidence in their ability to reproduce results, and guarantee performance, this situation should change rapidly.

SUMMARY

In summary, the immediate requirements for quality steel castings in the aircraft industry are as follows:

1. High strength steels capable of operating at temperatures as high as 1500 F.
2. Reproducibility of results.
3. Guaranteed strength levels.
4. Adequate inspection criteria.
5. Improved tolerances.
6. Improved cost and time estimates.

An analysis of these requirements reveals that they are goals which should be common to both the foundry industry and the aircraft industry. Therefore, with a sincere effort by both to achieve these goals, and with an understanding of each other's problems and limitations, there is no reason why these goals cannot be attained to the mutual benefit of both industries.

GRAIN REFINEMENT OF SOLIDIFYING METALS BY VIBRATION

By R. G. Garlick and J. F. Wallace

ABSTRACT

A number of different liquid metal compositions, including pure metals and alloys that solidify as solid solutions and eutectics, were vibrated under controlled solidifying conditions. Metallographic examination of the solidified ingots of these metals indicated that all were grain refined to some extent, although the amount of refinement varied considerably. The grain refinement increased for those metals exhibiting the larger solidification contraction. Data in technical literature generally confirmed this finding.

This latter fact provides substantiation of the theory that vibration, with its alternate high and low pressure waves, increases the nucleation rate of the solidifying phases. Mechanisms that describe how this grain refinement is caused by vibration during the solidification of pure metals, solid solutions and anomalous eutectics are discussed.

INTRODUCTION

Previous investigations have established that vibrations applied during the solidification of molten metals will result in degassing and a finer as-cast grain size. The amount of grain refinement, however, has been reported to vary considerably for different metals, and various frequencies and amplitudes of vibration. The grain refinement of pure metals,[1,2] solid solutions when present in a single phase alloy,[2-7] and metals or solid solutions when present as the hypo- or hypereutectic phase,[1,3,6-9] has been shown by the references listed for each.

The behavior of eutectics solidifying under the influence of vibration is considerably different. It has been reported that the eutectic constituents, particularly the minor eutectic phase, coarsened under the action of vibrational energy.[3,4,5,8]

Two theories have been offered to explain these effects of vibration on solidifying melts. The first theory postulates[1,9,10] that the grain refining effect of vibration is produced by fragmentation of primary crystals or dendrites, thereby providing an artificial source of stable nuclei. This fragmentation has been attributed to the high pressure cycle following cavitation,[1] or the viscous forces acting on the dendrites or crystals.[9,10] The second hypothesis explains this effect as an increased rate of nucleation produced by the alternate pressure and rarefaction waves in the melt.[4,5]

This latter hypothesis assumes that introducing vibrational energy into the melt will produce these alternate waves of high and low pressures. The existence of these moving or stationary waves (depending upon size of crucible, method of introduction, frequency, etc.) is generally accepted in the literature.[1,3-5,7-11] This latter theory when fully developed[4] is capable of explaining both the grain refinement of primary crystals and dendrites and coarsening of eutectic constituents that have been observed as a result of vibration.

The fragmentation theory, however, is incapable of explaining this coarsening of eutectic constituents and, therefore, does not appear to be justified, particularly since the phases refined are more ductile than those coarsened.

However, it should be noted that the influence of vibration on solidifying metals can be affected by several factors. These include—the peak to peak displacement of the vibrational energy in the melt, the amount of superheat of the metal when poured, the temperature of the mold, the rate of cooling in the mold and the presence of grain refiners or inoculants in the melt. The influence of the vibrational frequency and method of introducing the vibrational energy or grain refinement has been a subject of considerable disagreement.

Vibrational Energy Effect

There is considerable evidence[4,5,9] that if sufficient peak to peak displacement is obtained in the vibrations introduced into the melt, that the frequency and method of introduction is of small consequence. It has been shown, however, that the amount of grain refinement of primary dendrites and crystals[5,8,9] and coarsening of eutectics[8] is increased with greater energy on peak to peak displacement up to a certain point before attaining a maximum and becoming fairly insensitive to further increases.

In order to obtain the full influence of vibration, it is necessary that the vibrational energy have sufficient opportunity to act upon the solidifying melt. Accordingly, if the rate of solidification is too rapid, because of a cold mold or insufficient superheat, little influence will be observed.[5,8]

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TABLE 1 — CHEMICAL ANALYSES OF VARIOUS METALS AND ALLOYS INVESTIGATED

	Al	12% Si Al	4½% Cu Al	Brass	Mg	Zn	Gray Iron	Bi	Sb	Sn
Cu		0.10	4.51	64.56						
Al	bal.	bal.	bal.	<0.05	0.02	0.25				
Si	0.14	12.18	0.28				2.56			
Fe	0.03	0.49	0.28	0.02	0.02	0.002	bal.			
Zn		0.40	0.20	35.35	<0.02	bal.		<0.05	<0.05	<0.02
Mn		0.27	<0.02							
C							3.44			
P							0.087			
S							0.074			
Ti			0.008							
Pb						0.005		0.001	2.25	0.01
Sn				<0.05		<0.001		0.50	0.80	bal.
Sb								0.01	bal.	0.02
Bi								bal.	0.06	0.01

A theory was proposed to explain the influence of vibration on the grain size of the constituents in a solidifying melt.[4] This theory was developed as a result of prior work conducted under the sponsorship of Rodman Laboratory, Watertown Arsenal. In brief, this hypothesis stated that the alternate high and low pressure waves generated in a solidifying melt reduced the size of the critical nucleus necessary for solidification, thereby reducing undercooling and increasing the rate of nucleation.

The pressure wave favored the formation of solid when the solid was more dense than the liquid, and the rarefaction wave would favor the solid when less dense than the liquid because of the Le Chatelier-Braun principle. The former is, of course, the usual case. This theory further states that the rate of growth of the stable nuclei must be sufficiently rapid to permit these to exceed the critical radius for the wave of opposite magnitude.

This investigation was undertaken to provide further data that would substantiate or disprove this theory. In the course of this work, pure metals, and melts which solidify as solid solution and eutectics, were melted and vibrated under closely controlled conditions. If the theory advanced is correct, the amount of grain refinement attained by vibration should be influenced by the difference between the density of the liquid melt and solid metal immediately after solidification. For this reason, metals and alloys that exhibited a considerable range of solidification contraction were selected for this investigation.

PROCEDURE

The metals and alloys that were studied are listed here according to their mode of solidification.

Pure Metals		Solid Solutions	Eutectics
Aluminum	Magnesium	Aluminum-4.5% Copper	Aluminum-12% Silicon
Antimony	Tin	70% Copper-30% Zinc	4.20% Carbon equivalent gray iron
Bismuth	Zinc		

The composition of each metal is listed in Table 1. All pure metals were purchased as commercially pure ingot. The solid solution materials were melted from commercially pure ingot. The eutectics were 13 alloy for the Al-Si, and commercial, eutectic composition gray iron.

The aluminum, aluminum-base alloys, and the gray iron were melted in an acid-lined 10kc, 10kw high frequency induction furnace. The other metals were melted in separate clay-graphite crucibles in a gas-fired furnace. Only the magnesium metal was fluxed during melting. No degassing was conducted. To avoid masking the effect of vibration, no inoculants or grain refiners were added to any metal.

All metals were superheated to 200-300 F over the melting or liquidus temperature immediately prior to pouring, to permit a reasonably slow rate of solidification. Temperature measurements were taken with immersion thermocouples.

Metal Pouring

The metals were poured directly from the melting crucible into a smaller, tapered, clay-graphite crucible used as an ingot mold. These crucibles or molds were heated to approximately 700 F by means of a gas torch before pouring, for all except the lower melting temperature metals (tin and bismuth). The preheating was necessary to prevent rapid solidification. Each melt was split into either two or three equal portions, and each portion was poured into similarly preheated molds.

Each ingot poured was approximately 4 in. in diameter, 5 in. high (63 cu in.). The molds were bolted securely to a metal plate by means of steel clamps. Several series of ingots were poured for each metal.

Vibrations were applied to the solidifying melts by two methods. All metals were subjected to 60 cycle/sec vibrations by bolting one crucible to a metal table mounted over a 300 lb force output electromagnetic vibrator. The arrangement of the crucible on the 60 cycle vibrator is shown in Fig. 1. In addition, the aluminum and zinc were subjected to vibrations from a probe inserted into the melt through the top surface, as shown in Fig. 2. The probe was coupled to a 20kc, 110 watt magnetostriction generator.

The coupler used for vibration of aluminum was a ⅞-in. diameter titanium rod, wetted with aluminum metal before use. A 310 stainless steel ⅞-in. diameter coupling rod was employed for the vibration of zinc.

The molten metal from the induction furnace or melting crucible was poured into two crucibles when only 60 cycle vibration was employed, and into three crucibles when both 60 cycle and 20 kilocycle tech-

niques were utilized. In each case, one crucible was held statically to serve as a control ingot cast under similar conditions for comparison with the vibrated metal. The 60 cycle vibration was started before pouring each ingot. The ingot vibrated at 20kc was poured first, and then the vibrating probe introduced through the top surface of the melt.

This investigation was designed to explore the mechanism of grain refinement by vibration rather than to develop refinements in the technique. Accordingly, when it was established that approximately the same result was obtained with each type of vibration, the majority of the experiments were performed by the 60 cycle method. This latter technique is easily performed and controlled to assure uniform treatment.

Thermocouples were inserted in the approximate center of each ingot during the vibrating and solidification period. These thermocouple wires were contained in two-hole refractory insulators, but the hot junction was bare to provide direct contact with the solidifying melt. Chromel-alumel couples were used for all except the gray iron; platinum - platinum with 10 per cent rhodium was employed because of the higher melting temperature of this latter metal. Cooling curves were recorded on potentiometers throughout the period of cooling and solidification.

After the ingots had cooled to room temperature, they were removed from the crucibles and sectioned longitudinally. The longitudinal surfaces were polished and etched by suitable techniques to develop the grain structure. An area that was representative of the grain size of the ingot was selected and the number of grains determined within this area. Similar locations were selected when comparing the control and vibrated ingots of each metal. In the case of the ingots with a columnar structure, it was necessary to determine the number of grains intercepting a line and convert this to an area measurement.

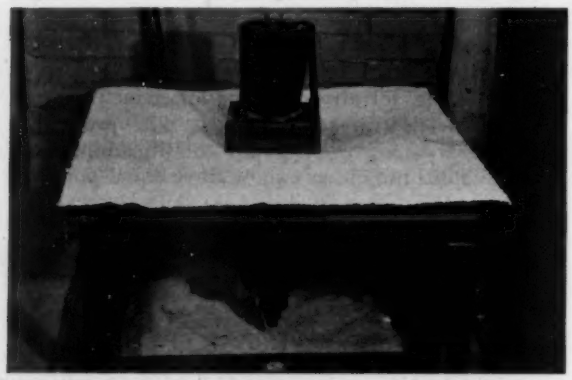


Fig. 1— 60 cycle vibrator and ingot mold arrangement.

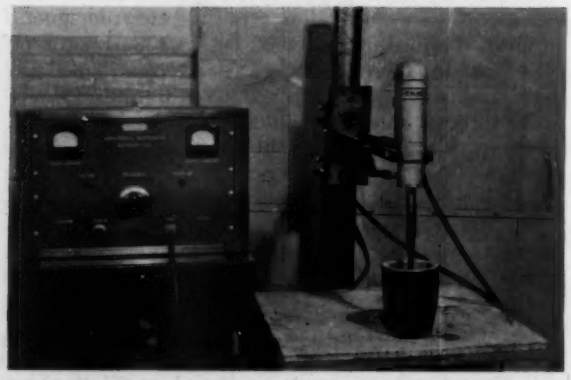


Fig. 2 — Arrangement of equipment for ultrasonic vibration of solidifying metals.

TABLE 2 — GRAIN REFINEMENT FOR VARIOUS SOLIDIFICATION CONTRACTIONS OBTAINED IN THIS INVESTIGATION

Metal or Alloy	Mode of Solidification	Solidification Contraction, %	Grains/Unit Area — Vibrated Grains/Unit Area — Unvibrated
Zinc	Pure metal	6.5	60
Aluminum	Pure metal	6.5	110
Magnesium	Pure metal	4.2	8
Tin	Pure metal	2.7	12
Antimony	Pure metal	1.4	1.8
Bismuth	Pure metal	—3.3	2.3
Al — 12% Si	Eutectic	3.8	3
Gray Iron	Eutectic	—1.5	1.5
Al — 4½% Cu	Solid solution	6.3	77
Brass	Solid solution	5.5	17

RESULTS AND DISCUSSION

The grain refinement obtained by vibration of each of the prime metals, solid solution and eutectics is shown in Table 2. This table lists the mode of solidification of the metal or alloy, amount of solidification shrinkage obtained with this material and a ratio of the number of grains contained in a given area of the vibrated ingot compared to the number of grains contained in a similar area of the unvibrated or control ingot.

In order to demonstrate the influence of vibration on the grain refinement of the pure metals, photomacrographs of representative, etched halves of the control 60 cycle and 20 kc vibrated ingots for aluminum and zinc are shown in Fig. 3. The macroetched ingots for the other commercially pure metals, antimony, bismuth, magnesium and tin, in the vibrated and unvibrated condition are shown in Fig. 4. The vibrated and control macrostructure of the alloys that solidify as solid solutions (Al-4.5 per cent Cu and 70 per cent Cu-30 per cent Zn), and the microstructure of the eutectics (Al-12 per cent Si and gray iron) are illustrated in Figs. 5 and 6, respectively.

In this latter case (Fig. 6) the eutectic cells were considered to be individual grains. The structure was so fine that microexamination is required to show the influence of vibrations on eutectics. Attention is called to the coarser eutectic constituents which accompany the finer cell structure of the vibrated ingots. The reason for this structure was previously described.

According to the theory that was postulated,[4] vibrations introduced into a solidifying melt would increase the rate of nucleation of solid phases in this melt because alternate high and low pressure waves are produced in the liquid. The high pressure wave would favor the existence of the most dense state, usually the solid state. The low pressure wave favors the lower density state, usually the liquid, but in some instances the solid, state.

It is also necessary to postulate that any stable nu-



Fig. 3 — Macrostructures of zinc (top row) and alumnimum (bottom row) ingots vibrated at 20 kc (left), 60 cycles (center) and unvibrated (right). Etched. 0.5 ×

cleus existing at a pressure wave would have to grow sufficiently rapidly, so that the critical nucleus size for the following wave of opposite pressure was exceeded before this latter wave reached this location.

If a series of experiments were conducted in which the significant variables for grain refinement were maintained within close limits, the theory previously postulated[4] predicts that the amount of grain refinement would be dependent on the difference between th density of the liquid and solid metal at the solidifying temperature. The known significant variables influencing the effect of vibration on grain refinemet were held reasonably constant in this investigation, and the effect of vibration under these controlled conditions was observed on numerous metals.

The resulting grain refinement of these pure metals, solid solutions and eutectics is plotted vs. per cent of solidification contraction in Fig. 7. In this graph, the amount of grain refinement or number of grains in a given area of the vibrated ingot, divided by the number of grains in a similar area of the control ingot, is shown on a logarithmic scale in order to highlight the metals in which only slight grain refinement occurs.

The data as plotted in Fig. 7 and listed in Table 2 show clearly that the amount of grain refinement under the controlled test conditions is dependent upon the amount of solidification contraction. Only slight refinement was experienced with antimony, and only slightly more with bismuth. Magnesium exhibited somewhat more refinement, 70 per cent Cu-30 per cent Zn still more and a large refinement is shown for aluminum, zinc and Al-4.5 per cent Cu. Tin was refined considerably more than any other metal or alloy with a similar amount of solidification contraction.

Some difficulty was experienced in measuring the grain sizes, but the best approximation is a refinement of one to 12, as shown in Fig. 7 and Table 2. No explanation for this different behavior is offered at this time. The grain refinement increases steadily as the change in volume upon solidification is raised from 1.5 to 6.6 per cent. The effect of vibration on the cell size of the eutectics is also approximately of this same order.

Gray iron, with only a small solidification expansion, is refined only slightly, and the cell size of Al-12 per cent Si with a 3.8 per cent solidification contraction exhibits somewhat more refinement. The refinement of these eutectics represents a special case that will be discussed subsequently.

In order to obtain a comparison of data obtained in this investigation with the other workers in this field literature was reviewed, and the grain refinement reported by numerous investigators were added to Fig. 7.

These refinement figures were limited to publications that either reported the refinement or contained photographs of sufficient clarity to permit these measurements.

Data were limited to those papers that employed no grain refiners and had sufficient energy for refining with enough superheating and mold preheat to allow time for the vibrational energy to act before solidification was complete. The details of the reference work, composition, type of vibration, grain refinement and solidification contraction for all data employed are listed in Table 3. It is noted that all of this work from these several references[1,2,4,5,6,12] agrees closely with the grain refinement vs. solidification contraction information from this investigation.

The thermal data for the most part were unsucces-

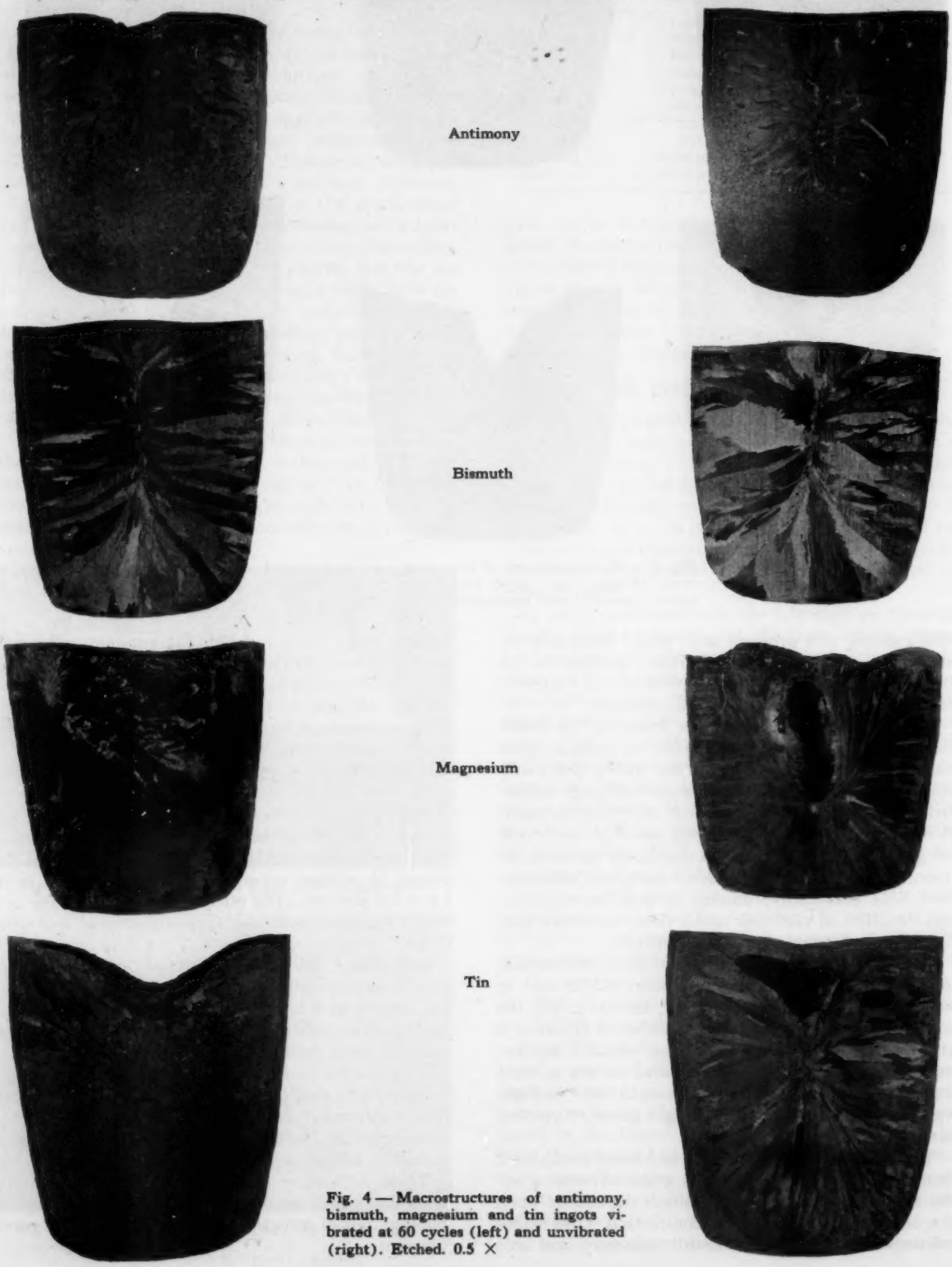
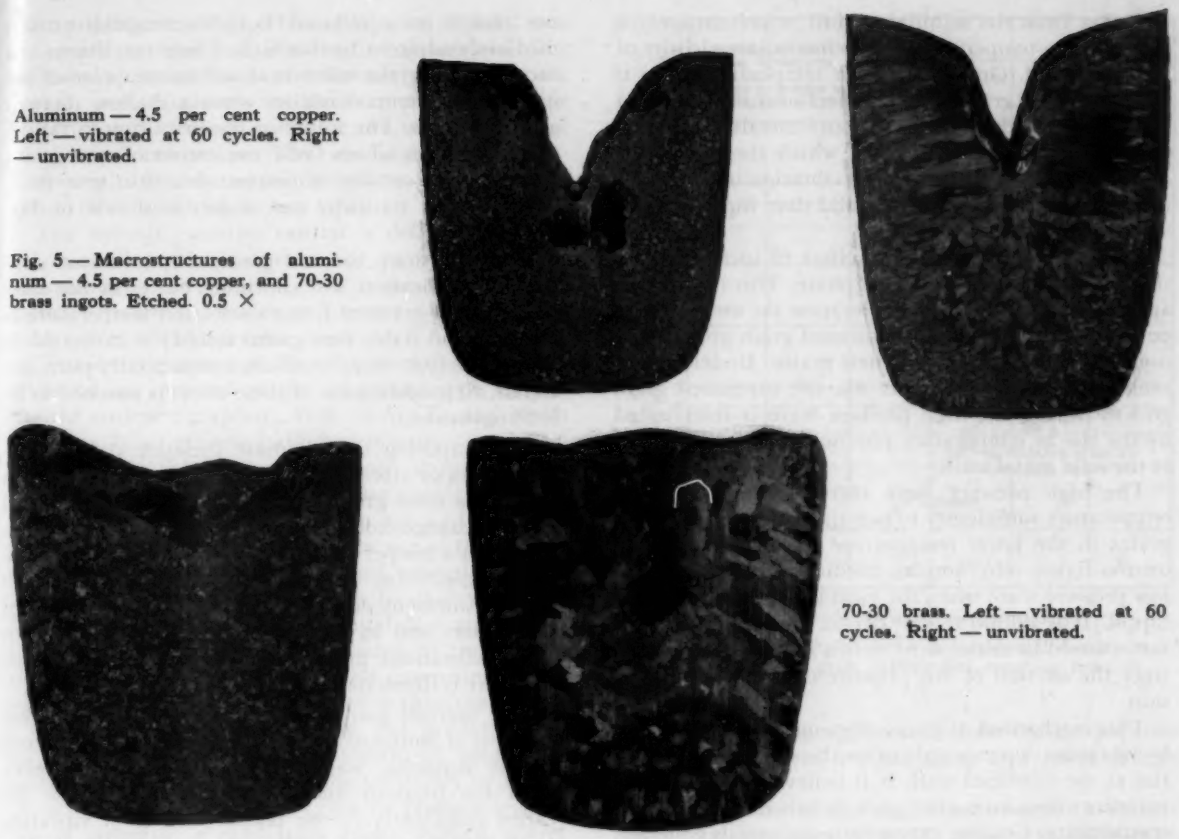


Fig. 4 — Macrostructures of antimony, bismuth, magnesium and tin ingots vibrated at 60 cycles (left) and unvibrated (right). Etched. 0.5 ×



Aluminum — 4.5 per cent copper. Left — vibrated at 60 cycles. Right — unvibrated.

70-30 brass. Left — vibrated at 60 cycles. Right — unvibrated.

Fig. 5 — Macrostructures of aluminum — 4.5 per cent copper, and 70-30 brass ingots. Etched. 0.5 ×

ful in obtaining any consistent difference between the vibrated and control ingots. This condition was attributed to the fact that differences in thermal behavior were so small that it was less than the sensitivity of the thermocouples and instruments. Definite indications were obtained, however, on bismuth, zinc and both eutectics, that the vibrated ingot undercooled somewhat less (2 to 6 F) than the control or unvibrated ingot before solidification was initiated.

The length of time required for complete solidification of the ingots was longer for the vibrated specimens in some cases and the unvibrated ingots in others. No consistent results of longer[8] or shorter[13] solidification times, such as reported in the Russian literature, were observed in this investigation.

MECHANISM OF GRAIN REFINEMENT

Pure Metals

Since the metals employed were of commercial purity, it is anticipated that these metals will solidify by heterogeneous nucleation. Even under these circumstances, however, a small amount of undercooling is necessary for stable nuclei to form, so that solidification can proceed.[14] The mechanism by which this solidification occurs to produce the grain refinement observed in pure metals could possibly be explained in a manner similar to solid solution solidification because of the impurities present.

If it is assumed, however, that the metals are of sufficient purity to avoid appreciable concentrations of solute in the liquid metal near the solidifying wall, then another mechanism must be found.

It has been postulated that the thermal conditions during the static solidification of the columnar zone in an ingot of pure metals are shown in Fig. 8a.[14] Figures 4 and 5 illustrate that columnar grains were formed for all the pure metals cast without vibration. These grains were obtained even though the amount of superheat and mold preheat were not conducive to high thermal gradients within the solidifying ingot.

The growth of the columnar grain continues in from the wall, since less undercooling is necessary for continued growth of the columnar grains than for nucleation of new grains and the thermal gradient in the liquid is positive.[14] Under these conditions, the explanation of the grain refinement of pure metals becomes difficult. However, this refinement, as shown in Figs. 4 and 5 of this paper, and in references[1,2] does occur.

It is believed that the thermal gradient in the liquid is not positive throughout the liquid, but that the heat

TABLE 3 — GRAIN REFINEMENT FOR VARIOUS SOLIDIFICATION CONTRACTIONS OBTAINED FROM OTHER INVESTIGATIONS

Metal or Alloy	Mode of Solidification	Solidification Contraction, %	Grains/Unit Area — Vibrated Grains/Unit Area — Unvibrated	Source
Magnesium	Pure metal	4.2	9	(2)
Al — 12% Si	Eutectic	3.8	5	(4)
Al — 4½% Cu	Solid solution	6.3	100	(5)
Brass	Solid solution	5.2	20	(4)
310 Stainless Steel	Solid solution	3.0	2 — 3	(6)
4340 Steel	Solid solution	2.8	2	(12)

of fusion from the solidifying chill or columnar zone increases the temperature in the immediate vicinity of this solid wall. Once this higher temperature zone is postulated, the grain refining effect or interruption of columnar growth observed in pure metals can be explained. The conditions under which the nucleation of new grains will occur during vibration of a solidifying pure metal with a greater solid than liquid density are shown in Figure 8b.

The pressure wave has the effect of increasing the stability of the solid or denser phase. This means that applied vibrations effectively increase the melting temperature, temperature for continued grain growth and nucleation temperature for new grains. Under the arrangement shown in Figure 8b, the continued grain growth during the high pressure wave is interrupted by the rise in temperature produced by heat of fusion at the solid metal wall.

The high pressure wave increases the nucleation temperature sufficiently to permit nucleation of stable grains in the lower temperature region, as indicated on the figure (8b). Similar conditions prevail for the low pressure wave when the solid is less dense than the liquid. It is doubtful whether as much grain refinement would be anticipated in this latter case, however, since the amount of low pressure is limited by cavitation.

This mechanism of grain refinement of pure metals by vibration depends only upon the slight temperature rise at the solidified wall. It is believed that this temperature rise occurs since pure metals melt at a single temperature. Cooling curves for pure metals generally show this temperature rise as solidification takes place.

Solid Solutions

The solidification of solid solutions occurs with a concentration of solute immediately adjacent to the solidifying wall of dendrites. This concentration of solute results in a reduced liquidus temperature immediately adjacent to this wall. These conditions can account for a grain refinement and interruption of the growth of column dendrites when a shallow thermal gradient exists. The solute or impurity concentration, and conditions which favor nucleation of new grains in the melt so that columnar dentritic growth is stopped in a statically cast ingot, is shown in Fig. 9a.[1]

With the sharp thermal gradient t_1 columnar dendritic solidification will continue, but with the shallow thermal gradient t_2 the nucleation temperature is attained and stable new grains solidify as indicated in this figure (9a). Since these are commercially pure materials, all solidification of these alloys is assumed to be heterogeneous.

Superimposing the alternate pressure and rarefaction waves of vibration upon this system will tend to produce a finer grain sooner or, in other words, will cause a change from columnar to equiaxed solidification with a steeper thermal gradient than in the unvibrated condition. Experimental evidence that this grain refinement does occur is contained in Fig. 5 of this report and in references.[2-7] The mechanism by which vibrations produce this finer, more equiaxed structure is illustrated in Fig. 9b.

The thermal gradient t_1 is too steep to permit nucleation of equiaxed grains away from solidifying columnar dendritic wall without vibration. However, both the liquidus and nucleation temperature are raised sufficiently by the pressure waves of vibration so that nucleation of small, stable equiaxed grains occurs at the location shown in Fig. 9b.

Eutectics

The somewhat more complicated mechanism by which eutectics solidify has been reviewed in two re-

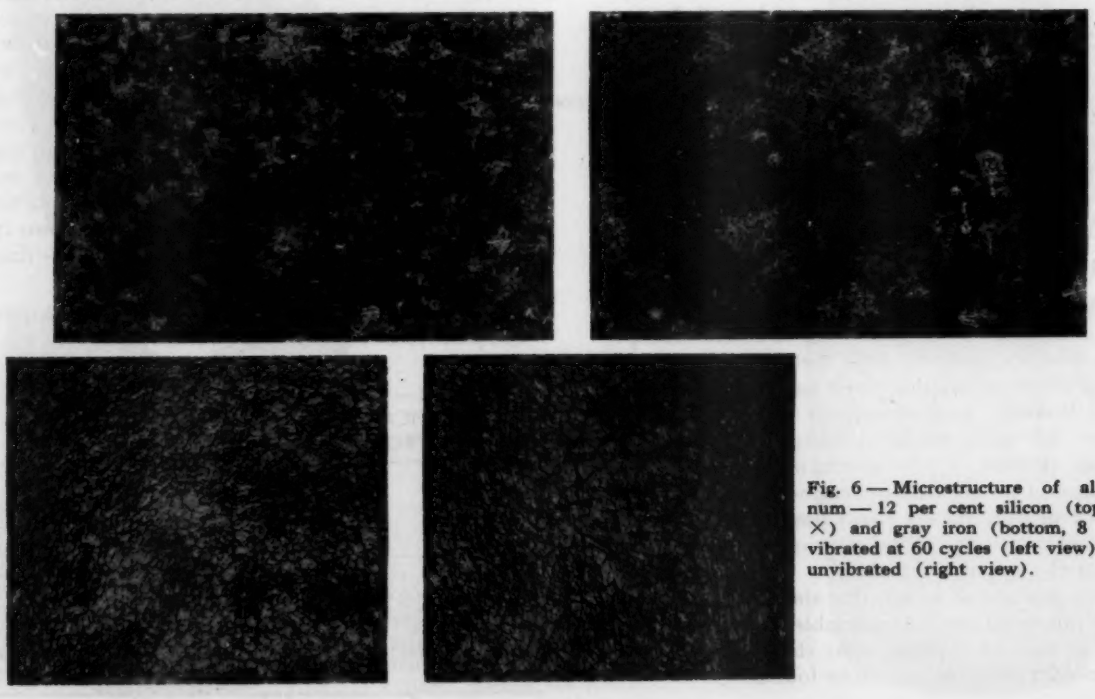


Fig. 6 — Microstructure of aluminum — 12 per cent silicon (top 10 ×) and gray iron (bottom, 8 ×), vibrated at 60 cycles (left view) and unvibrated (right view).

cent publications.[15,16] These alloys are divided into two types—normal and abnormal or anomalous. The primary phase nucleates the secondary phase in the former case. However, the primary phase is unable to serve as nucleating agent for the secondary phase in the analomous type. This seemingly small difference results in substantial deviations in the solidification mechanisms of the two types.

The normal eutectics exhibit a definite crystallographic orientation between the two phases. The anomalous eutectics do not exhibit this regular orientation between the major and minor phases and usually undercool to a greater extent.

This investigation and the other work that has been reviewed are confined to the anomalous type, including the austenite-graphite, Al-Si, Al-Al_3 Fe, and Pb-Sb eutectics. In all of these published investigations[3,4,5,8] as well as in this work, it has been shown that vibration will coarsen the eutectic constituents.

In addition, it has been shown in Fig. 6 of this work and reference[4] that the cell or grain size of the austenite-graphite and Al-Si eutectics are refined and the amount of undercooling reduced by this vibration. In view of the fact that experimental data are limited to anomalous eutectics, the discussion of the mechanism of refinement of the cells and coarsening of the eutectic constituents will be limited to this type.

The main nucleation problem, and primary cause for undercooling in the solidification of anomalous eutectics, is the nucleation of the secondary or minor phase.[15,16] Since the primary eutectic phase is incapable of acting as a nucleation agent, foreign nuclei from the melt must provide these sources for heterogeneous solidification.

If the major phase is considered to have been previously nucleated from the eutectic liquid, this solid particle is surrounded by a liquid at a higher temperature than the unaltered liquid eutectic because of heat of fusion and with a higher concentration of the second phase. When this liquid becomes sufficiently supersaturated with the minor phase, nucleation of this phase occurs at a suitable foreign particle. By the application of vibration to this solidifying eutectic, the melting and nucleation temperature of the supersaturated liquid surrounding the major phase is effectively raised in a manner similar to that described for the pure metals and solid solutions.

Vibrations are capable of assisting nucleation of the minor phase immediately adjacent to (although probably not in contact with) the stable nucleus of the major phase. By this action, vibration of a solidifying anomalous eutectic reduces undercooling and increases the number of stable nuclei of the minor phase. This latter action results from the fact that more foreign nuclei are rendered effective by increasing the supersaturation or nucleation temperature.

The effect of vibrating this type of eutectic is similar to inoculation. The number of eutectic cells is increased and the undercooling is reduced. As a result the eutectic solidifies much more slowly than in the static uninoculated state, because the heat of fusion quickly overcomes the slight undercooling and solidification becomes dependent on the rate of heat removal from the ingot. Under these conditions, the eutectic consitituents agglomerate into coarser particles with their reduced surface areas and lower energy state.

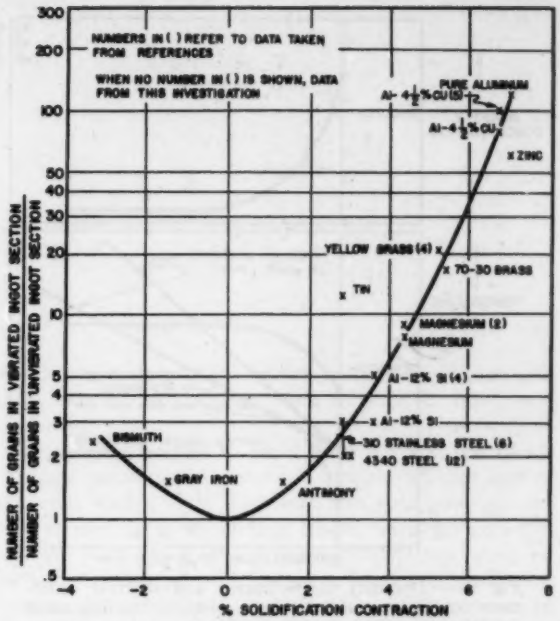


Fig. 7 — Effect of difference in liquid-solid density at melting point on grain refinements resulting from vibrations.

The selection of the significant change in density upon solidification of a eutectic poses a problem. The amount of refinement is measured in terms of the number of eutectic cells or grains. However, the minor phase is generally the difficult phase to nucleate, and it appears that vibration exerts its major influence on eutectic solidification by assisting in the nucleation of this minor phase. For this reason, some considera-

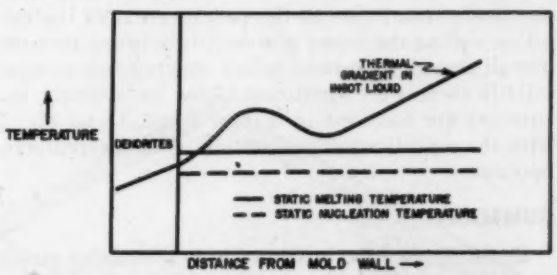


Fig. 8a — Thermal conditions in ingot during solidification of the columnar zone in a pure metal without vibration.[14]

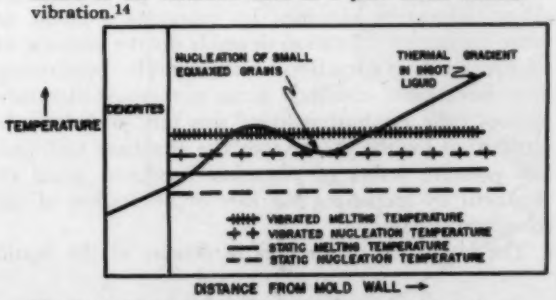


Fig. 8b — Influence of vibration on melting and nucleation temperature in a pure metal and nucleation of equiaxed grains.

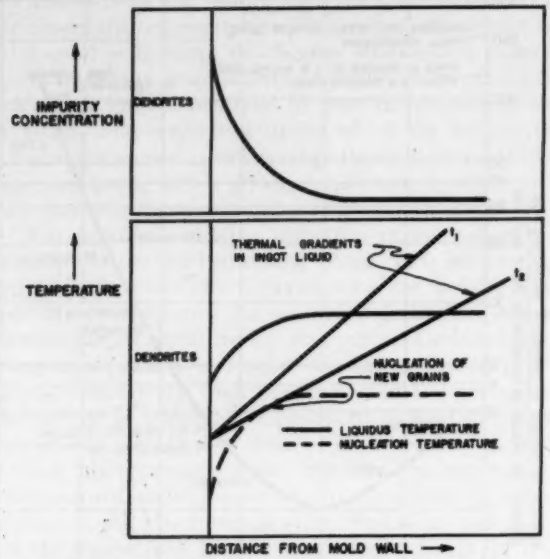


Fig. 9a — Impurity concentration and thermal conditions during transition from columnar to equiaxed solidification without vibration.[14]

tion must be given to the significance of the density change involved in the solidification of this minor phase.

It is noted that solidification contraction of the gray iron eutectic varies from a slight expansion to a small contraction, depending on the silicon content. The solidification of graphite from the liquid, however, involves a large expansion. The Al-Si eutectic solidifies with a considerable contraction (3.8 per cent)), whereas silicon solidifies from eutectic liquid with a slight expansion.

In view of the fact that the actual grain size measured is that of the eutectic cells, and vibration undoubtedly contributes to the solidification of the major as well as the minor phase, it is believed that the overall change in volume which occurs upon eutectic solidification is the significant factor. Accordingly, the eutectics are included in Tables 2 and 3 and Fig. 7 with the solidification contraction of the overall composition.

SUMMARY

By the application of vibration to solidifying metals including pure metals, solid solutions and eutectic compositions, it has been shown that the amount of grain refinement obtained by vibrating a metal or alloy during solidification depends on the amount of solidification contraction. Considerable confirming data have been obtained from numerous literature sources. The establishment of this fact provides substantiation for the theory that the alternate high and low pressure waves of vibration produced grain refinement by increasing the rate of nucleation of the solidifying phases.

The larger the differences in density of the liquid and solid phases, the greater the influence of the pressure waves. The mechanism whereby this increased rate of nucleation can refine the grain size of solidifying pure metals, solid solutions and anomalous eutectics is presented.

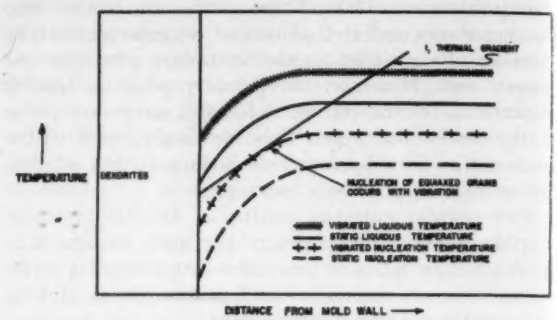


Fig. 9b — Influence of vibration on liquidus and nucleation temperatures in a solid solution and nucleation of equiaxed grains.

ACKNOWLEDGMENT

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REFERENCES

1. P. D. Southgate, "Action of Vibration on Solidifying Aluminum Alloys," *Journal of Metals, vol. 9,* p. 514 and references, 1957.
2. J. B. Jones, "Ultrasonics Applied to Solidification and Solid State Transformation," *Af Technical Report No. 6675,* WADC, 1951.
3. W. Rostoker and M. J. Berger, "Effects of Vibration During Solidification of Castings," *Foundry, vol. 81,* p. 100, July 1953.
4. A. H. Freedman and J. F. Wallace, "The Influence of Vibration on Solidifying Melts," AFS TRANSACTIONS, *vol 65,* p. 578, 1957.
5. R. S. Richards and W. Rostoker, "The Influence of Vibration on the Solidification of an Aluminum Alloy," *A.S.M. Transactions, vol. 48,* p. 884, 1956.
6. Y. B. Guverich, et al, "Effect of Ultrasound on Structure and Properties of Steel Ingots," *Stal', vol. 17,* p. 406, 1957.
7. N. P. Nikolaichik and E. N. Nikolaichik, "Quality of Steel Ingots Improved by Ultrasound Treatment," *Stal', vol. 17,* p. 322, 1957.
8. V. V. Zaboleev and G. I. Pogodin-Alekseev, "Effect of Ultrasonic Vibrations on the Structures of Eutectic Alloys," *Metallovedenie: Ob. Met.,* Jan., p. 2, 1958.
9. G. Schmid and A. Roll, "The Effect of Intense Sound Waves on Metallic Melts," *Z. Electro-chem, vol. 45,* p. 796, 1939.
10. G. Schmid and A. Roll, "Ultrasonics – Their Effect on Molten Zinc," *Metal Industry, vol. 60,* No. 15, p. 252, April 10, 1942.
11. I. I. Teumin, "Theory and Calculation of Induction of Elastic Oscillations Molten Metals," *Problems of Metallography and Metal Physics,* 4th Collection of Papers, p. 50, 1955.
12. Data from unpublished research sponsored at Case by Rodman Laboratory, Watertown Arsenal.
13. S. Sokoloff, "The Effect of Supersonic Waves on the Solidification of Molten Metals," *R.T.P. Translation,* No. 2568, 1958.
14. J. L. Walker, "Structure of Ingots and Castings," and references, *Liquid Metals and Solidification,* A.S.M., p. 319, 1958.
15. W. A. Tiller, "Polyphase Solidification," *Liquid Metals and Solidification,* A.S.M., p. 276, 1958.
16. A. de Sy, "Considerations of the Mechanism of the Solidification of Gray Iron," AFS TRANSACTIONS, *vol. 67,* 1959.

RECLAIM SODIUM SILICATE BONDED SAND

By George C. Warneke

ABSTRACT

The objective of this project was to determine whether or not sodium silicate bonded sands could be reclaimed and, if so, what would be involved, and what could be expected as to repetitive dimensional accuracy of the sodium silicate bonded mold. A pneumatic scrubber and an aerated muller were used in the reclamation tests. A casting from green sand was compared for dimensional accuracy with a casting from sodium silicate bonded sand.

INTRODUCTION

The sodium silicate process has been in use for some time. To obtain the answers to two objectives which it was felt had never been thoroughly investigated, a project was initiated to find out whether or not sodium silicate bonded sands could be reclaimed, and if so what would be involved and what could be expected in repetitive dimensional accuracy of a sodium silicate bonded sand mold.

The sodium silicate process has had many pros and cons. Much publicity has been given the cost and savings factors of molds and cores free from any sag or distortion, and the freedom from use of driers, ovens and large storage areas, as well as the personnel to service them. One factor, and a basic and costly one, that it was felt had never been investigated was the basic core or mold sand used.

If this sand had to be discarded after each use, the use of sodium silicate as a bonding agent could become economically prohibitive, even though the process was of value in other areas.

PROCEDURE

In order to produce castings from sodium silicate bonded sands at the lowest economic level, the sand necessarily must be reclaimable. The casting chosen to determine the merits and/or faults of a reclaimed sodium silicate bonded sand was the Steel Founders' Society of America standard test block. Figure 1 is a sketch of this test block casting.

The mold is shaped and gated to produce severe conditions against the sand interface. The casting is gated into the thin wing causing the metal to cover, uncover and re-cover the bottom of the thin wing area before running over the edge and subjecting itself to the side and bottom grooves. This casting was used to evaluate the relative erosion, penetration and burn-on conditions of the reclaimed sodium silicate bonded sand after each repetitive sand reclamation.

The first mix used to mold these S.F.S.A. test block castings contained 6 per cent sodium silicate and 2 per cent kaolin clay. Although this mix gave an excellent casting as to surface finish, with no penetration or burn-on on the bottom of the thin wing or in the bottom of side grooves, shake-out was extremely difficult. This sand with the high percentage of sodium silicate was shaken out in the form of large, hard chunks. Upon crushing these large chunks the sand grains were fractured making this mix with a high sodium silicate content unsatisfactory for reclamation, or from the standpoint of shake-out properties.

The next step was to lower the sodium silicate content, so that the shake-out properties would be improved and not create any difficulties, as well as to produce a sand which could easily be crushed, a necessary step between shake-out and reclamation. A figure of 4 per cent sodium silicate was chosen, as this amount, or slightly less, is used in many foundries for normal sodium silicate core and mold production.

The S.F.S.A. test block castings which resulted from this mix were not 100 per cent defect free as were the 6 per cent castings. However, these 4 per

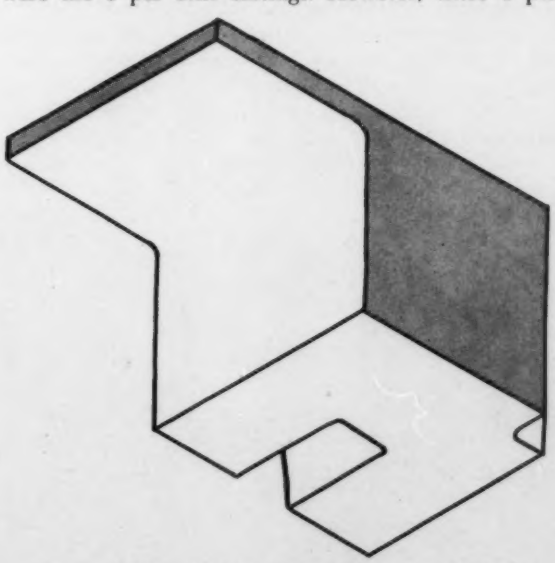


Fig. 1 — Sketch of test block casting used for this study.

G. C. WARNEKE is Foundry Engr., National Malleable & Steel Castings Co., Melrose Park Works, Melrose Park, Ill.



Fig. 2 — Casting made from new Ottawa sand having an AFS Grain Fineness Number of 48.41.

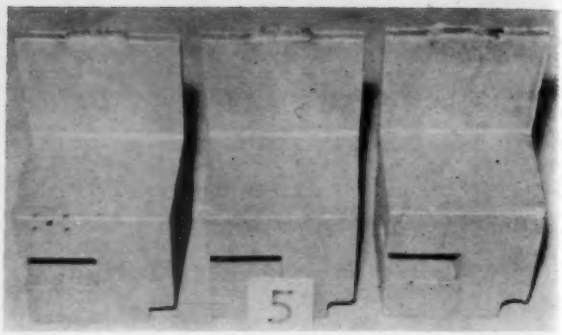


Fig. 3 — Castings made after the sand's first reclamation in the pneumatic scrubber.

cent castings could easily be used as original sand castings, for comparison purposes, with castings made from the same sand after reclamation.

Figure 2 shows the castings made from new Ottawa sand having an AFS Grain Fineness Number of 48.4. The additives used were 4 per cent sodium silicate and 2 per cent kaolin clay. Note the erosion on the bottom of the thin wing, although the bottom and side grooves are free of defects. These were the standard castings against which test block castings, made with the same sand after its reclamation, were to be compared. These castings made from a mix with a 4 per cent sodium silicate content were easily shaken out, and the sand was readily crushed without grain fracture.

RECLAMATION PROCESSES

The 4 per cent sodium silicate bonded sand was then divided into two equal parts and reclaimed by two different methods:

1) In a pneumatic scrubber.
2) In an aerated muller.

Reclamation by the pneumatic scrubber was done at the rate of 1000 lb/hr in a two-compartment unit. Six groups of castings were made; the original, and a group after each of the five reclamations.

The sodium silicate bonded sand used to make these castings was cured with carbon dioxide, the gas entering the mold through slotted vents in the pattern plate from a plenum chamber under the pattern plate. Before each re-use of the sand, it was tested for usable sodium silicate; none was found at any time. The casting to sand ratio was 1 to 3; perhaps with a higher ratio some sodium silicate may be available for re-use. Twenty per cent new sand was added to each batch of reclaimed sand.

This was done to accomplish a two-fold purpose:

1) To make up the approximate 15 per cent loss shared between the reclaimer and the crusher, where $\frac{1}{8}$-in. "peanuts" were screened from the sand after crushing.
2) To be able to vary the Grain Fineness Number of this 20 per cent addition, so when it was blended with the returned reclaimed sand the fineness number of the reclaimed sand could be kept in the range of the original new sand.

Pneumatic Scrubber Reclamation

In Fig. 3 are castings made from the sand after its first reclamation in the pneumatic scrubber. As can be seen, the castings appear similar to the original new sand castings as to surface finish on the bottom of the thin wing, the same erosion condition in area and degree of severity and the absence of defects in the bottom and side grooves.

Figure 4 shows castings made from the sand after its second reclamation in the pneumatic scrubber. They are identical with original castings, and with castings made after the first reclamation.

In Fig. 5 are castings made after the sand's third reclamation. They are slightly improved over the original castings with regard to the erosion condition at the bottom of the thin wing. No explanation is offered as to why this situation occurred. The sand, as to grain fineness and distribution, was virtually the same as when it was used to make the original castings.

The foundry practice was duplicated as closely as

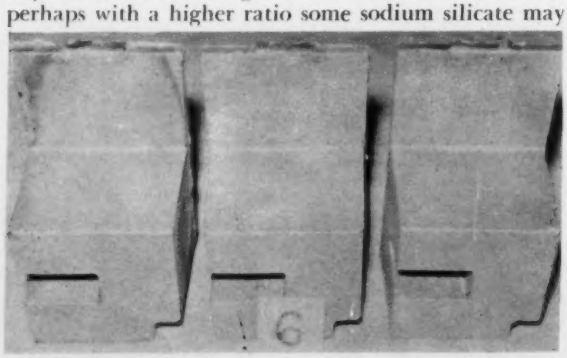


Fig. 4 — Castings made after second sand reclamation in the pneumatic scrubber.

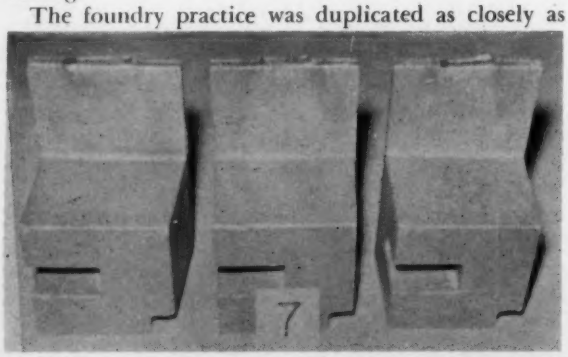


Fig. 5 — Casting made after third reclamation of sand in the pneumatic scrubber.

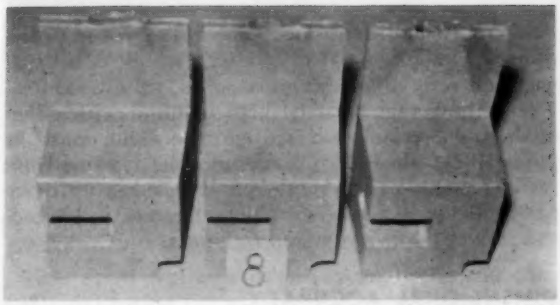


Fig. 6 — Castings made after fourth reclamation of sand in the pneumatic scrubber.

possible, as was the pouring practice. The metal temperature was the same as used in the original castings.

Figure 6 shows castings made from sand after the fourth reclamation. Again, these castings are similar to the original new sand castings, having the erosion condition on the bottom of the thin wing.

In Fig. 7 are castings made after the sand's fifth reclamation. Again, these castings are similar in appearance to the original new sand test block castings.

Conditions that were discernable in looking at the actual castings, and which could not be reproduced well on the photographic negatives, were the slight differences between the groups of castings even though they may be said to appear similar. In all cases, except one, a general improvement in surface conditions of a group of castings could be traced to that group of castings coming from a sand that had a slightly higher grain fineness number.

Aerated Muller Reclamation

The second method of reclamation employed an aerated muller. This was done at the rate of 350-lb batches mulled and aerated for 20 min, and discharged through a cyclone. Again, six groups of S.F.S.A. test block castings were made—the original with new sand, and a group after each of the five reclamations of the sand.

These groups of castings were made simultaneously with groups of castings from sand reclaimed in the pneumatic scrubber, in order that molds made with sand after its first reclamation in the aerated muller, and molds made with sand after its first reclamation in the pneumatic scrubber, could be poured in the same heat.

This procedure was followed with the second, third, fourth and fifth reclamations by each method. The same foundry practice, bond additions, blending procedure and sand mulling practice was used as was used in the sand that was reclaimed by the pneumatic scrubber. This procedure was followed to minimize the many variables.

Figure 8 shows castings made with the sand after its first reclamation in the aerated muller. The surface appearance on the bottom of the thin wing is slightly better than that on the original new sand castings. This is attributed to the fact that this sand had a grain fineness number 7 points higher than that of the original sand. It is felt that this was caused by the negative pressure not being high enough in the muller, and to the many fines retained in the sand.

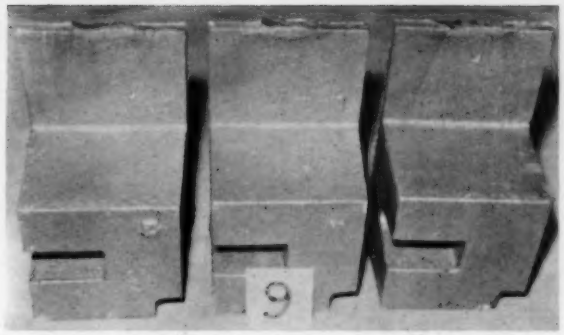


Fig. 7 — Castings made after fifth sand reclamation in the pneumatic scrubber.

In Fig. 9 are castings made with the sand after its second reclamation in the aerated muller. In appearance they are similar to the original new sand castings, having the slight erosion condition on the bottom of the thin wing.

Figure 10 shows castings made with the sand after its third reclamation. Notice the improved surface condition. These castings from reclaimed sand are mates to the castings from sand reclaimed in the pneumatic scrubber. As noted previously, even after extensive checking, no cause can be found for these castings having the improved surface after the third reclamation by either method.

Figure 11 shows castings made from the sand after its fourth reclamation. As can be seen, these castings have the same general surface conditions, with the slight erosion condition in the bottom of the thin wing and the absence of defects in the bottom and side grooves, as did the original new sand castings.

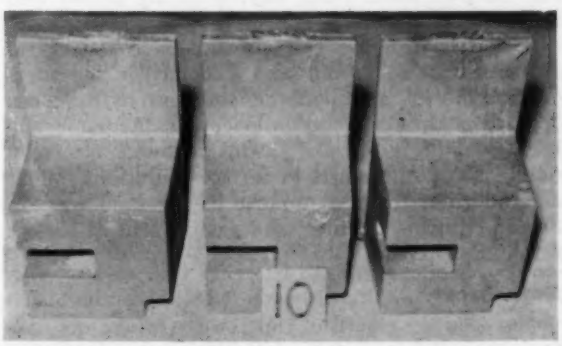


Fig. 8 — Casting made after the sand's first reclamation in the aerated muller.

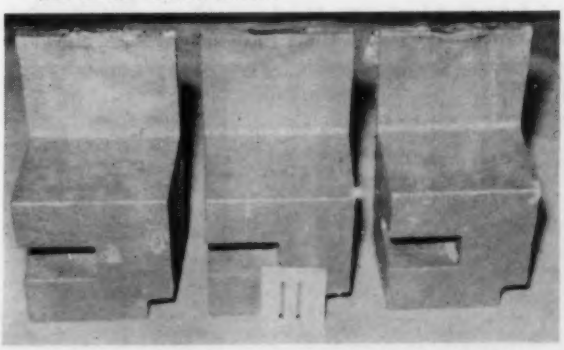


Fig. 9 — Castings made after second sand reclamation in the aerated muller.

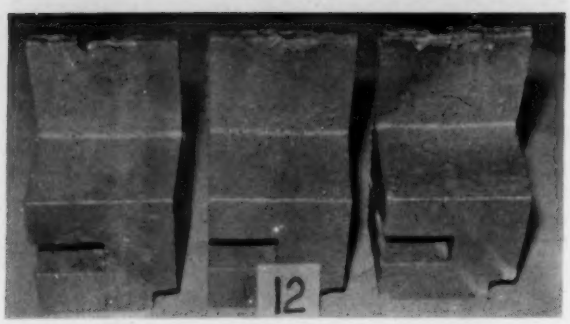


Fig. 10 — Castings made after third reclamation of sand in the aerated muller.

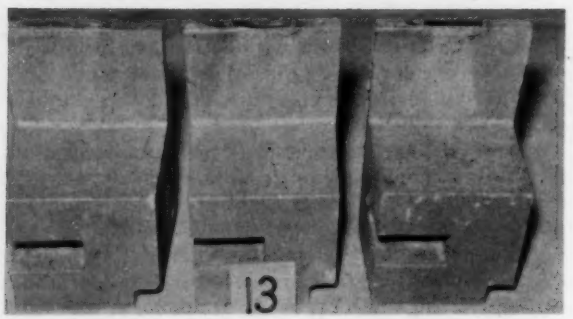


Fig. 11 — Castings made after fourth sand reclamation in the aerated muller.

In Fig. 12 are castings made from the sand after its fifth reclamation in the aerated muller. Again, the castings have a similar surface appearance on the bottom of the thin wing as did the original castings and castings made from sand after its second and fourth reclamations.

A similar situation existed with these castings as did with the others from sand reclaimed in the pneumatic scrubber, although the photographs have not detailed slight differences too closely. Actual examination of the groups of castings showed a correlation of slight differences in surface conditions following the grain fineness number of the sand used in the experiments. A finer sand gave a general surface finish improvement.

Conclusions

Based on the results of this project, where the base Ottawa sand was reclaimed five times in an aerated muller and a pneumatic scrubber, and used to cast S.F.S.A. standard test blocks (used to compare the reclaimed sand with the original new sand), it appears that sodium silicate bonded sands can be reclaimed in an aerated muller or a pneumatic scrubber.

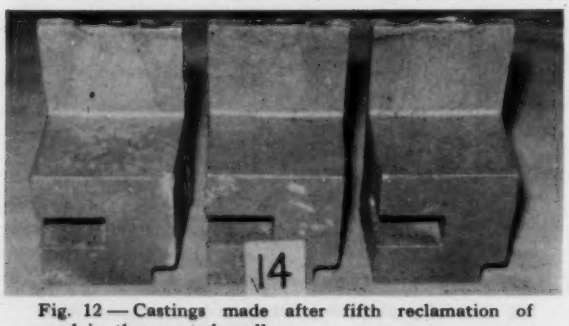


Fig. 12 — Castings made after fifth reclamation of sand in the aerated muller.

One thing that must be kept in mind with either method of reclamation is that the sand must be physically crushed after shake-out. The sand will not readily crumble as do most foundry sands after being passed through a sand tumbler and returned either to a storage hopper or a sand reclaimer.

DIMENSIONAL ACCURACY

This section deals with the dimensional accuracy that can be obtained from a sand mold using a sodium silicate binder. By "dimensional accuracy" is meant repetitive dimensional accuracy from any size lot of repetitive productive castings.

Figure 13 shows the casting chosen to be examined dimensionally in six different places. It is a production casting normally cast in green sand.

Figure 14 is a schematic view of the casting which was used to compare the repetitive dimensional accuracy of sodium silicate bonded molds with that of the green sand molds. As can be seen, dimensions were taken at *A*, between a pair of vertical inside walls of two pockets and at *B*, (90 degrees to dimension *A*) between a pair of outside vertical walls. Dimension *C* is comparable to dimension *B* except that it is in the cope. Dimension *D* is the overall outside length of the mold.

Dimension *E* is the depth of the drag from the parting line to the lowest point, and dimension *F* is an overall dimension of cope and drag total height. The dimensions involved cover all areas where the castings may be distorted either in molding or pouring operations. A total of 84 of these castings was made in sand molds bonded with sodium silicate. A similar number of regular production green sand castings were selected at random to be compared with these castings from sodium silicate bonded sand molds.

Dimension *A* on castings from sodium silicate bonded sand molds averaged 7.6 in., all dimensions being taken with a vernier caliper. When all 84 dimensional readings were subjected to statistical analysis, the average variation which could be expected 95 per cent of the time was a tolerance of ± 0.016 in. of the average dimension. This was established by analyzing the data sheets, and taking the average tol-

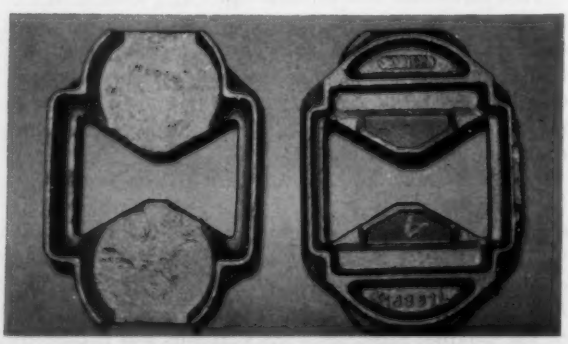


Fig. 13 — Casting chosen to the examined dimensionally in six different places. It is a normal production casting usually cast in green sand.

erance plus or minus twice the standard deviation, being 0.008 in.

Dimension *A* on castings made by conventional green sand molding methods had an average of 7.615 in. When all 84 readings of dimension *A* of the green sand molded castings were subjected to statistical analysis, the average variation which could be expected 95 per cent of the time was a tolerance of ±0.028 in. of the average dimension.

This again was established by analyzing the data sheets and taking the average tolerance plus or minus twice the standard deviation of green sand, being 0.014 in. Therefore, the use of a sodium silicate bonded sand in preference to conventional green sand would permit the molding of this 7⅝ in. dimension with the range of variation cut approximately in half.

Dimension *B*, at 90 degrees to dimension *A*, on castings from sodium silicate bonded molds averaged 7.901 in. Again, when all 84 of these dimensional readings were subjected to statistical analysis, the average variation to be expected 95 per cent of the time was a tolerance of ±0.016 in. of the average dimension. This same dimension on the green sand castings averaged 7.874 in. The statistical analysis showed the average variation to be expected 95 per cent of the time was a tolerance of ±0.028 in.

As in the previous case, molding with a sodium silicate bonding agent would permit holding as-cast tolerances only approximately half as great as found with conventional green sand molding.

Dimension *C* is comparable to dimension *B*, except that it is in the cope. In castings from sodium silicate molds this dimension averaged 7.984 in. When the 84 readings taken on all castings at this location were statistically analyzed, the average variation to be expected 95 per cent of the time was ±0.038 in. This is not in relationship to a previous dimension of approximately the same magnitude although in a different location.

Also, there is no correlation between this and the dimension taken in the same location on green sand castings where the average dimension was 7.964 in., with a statistically analyzed tolerance range of 0.028 in. No explanation is offered for the increased tolerance range at this location on the sodium silicate bonded sand castings.

Dimension *D* has an outside length dimension average of 12.145 in. on castings from sodium silicate bonded sand. The statistical analysis of the 84 dimensions taken at this location showed the average variation to be expected 95 per cent of the time was ± 0.026 in. Comparing the green sand castings at this location, the average dimension was 12.180 in. and the tolerance to be expected 95 per cent of the time was ±0.050 in.

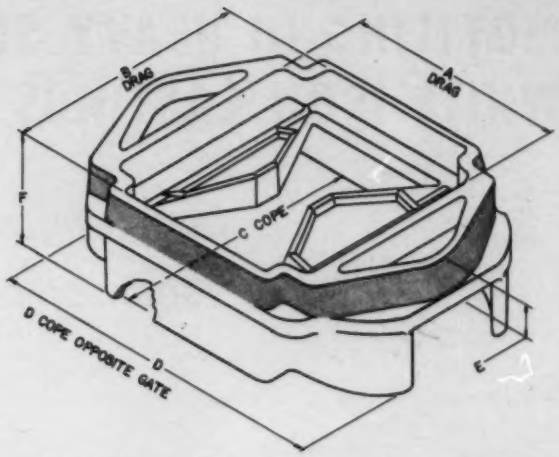


Fig. 14 — Schematic view of casting shown in Fig. 13.

As in the first two cases, the use of a sodium silicate bonded sand gave dimensional variations only half as great.

Dimension *E* is a depth dimension of the drag half of the casting from the parting line to the bottom of the mold, and had an average dimension on sodium silicate bonded sand castings of 1.362 in. When the readings from 84 castings in this area were analyzed the average variation to be expected was ±0.028 in. Comparing this with the same dimension on green sand castings, it averaged 1.367 in. and had a tolerance range of ±0.032 in.

Although not as great as in previous cases, the sodium silicate bonded mold gave smaller tolerance variations.

The last dimension, dimension *F*, although averaging 3.318 in. on sodium silicate bonded sand castings could not be properly analyzed due to a skewed distribution. The molds were all weighted uniformly and were subjected to different pouring pressures, which caused no serious horizontal mold wall deviations, but were affected vertically by the static pressure of the metal being poured. Therefore, no statistical analysis of a standard deviation could be properly established. The same situation occurred with the green sand molds where the average dimension of 84 castings was 3.318 in.

Conclusions

In the casting just analyzed, allowing for the one exception, a sand mold bonded with sodium silicate will provide greater dimensional stability, and a smaller as-cast tolerance range, than will the conventional green sand practice.

MOTTLING IN HEAVY SECTION WHITE IRON CASTINGS

first progress report on an investigation sponsored by AFS Malleable Division

By C. R. Loper, Jr. and R. W. Heine

ABSTRACT

There is an increasing demand on malleable iron foundries to produce castings of a larger section size. However, the problems of producing the casting without mottling and with adequate annealability arise.

To seek a solution to these problems a research project is being sponsored by the American Foundrymen's Society Malleable Div. at the University of Wisconsin. The project is aimed at producing heavy section castings from an iron of such composition and/or melting practice that mottling or primary graphitization will not occur during the freezing of the casting. However, the iron must have sufficient graphitizing tendency in the solid state so that it may be malleabilized by a commercially feasible heat treatment. This paper is the first progress report emanating from this project.

LITERATURE SURVEY

In general, articles published in the literature[1-17] which have a bearing on the problem state that carbon and silicon exert a dominating effect on mottling tendency in a given section size. Lowering the carbon and silicon to sufficiently small percentages will prevent mottling and cause a white fracture. Slightly higher percentages of carbon and silicon may or may not cause mottling, depending on section size and melting practices. At still higher percentages of carbon and silicon mottling will occur consistently.

This relationship between chemical composition and mottling tendency can be established by correlating the appearance of mottling with certain limiting maximum carbon and silicon contents for a given casting section size and a given melting practice. Among the criteria for combination of carbon, silicon and section size which have been suggested in the literature are:

a) $\%C + 2.85 \times \log \%Si = 2.703 - 1.06 \log D$;[2] where D = section diameter for bars in in. For a 2.0 in. diameter bar, $\%C + 2.85 \log \%Si = 2.703 - 1.6 \times 0.302 = 2.22$.

b) $\%C + 2.85 \times \log \%Si = b$;[3] where b is a constant which depends on section size. For irons produced in three foundries, a statistical study revealed that elimination of mottling in a 2.0 in. diameter section occurs when b = 2.55.

c) $4.25 \times \%C + 1.82 \times \%Si = 12.20$ or less; to eliminate mottling in a 2.0 in. diameter bar.[4]

d) $\%C + \%Si = k$; this is an arbitrary relation where a limiting value of k is assigned by a particular foundry based on their experience with a test bar. For a 2.0 in. diameter bar, constant k usually is in the range of 3.6 to 4.0.

e) $4.10 \times \log \%C + \log \%Si = 1.602$;[1] where the data from formulas (b) and (c) are also found to fit in the normal carbon and silicon percentage range in malleable iron.

In the above relationships (a) through (e), the combination of carbon and silicon percentage on the left side of the equation must be below the value expressed on the right side of the equation, in order to obtain a white cast iron regardless of melting variables. Three of these formulas are plotted in Fig. 1, along with the eutectic carbon and silicon percentage, $\%C + \frac{1}{3}\ \%Si = 4.30$, and the maximum solubility of carbon in γ iron alloyed with silicon ($\%C + \frac{1}{6} \times \%Si = 2.0$).

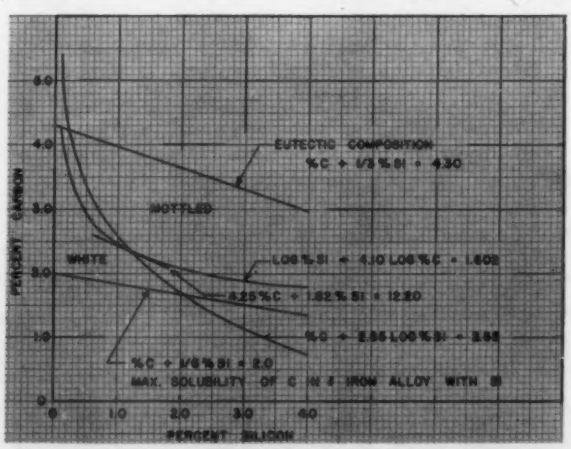


Fig. 1 — Comparison of three criteria of mottling, in a 2.0 in. diameter bar with eutectic carbon and silicon concentrations, and with maximum solubility of carbon in austenite as affected by per cent silicon.

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Theoretically, the correct equation should have a curve approaching the eutectic carbon content (4.30 per cent), or slightly above, as silicon approaches zero, since it has been observed that hypereutectic irons will always cast with structures containing primary graphite even in sections under 1/8-in. diameter quenched in water. The correct equation should also have a curve approaching or slightly below the maximum solubility of carbon in austenite as silicon content increases. This is necessary since mottling is regarded as the precipitation of graphite during freezing of the eutectic. No eutectic will be formed at compositions much below those of the solubility level of carbon in silicon alloyed iron.

A comparison of the curves (b) $\%C + 2.85 \times \log \%Si = 2.55$, (c) $4.25 \times \%C + 1.82 \times \%Si = 12.20$ and (e) $4.10 \times \log \%C + \log \%Si = 1.602$, is presented in Fig. 1. Equation (b) appears to cross the curve of maximum solubility of carbon in austenite at too low a silicon percentage to be valid. Since the data[8] from which curve (b) is taken also fits equations (c) and (e), it appears that equation (e) is the best interpretation of available data on the limits of C and Si for mottling in a 2.0 in. diameter test bar.

A better curve would be one where the equation is asymptotic both to the zero per cent silicon line near the eutectic composition and the maximum solubility of carbon in austenite curve.

The preceding discussion concerns comparisons in the literature regarding limiting carbon and silicon contents for mottling in a 2.0 in. diameter section. A base curve which is asymptotic to the zero per cent silicon ordinate, and the line indicating maximum solubility of C in γ iron ($C + 1/6 \times \%$ Si) is indicated as theoretically correct for the 2.0 in. diameter section. A similar base curve of limiting carbon and silicon contents should exist for heavier sections such as the 4 x 4 x 8 in. block casting selected as a test bar for mottling in this project.

The first phase of the project consisted of determining this base curve of carbon and silicon for producing a white fracture in a 4 x 4 x 8 in. block casting. The effects of other variables, for example bismuth, can then be related to this base curve.

EXPERIMENTAL PROCEDURE

The mottling test casting selected by the committee was a block 4 x 4 x 8 in. long made in a green sand mold, and fed by a 9 x 4¾ in. diameter riser. The drag side of the pattern is shown in Fig. 2. A thermocouple protected by a fused silica tube was inserted into the center of the 4 x 4 x 8 in. block from the top, in order to obtain cooling curves on the casting. Thirty to 35 min were required after pouring for complete solidification of the 4 x 4 x 8 in. block. Mottling data of this casting were obtained from the fracture between the riser and the 4 x 4 x 8 in. block, and from fracture at the center of the block.

The fracture was called white only if it showed no evidence whatever of mottling. It was called lightly mottled when about 6 to 12 mottle specks showed on the 4 x 4 in. section. Heavily mottled is the term used to describe a fracture about 50 per cent white and 50 per cent gray. Mottled describes a fracture which is about 75 per cent white.

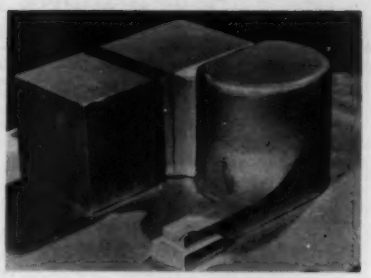


Fig. 2 — Drag view of the pattern of the 4 x 4 x 8 in. casting from which mottling data were obtained.

TABLE 1 — CHEMICAL ANALYSIS AND SURFACE AREA OF CHARGE MATERIALS

Material	Composition, %					Average Surface Area, in.²/lb
	C	Si	Mn	S	P	
Pig Iron	4.20	1.62	0.80	0.043	0.13	8.0
Sprue	2.36	1.18	0.31	0.090	—	17.0
Steel	0.24	0.09	0.66	0.04	—	22.4

All heats were melted in a high frequency induction furnace under a controlled atmosphere. With the exception of a few high carbon, low silicon heats, this atmosphere consisted of 21 per cent carbon dioxide and 79 per cent nitrogen. The CO_2 - N_2 atmosphere was selected since it represents the most oxidizing melting condition without free oxygen, or the case where perfect combustion of carbon takes place.

High carbon, low silicon heats 22, 25, 30 and 31 were melted in a 35 per cent carbon monoxide, 65 per cent nitrogen atmosphere because of their tendency to boil when melted under the 21 per cent CO_2-79 per cent N_2 atmosphere. The 35 per cent carbon monoxide atmosphere represents the case where a solid fuel is completely burned to CO with no excess O_2.

Table 1 presents the chemical analysis and approximate surface area of the charge ingredients. With a few exceptions, 50 per cent sprue was used in all charges. The chemical composition was adjusted through proper proportioning of pig iron, sprue, steel and 75 per cent silicon ferro-silicon or graphite obtained from crushed electrodes. The pig iron was sawed into pieces 4 to 8 in. long for convenience in charging. The steel consisted of scrap nuts and bolts. All charge ingredients were thoroughly cleaned of adhering dirt and scale.

Charging Procedure

Charging procedure was as follows:

1) The pig iron was placed in the bottom of the furnace.
2) Ferro-silicon, if required, was placed on top of the pig.
3) Sprue was then placed on top of the pig.
4) As melting progressed steel was added.
5) Any graphite required, was added to the charge after the melt reached a temperature of 2570-2800 F, and was cleaned of slag. At this temperature the graphite entered readily into solution.

Melt down time for all heats was about 65-70 min. The bath was brought up to 2800 F in a total of 80-

TABLE 2 — CHEMICAL ANALYSIS AND FRACTURE RESULTS OF HEAT POURED — NO ADDITION MADE

Heat No.	Chemical Analysis, %[1] C	Si	Fracture Results[2] 4x4x8 in. casting	1⅞ in. D. bar
1	2.50*	1.09*	G	M
2	2.50*	1.32*	G	M
8	2.39*	1.40*	G	M
11	2.44*	1.05*	M	W
12	2.34*	1.05*	M	LM
13	2.20*	1.41*	HM	M
14	2.12*	1.21*	M	W
15	1.85	0.86	W	W
16	2.34*	0.95*	M	W
17	2.10*	0.82*	LM	W
18	2.10	0.87	M	W
19	1.90*	1.12*	LM	W
20	2.00	1.15	M	W
21	1.85*	1.41*	LM	W
22	3.00	0.40	M	W
24	1.50	4.00	G	HM
25	3.03*	0.25*	LM	W
26	1.13	4.00	W	W
27	1.32*	3.54*	W	W
28	1.63	2.00	HM	—
29	1.46*	3.94*	HM	HM
30	2.60*	0.25*	W	—
31	2.80	0.31	W	W
54	2.11	1.65	G	LM
55	2.11	1.65	G	LM

1. Estimated chemical analyses, unless marked with an asterisk for which actual chemical analyses were obtained.
2. G = gray fracture; HM = heavily mottled fracture, almost gray; M = mottled fracture; LM = lightly mottled fracture; W = 100 per cent white fracture.

90 min from the start of the heat. The heat was then held at 2800 F for 60 min. The furnace was turned off and the melt allowed to cool. In about 3 min the bath dropped to a temperature of 2750 F, and in another 2 min the bath temperature dropped to 2700 F.

Those heats to which no additions were made were poured at 2700 F. In the case of heats having a bismuth addition, the bismuth was added at 2750 F, and the heat poured at 2700 F. The bismuth added was wrapped in a sheet of mild steel, wired to a fused silica rod and subsequently immersed in the bath. As a result of the melting practice followed, no oxidation losses of silicon occurred from charge to final analysis. Total decarbonization amounted to about 0.30 per cent for each heat.

The final analysis is listed in Table 2. Table 2 lists actual chemical compositions for 15 heats and estimated compositions for 10 heats. Only the significant heats were analyzed. However, past experience has shown that estimated analyses are always correct within a few points of carbon and silicon.

RESULTS

The effect of carbon and silicon percentages on the fracture of the 4 x 4 x 8 in. casting is listed on Table 2, and shown graphically in Fig. 3. Figure 3 also shows the line representing the maximum solubility of carbon in austenite. The curve representing the maximum carbon and silicon contents which will produce a white fracture in the 4 x 4 x 8 in. casting is seen to be asymptotic to the 0 per cent silicon line and the curve of maximum solubility of carbon in austenite.

Increasing the carbon or silicon percentage from this base line in Fig. 3, in the region less than 2.50 per cent silicon, will result in a slightly mottled fracture. Further increase will result in heavier mottle and finally a fully gray fracture. The base curve for carbon and silicon limits for gray fractures is also shown in Fig. 3. Carbon and silicon contents in excess of those on this base curve will produce a gray fracture in the 4 x 4 x 8 in. casting.

Silicon contents of greater than 2.50 per cent produce an interesting result. From Fig. 3 it may be noticed that above 2.50 per cent silicon the base curves for white and gray fractures appear to be coincident. This means that a small increase in carbon content changes the fracture from fully white to completely gray.

BISMUTH EFFECT

In order to produce heavy section malleable castings the base curve on Fig. 3 for white fractures must be raised to higher silicon and carbon contents. This is necessary in order to obtain sufficient graphitizing tendency on annealing. Ladle additions of bismuth have been used for this purpose in malleable foundries, as reported in references.[5,15] The effect of bismuth additions to the metal may be seen in Fig. 4, the data for Fig. 4 being presented in Table 3.

This curve represents the shift in base line brought

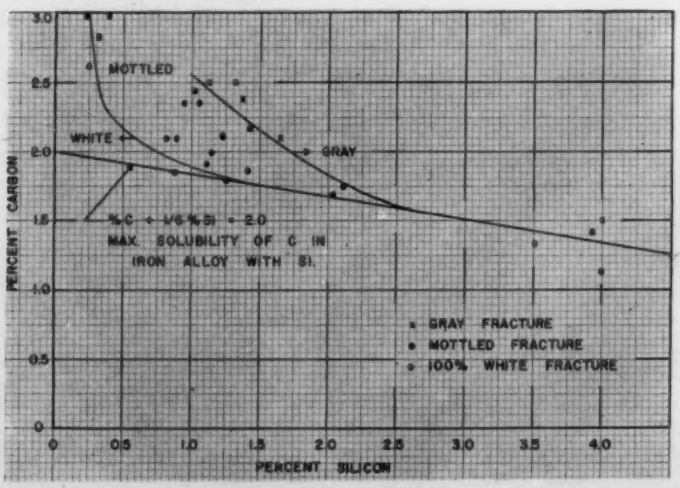


Fig. 3 — Carbon and silicon content effect on the fracture obtained in the 4 x 4 x 8 in. casting showing the maximum carbon and silicon percentages which will yield a fully white fracture, and the minimum carbon and silicon percentages which will yield a completely gray fracture.

about by the addition of 0.01 per cent bismuth, according to the procedure described earlier. The effect of bismuth is seen to be much more pronounced at lower silicon contents than at higher silicon contents, according to Fig. 4. In fact, it appears that beyond about 2.40 per cent silicon, little or no effect on the fracture will be obtained from 0.01 per cent bismuth addition.

Fracture results on casting of an analysis just to the right of the curve for 0.01 per cent added bismuth appear to be identical to those just to the right of the base line curve when no bismuth is added. The effect of 0.01 per cent bismuth in the area of the shifted base line curve, presented in Fig. 4, is to change the fracture from very heavily mottled to fully white, or from completely gray to very lightly mottled. This may be noted by comparison of Figs. 3 and 4.

The effect of bismuth additions larger than 0.01 per cent is shown in Fig. 5. In the compositions studied, bismuth in amounts of 0.02 to 0.10 per cent produced no greater effect than that obtained from 0.01 per cent bismuth. Fracture results from these increased percentages of bismuth were similar to those of 0.01 per cent bismuth additions at corresponding carbon and silicon contents.

TEMPERATURE OF BISMUTH ADDITION EFFECT

The effect of adding bismuth at 2600 F rather than 2750 F was studied with results reported in Table 4. Bismuth additions of 0.01 per cent at 2600 F resulted in gray fracture in heats 67, 68 and 70, whereas a lightly mottled fracture would be expected in heat 67 and a heavily mottled fracture would be expected in heats 68 and 70. Heat 67 had an analysis of 2.30 per cent C and 1.65 per cent Si, and would normally show a lightly mottled fracture from the addition of 0.01 per cent bismuth at 2750 F (Fig. 3).

However, it had a gray fracture when the bismuth was added at 2600 F. Heats 68 and 70 had an analysis of 2.58 per cent C and 1.65 per cent Si, and would normally show a heavily mottled fracture from the addition of 0.01 per cent Bi at 2750 F. However, it had a gray fracture when the bismuth was added at 2600 F. Thus, it is seen that the temperature of bismuth addition is exceedingly important in obtaining its full affect on raising carbon and silicon limits for mottling.

WATER VAPOR EFFECT

It has been previously reported that water vapor in the furnace atmosphere raised the base curve for a white fracture in a 2.0 in. diameter bar.[2] The effect of water vapor on the base line for white fractures in the 4 x 4 x 8 in. test casting is presented in the data in Table 4. Water vapor in the amount of 12 grains cu ft of gas in a $CO_2 \cdot N_2$ atmosphere, heat No. 43, produced a lightly mottled fracture in an analysis of 2.10 per cent C, 1.63 per cent Si identical with that to be expected from 0.01 per cent Bi addition, as shown in Fig. 4.

Heat No. 47, containing 2.14 per cent C and 1.40 per cent Si, to which 25 grains water vapor/cu ft of

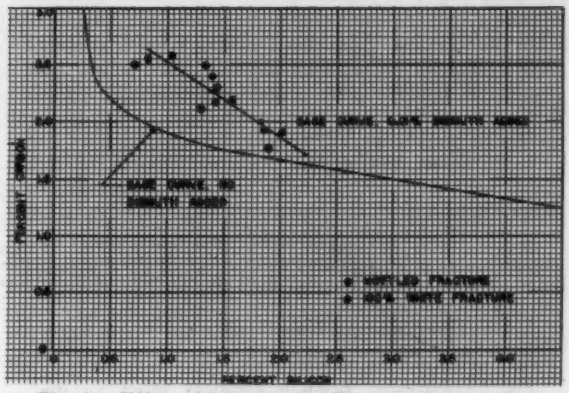


Fig. 4 — Effect of 0.01 per cent Bismuth added on the base curve of maximum carbon and silicon percentages which will yield a fully white fracture in the 4 x 4 x 8 in. casting.

TABLE 3 — CHEMICAL ANALYSIS AND FRACTURE RESULTS OF HEAT POURED, 0.01 PER CENT BISMUTH ADDED

Heat No.	Chemical Analysis, %[1] C	Si	Fracture Results[2] 4x4x8 in. casting	1⅞ in. D. bar
3	2.50*	1.33*	M	W
6	2.42*	1.39*	M	W
9	2.34*	1.42*	M	M
33	1.77*	2.00*	M	M
34	2.18*	1.41*	W	W
35	2.68*	1.03*	M	W
36	1.77*	1.80*	W	W
37	2.11	1.30	W	W
38	2.50	0.70	W	W
39	1.77	1.90	W	W
40	2.18*	1.57*	LM	W
41	2.50*	0.85*	W	W

1. Estimated chemical analyses unless marked with an asterisk for which actual chemical analyses were obtained.
2. M = mottled fracture; LM = lightly mottled fracture; W = 100 per cent white fracture.

TABLE 4 — CHEMICAL ANALYSIS AND FRACTURES, VARIOUS ADDITIONS

Heat No.	Chemical Analysis, %[1] C	Si	Addition	Fracture Results[2] 4x4x8 in. casting	1⅞ in D. bar
4	2.40*	1.30*	0.02 Bi	M	W
5	2.36*	1.28*	0.05 Bi	M	W
7	2.40*	1.49*	0.10 Bi	M	—
10	2.41*	1.41*	0.10 Bi	M	W
42	2.11	1.65	0.05 Bi	LM	W
43	2.10*	1.63*	12 gm H_2O/cu ft gas	LM	W
46	2.19*	1.64*	12 gm H_2O/cu ft gas plus 0.01% Bi	LM	LM
47	2.14*	1.40*	25 gm H_2O/cu ft gas	W	W
52	2.11	1.65	0.05 Bi	LM	W
60	2.11	1.65	0.01 Bi_3	W	—
67	2.30	1.65	0.01 Bi_3	G	G
68	2.58	1.65	0.05 Bi	G	G
69	2.58	1.65	0.05 Bi_3	HM	—
70	2.58	1.65	0.05 Bi	G	—

1. Estimated chemical analyses unless marked with an asterisk for which actual chemical analyses were obtained.
2. G = gray fracture; HM = heavily mottled fracture; almost gray; M = mottled fracture; LM = lightly mottled fracture; W = 100 per cent white fracture.
3. Bismuth additions made and castings poured at 2600 F compared to normal procedure of adding bismuth and pouring at 2750 F.

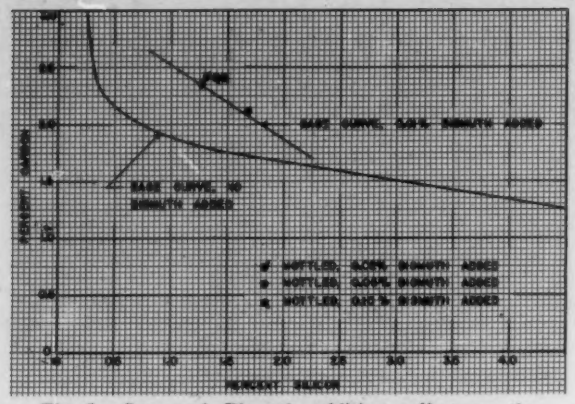


Fig. 5 — Increased Bismuth additions effect on the base curve of maximum carbon and silicon percentages with 0.01 per cent Bismuth added, which will yield a fully white fracture in the 4 x 8 x 8 in. casting.

gas in a CO_2 - N_2 atmosphere were added, produced a white fracture identical with that to be expected from 0.01 per cent Bi additions, shown in Fig. 4. Addition of 12 grains water vapor/cu ft gas in a CO_2 - N_2 atmosphere and 0.01 per cent bismuth to heat No. 46 of 2.19 per cent C, 1.64 per cent Si, produced a lightly mottled fracture identical with that obtained from water vapor alone or 0.01 per cent bismuth alone.

Therefore, it would appear, that similar results can be obtained from a controlled amount of water vapor in the furnace atmosphere, as can be obtained from the use of bismuth additions in so far as mottling is concerned. Water vapor, however, has been proved to interfere with second stage graphitization.[1]

SUMMARY

Base line curves representing the maximum carbon and silicon content which will produce a white fracture in the 4 x 4 x 8 in. casting, and the minimum carbon and silicon content which will produce a gray fracture in the 4 x 4 x 8 in. casting have been presented in Fig. 3.

The effect of bismuth additions on the limiting carbon and silicon contents for white fracture in the 4 x 4 x 8 in. test casting has been shown in Fig. 4. Percentages of bismuth additions over 0.01 per cent have been shown to be of no value in reducing mottling in the 4 x 4 x 8 in. test casting insofar as obtaining a white fracture is concerned. Too low a temperature of addition (2600 F) reduces the effectiveness of the bismuth additions. Water vapor has been shown to be as effective as bismuth in raising carbon and silicon limits for mottling, according to the melting practices followed in this research.

Future studies on the casting of heavy section malleable will be aimed at ladle additions and/or melting practices that will raise the carbon and silicon limits for mottling above the base line curves presented in this report.

This paper is the first annual progress report of this project. As such, it should be recognized that the various graphs and conclusions stated, represent the best analyses of the data obtained thus far and they may be modified if future work so indicates.

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REFERENCES

1. R. W. Heine, *Melting Variable Effects on Malleable Iron Properties,* American Foundrymen's Society, pp. 13-15, 1959.
2. R. Schneidewind and R. G. McElwee, "Composition and Properties of Gray Iron," AFS TRANSACTIONS, *vol. 58,* p. 312, 1950.
3. H. A. Schwartz, W. K. Bock, and J. D. Hedberg, "Graphitization in the Malleable Iron Industry," AFS TRANSACTIONS, *vol. 61,* p. 228, 1953.
4. E. H. Belter and R. W. Heine, "Effects of Charge Materials and Melting Conditions on Properties of Malleable Iron," AFS TRANSACTIONS, *vol. 65,* p. 134, 1957.
5. A. L. Boegehold, U.S. Patent No. 2,370,225, Feb. 27, 1945.
6. L. Crome, "Some Notes on Graphitization," *Foundry,* p. 98, April 1946.
7. H. A. Schwartz, "Solved and Unsolved Problems in the Metallurgy of Blackheart Malleable," *Foundry,* p. 74, July 1948.
8. H. Marrogh and W. J. Williams, "Graphite Formation in Cast Irons and in Ni-C and Co-C Alloys," *Journal of Iron and Steel Inst., vol. 155,* p. 321, 1947.
9. H. A. Schwartz, "Chemical Composition of Malleable Iron," AFS TRANSACTIONS, *vol. 54,* p. 101, 1946.
10. B. F. Brown and M. F. Hawkes, "Kinetics of Graphitization in Cast Iron," AFS TRANSACTIONS, *vol. 59,* p. 181, 1951.
11. E. A. Lange and R. W. Heine, "Some Effects of Temperature and Melting Variables on Chemical Composition and Structure of Gray Irons," AFS TRANSACTIONS, *vol. 59,* p. 472, 1951.
12. E. A. Lange and R. W. Heine, "Effect of Melting Furnace Atmosphere on Malleable Iron," AFS TRANSACTIONS, *vol. 61,* p. 107, 1953.
13. A. Berman and V. Konelie, "Decomposition of Cementite During Solidification of Cast Iron," *Journal of Iron and Steel Inst.,* p. 385, April 1954.
14. H. A. Schwartz and J. D. Hedberg, "Graphitization of Rims in Malleable Iron," AFS TRANSACTIONS, *vol. 62,* p. 282, 1954.
15. A. L. Boegehold, "A New Composition for Fast Malleable," *Metal Progress,* p. 73, Oct. 1956.
16. A. H. Zrimsek, E. H. Belter and R. W. Heine, "Effects of Charge Proportions, Furnace Atmosphere Flow Rate, and Melt Down Time on Properties of Malleable Iron," AFS TRANSACTIONS, *vol. 64,* p. 72, 1956.
17. J. E. Hanes and O. Kondic, "Origin of Graphite in Cast Irons," *Foundry Trade Journal,* p. 267, Feb. 28, 1957, p. 311, March 7, 1957.

PRESSURE TIGHTNESS OF 85-5-5-5 BRONZE

relationship of thermal gradients during solidification

PROGRESS REPORT OF AFS BRASS and BRONZE DIVISION RESEARCH COMMITTEE

By Richard A. Flinn and Chester R. Mielke

ABSTRACT

Previous work established the marked influence of different types of chilling as well as risering upon the pressure tightness of 85-5-5-5. The purpose of this investigation was to measure the thermal gradients during solidification to provide a quantitative evaluation of different risering and chilling combinations.

The data indicate that pressure tight castings were obtained in 2 in. square bar sections when the thermal gradient exceeded 60 F/in. By the use of tapered chills it was possible to attain this gradient in bars up to 9 in. in length.

An end chill produces gradients over 60 F/in. for 4 in. along the bar, while the uninsulated risers develop gradients of this magnitude for less than 1 in. The combined effect of riser and tapered sidewall chills results in a gradient of 60 F/in. or more for 3 in. from the riser.

Preliminary experiments involving the effects of riser size and bar thickness are discussed.

INTRODUCTION

The purpose of this project, which has been supervised by the Brass and Bronze Division of the AFS,* has been to determine the principal variables responsible for control of pressure tightness in 85-5-5-5 bronze. This alloy was selected because it is the most widely used copper base alloy. However, the procedures have deliberately been developed so that any other copper base alloys can be readily evaluated in similar fashion.

Previous reports[1,2] indicate that pouring temperature and chill geometry are important in controlling pressure tightness.

The objectives of the present work were to determine quantitatively the thermal gradients produced by the riser and the chills as the following features of mold design were changed:

1) Casting length.
2) Riser size.
3) Casting thickness.

In this way the relative effectiveness of risers of different geometry and chills of different design can be evaluated for practical application to casting problems.

PROCEDURE

The details of procedure have been described in previous reports.[1,2] Briefly, a constant metal history was maintained by using charges composed entirely of ingot from two interrelated heats, donated by the Brass and Bronze Ingot Institute. Melting was conducted in a gas-fired crucible furnace under oxidizing conditions. Zinc and phosphorus copper additions were made as described previously. Temperatures were measured at tap and pouring with an immersion chromel-alumel thermocouple and calibrated high-speed recorder.

In addition to the standard 12 x 2 x 2 in. bar with 4 in. diameter riser, variations in riser size and in bar thickness were planned as indicated on the appropriate data sheets.

Pressure tightness was evaluated, as discussed previously, in a 2 x 12 x $\frac{1}{32}$-in. specimen cut horizontally through the center of the test casting. Of course, in the special 6 in. and 9 in. long bars this test section was only as long as the bar. When thermocouples were employed at the center line, the section for pressure testing was displaced less than $\frac{1}{4}$-in. to avoid the thin silica protection tubes. Previous work indicated that in a 2 x 2 in. square section this displacement will not affect the pressure test results appreciably.

To determine thermal gradients thermocouples were inserted in the mold, as illustrated in Fig. 1. Thermocouple locations are indicated on the graphs showing thermal gradients.

Thermocouple Technique

Since considerable development work was necessary to obtain really sensitive thermocouples, a brief description of the final technique may be of interest. To minimize thermal lag, 28 gage (0.0126 in.) chromel and alumel wires were butt welded using a condenser discharge arc. The thermocouple was then slipped into a fused silica protection tube which had been drawn from 0.075 in. O.D. and 0.039 in. I.D.

*Membership of the Brass and Bronze Research Committee: F. L. Riddell, *Chairman;* E. F. Tibbetts; R. A. Colton; G. LeBrasse; R. W. Ruddle; C. W. Ward; H. Bishop; S. C. Massari, *AFS Technical Director;* R. B. Fischer; and R. J. Keeley.

R. A. FLINN is Prof. and C. R. MIELKE is special student, Dept. of Chemical and Met. Engrg., University of Michigan, Ann Arbor.

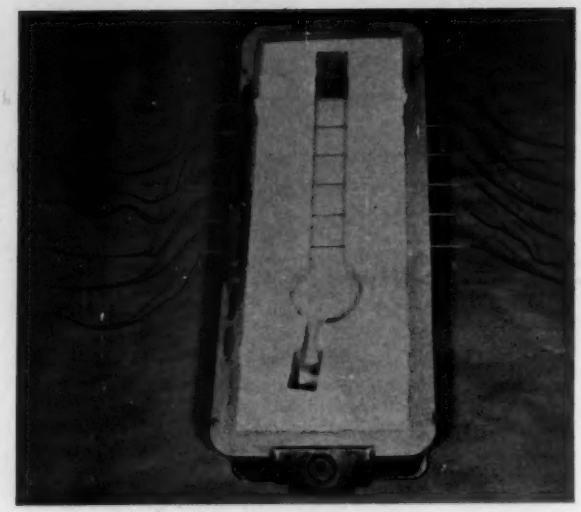


Fig. 1 — Drag half of mold showing location of thermocouples.

to an O.D. of approximately 0.045 in. and 0.020 in. I.D.

With this technique thermocouple assemblies of small thermal capacity and good sensitivity were obtained. These thermocouples reached the metal temperature by the time the mold was filled (12 sec).

To permit the insertion of these delicate tubes in the mold, pattern and flask alterations were made, as illustrated in Fig. 2. Therefore, oversize holes were present in the mold and flask. After locating the thermocouples with the butt weld at the centerline of the bar, the tube position was fixed with a small amount of core paste. An assembly ready to pour is illustrated in Fig. 3. Examination of the castings showed no shifting of the thermocouples.

The thermocouples were connected to a rotating silver commutator leading to a high-speed recorder. In this way each couple gave a reading every 6 sec. A typical recorder trace is shown in Fig. 4.

THERMAL GRADIENTS CALCULATIONS

From the cooling curve data of Fig. 5 it is possible to compute thermal gradients in several ways. The tangent method (Method 1) is to determine the temperature at different stations at the different intervals after pouring, as illustrated in Fig. 6. Then, from these curves the thermal gradient at any desired station and temperature can be determined by drawing a tangent and taking the slope. A sample calculation is shown on the graph. While this meets the proper definition of a gradient dT:dx, it has several disadvantages.

First, if the gradient is taken at the solidus the colder, solid material influences the numerical value of the gradient through its effect upon the curve. However only the hotter, still liquid material is of importance in feeding. Second, the tangent must be drawn to a curve derived from a second interpolation of the measured data, and is therefore not as accurate as if the raw data could be used.

To meet these objections an alternate and much simpler method may be used (Method 2). If the gradient at a given station is desired at the solidus, for example, the temperature difference to the nearest (hotter) station is determined from the original cooling curves and divided by the distance. This calculation really gives a secant rather than a true tangent, but has the advantage of simple direct computation. It also measures only the upstream gradient to the source of liquid metal. It is true that if the distance to the upstream thermocouple varies from 1 in. to 2 in., the variation between the secant and the tangent methods will vary.

Computations by both methods are illustrated in Figs. 5 and 6.

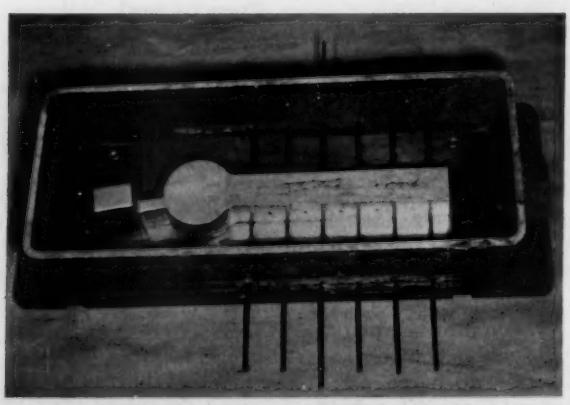


Fig. 2 — Pattern and flask modifications for thermocouple insertion.



Fig. 3 — Mold assembly and recording equipment.

EFFECT OF TEMPERATURE SELECTED FOR THERMAL GRADIENT MEASUREMENT

The liquidus temperature will vary in the range 1824 F - 1852 F depending upon cooling rate as indicated in the graphs. Because of the long freezing range and high conductivity the solidus is not indicated by the graphs. A constant reference solidus of 1568 F[8] was used. The choice of absolute temperature for the solidus is not critical because of the slow change in cooling rate in this temperature range.

In addition to calculating the thermal gradient at the solidus, determinations at other temperatures between liquidus and solidus were also made. In addition to the general interest of these data, these calcu-

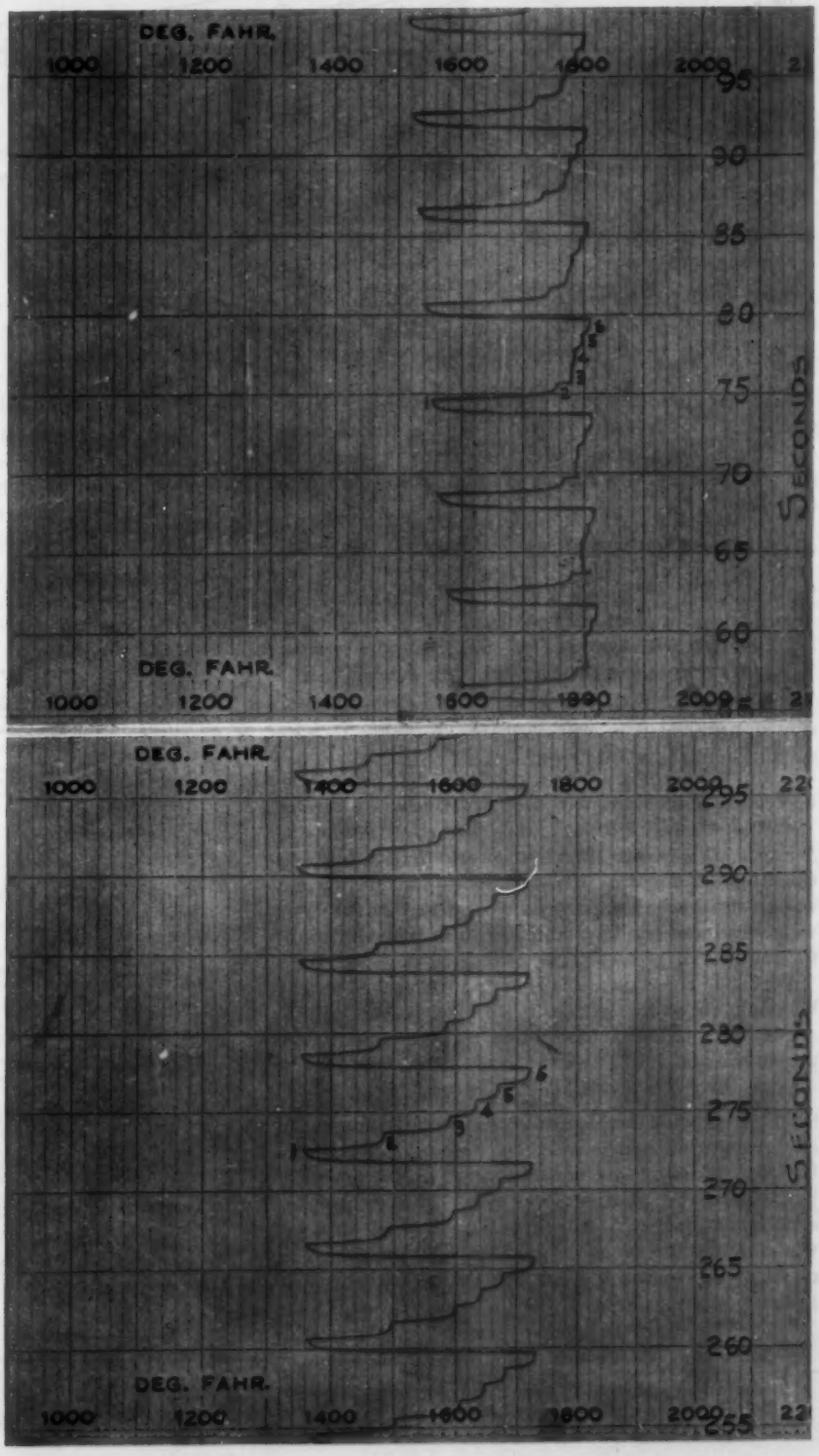


Fig. 4 — Typical recorder trace showing a small gradient 50 sec after pouring (top), and a larger gradient at 250 sec after pouring (bottom).

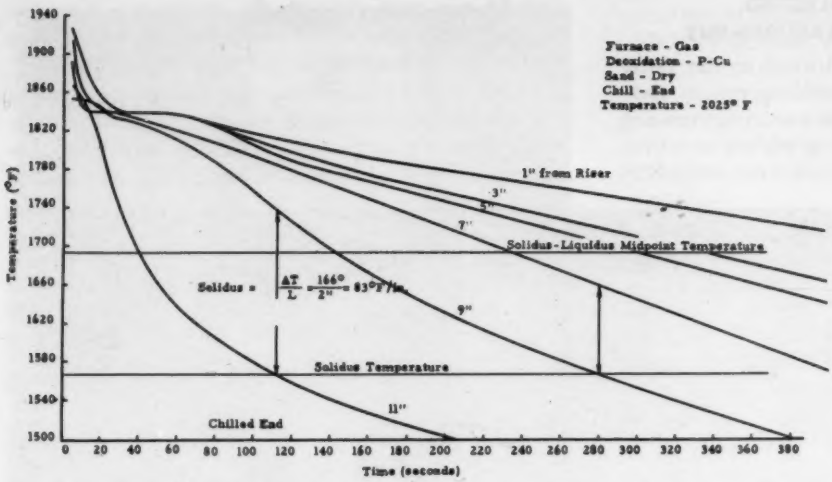


Fig. 5 — Typical cooling curves, secant method, 2 in. bar.

lations were performed because of the complex nature of the 85-5-5-5 bronze system. The last portions to solidify are richer in lead and may not require extensive feeding.

Also, a satisfactory gradient may be necessary in the middle of the freezing range to reduce the distance of liquid metal transport through the dendritic growth. In general a higher gradient is developed at the solidus.

DISCUSSION OF RESULTS

The results may be considered in three sections:

1) 6, 9 and 12 in. bars with wedge chills.
2) 12 in. bars with varying riser size.
3) 12 in. bars of different thickness.

In each case the effect of mold design upon pressure tightness will be discussed first, followed by consideration of the thermal gradients.

6, 9 and 12 in. Bars with Wedge Chills (poured at 2025 F)

Pressure Tightness. Previous work had indicated that the 12 x 2 x 2 in. bar was not sound even when wedge chills were employed. Therefore, shorter bars were produced to determine the limiting conditions for soundness.

In Fig. 7 the data show that 6 and 9 in. bars may be made pressure tight using the wedge chill technique. The 12 in. bar exhibited leakage, as noted in previous work.

Thermal Gradients. The thermal gradient data, Figs. 8 and 9, provide helpful quantitative information in explaining the leakage rates. In the sound 6 and 9 in. bars the minimum gradient was 60 F/in. In the 12 in. bar where the gradient was over 60 F/in. solidus sound material was also encountered, while leakage was noted where a lesser gradient was present. The effect of shortening the bar was to increase the minimum gradient as follows:

Bar size, in.	Approx. minimum gradient at solidus, F/in.
12	36
9	60
6	150

The effects of the riser and the end chill are also of interest. In the 12 in. bars, the gradient at 3 in. from the riser is 60 F/in., while 3 in. from the end chill the gradient is also about 60 F/in. The gradient produced by the end chill is less than if the walls were of sand. This effect is due to the fact that the sidewalls are almost completely graphite, and therefore the gradient obtained is due mostly to end effects, that is, the presence of the end surface.

In the case where only an end chill is present (sand

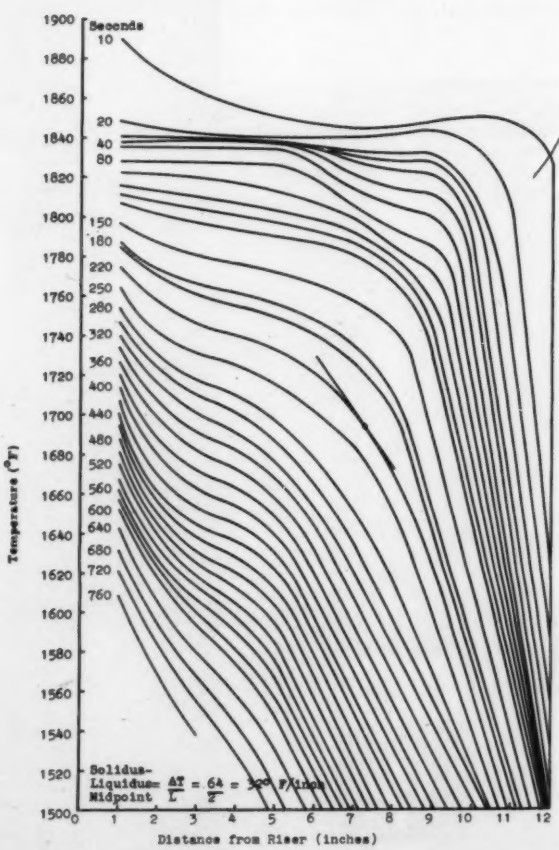


Fig. 6 — Typical iso-time cooling curves, tangent method, 2 in. bar.

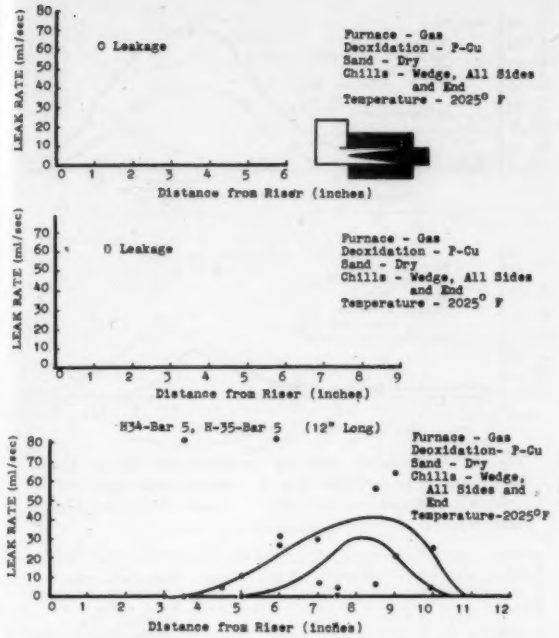


Fig. 7 — Effect of length of bar upon leak rate (wedge chills), for different bar lengths. Top — 6 in. long, center — 9 in. long, bottom — 12 in. long.

sidewalls) the gradient is due to the combined effects of the end surface and the difference in sidewall material. However, by the use of tapered graphite chills it is possible to ration the end wall effect to produce a satisfactory gradient further into the mold. This point is confirmed by the data of the next section.

Variations in Pouring Temperature Effect

Preliminary data indicating the effects of variations

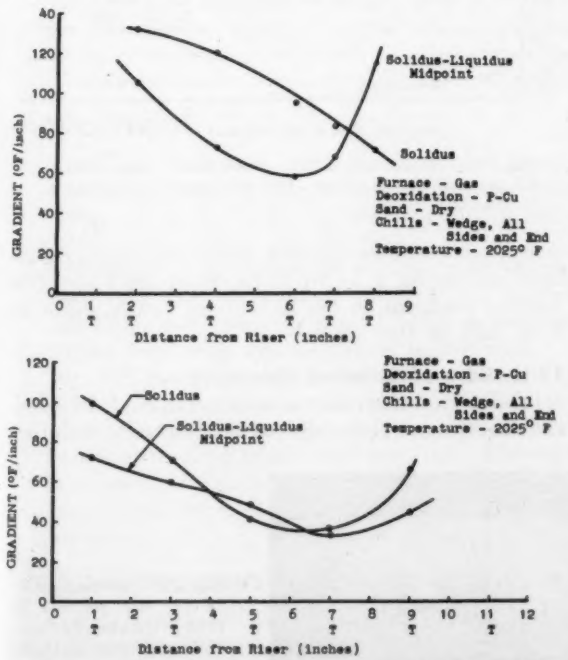


Fig. 9 — Thermal gradients in 9 in. (top) and 12 in. (bottom) bars.

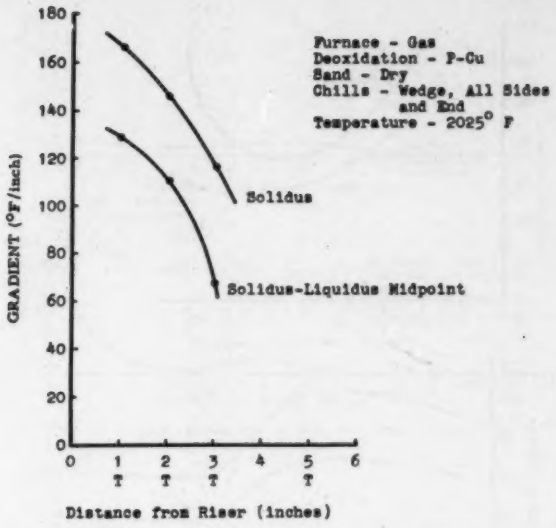


Fig. 8 — Thermal gradients in 6 in. bar.

in pouring temperature are given in Figs. 10, 11, 12 and 13.

Two bars, 9 in. and 12 in. long, respectively, were poured at different and higher temperatures (Figs. 10 and 11). The 12 in. bar poured at 2275 F exhibited severe leakage, probably due to the greater gas content of the metal in the bar. The greater reaction rate, and longer reaction time between metal and mold, probably leads to the higher gas content. These data indicate that the effects of a good mold design may be nullified if excessive pouring temperatures are employed.

It is interesting that somewhat greater thermal gradients were obtained at the higher pouring temperatures. The greater solution, and later evolution of gas, was apparently more important than the thermal gradient in the 12 in. bar. A photograph of the midsection of this bar is shown in Fig. 12. The 9 in. bar poured only 150 F above the standard temperature of 2025 F was still sound.

Additional experiments were performed to determine the effect of reducing pouring temperature below the standard of 2025 F. Previous work[1,2] indi-

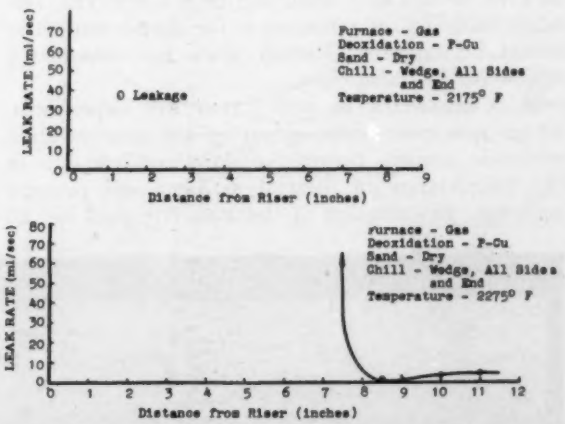


Fig. 10 — Effect of increased pouring temperature upon leak rate for 9 in. long bar (top) and 12 in. long bar (bottom).

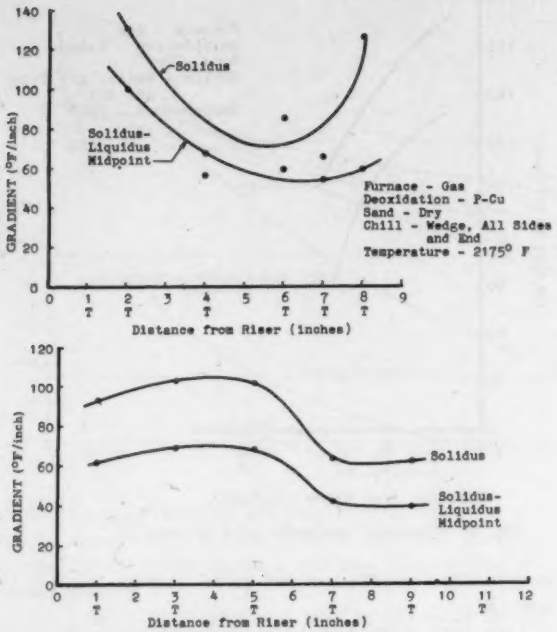


Fig. 11 — Thermal gradients of 12 in. (bottom) and 9 in. (top) bars poured at higher than standard temperature.

cated that this was helpful for bars cast with sand sidewalls and end chills. A small reduction in leakage was encountered, as shown in Fig. 13.

12 in. Bars With Varying Riser Size

The results of preliminary experiments in this field are given in Figs. 14 and 15.

A variation from −44 per cent to +37 per cent in riser size, referred to the standard riser, had the following effects. The larger riser provided a sounder section at the risered end of the bar than the smaller riser, but no improvement over the standard riser. The leakage is similar in all three cases at stations further than 3 in. from the riser.

The thermal gradient data provide an interesting quantitative verification of these effects. The (solidus) gradient at 7 in. from the riser, for example, is 44 F/in. in bars with small and large risers. This certainly indicates the influence of the riser is limited as would be expected. Further work is contemplated employing insulated risers.

It is important to notice that this experiment, which was quite independent of the first (dealing with bar length), provides striking confirmation of the interrelation of thermal gradient and pressure tightness. Examination of the leak rate data for all

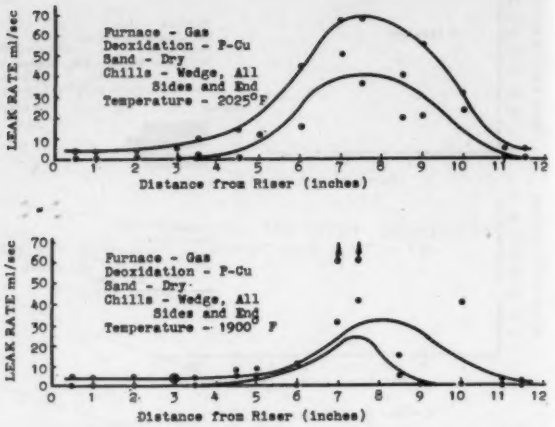


Fig. 13 — Reduced pouring temperature effect upon leak rate. Top — H31, bar 1 (black dot) and bar 2 (circle). Bottom — H31, bar 3 (black dot) and bar 4 (circle).

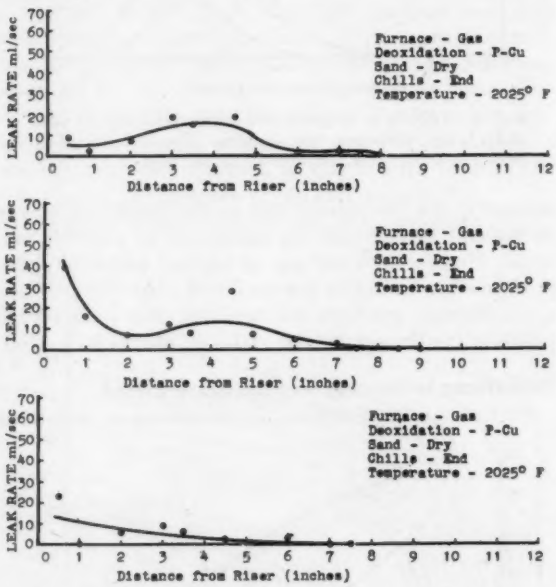


Fig. 14 — Riser size effect upon leak rate. Top — 37 per cent larger; center — 44 per cent smaller; bottom — standard size.

three bars indicates that the sound material extends from the region 8 in. to 12 in. from the riser. The thermal gradients in the bars attain values above 60 F/in. only in this range.

12 in. Bars of Different Thickness

Preliminary experiments were conducted to evaluate the effect of reducing the bar thickness to values

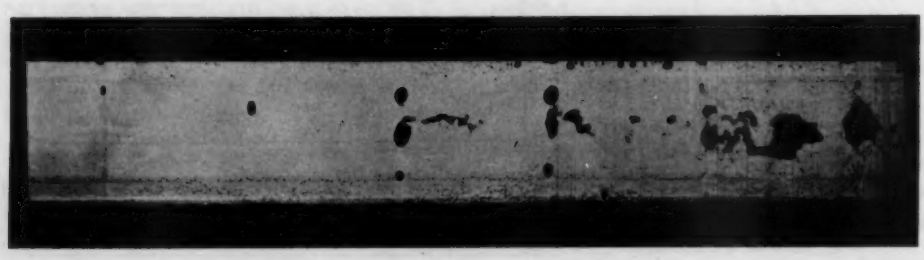


Fig. 12 — Midsection of 12 in. bar poured at 2275 F showing gas porosity, riser end on right.

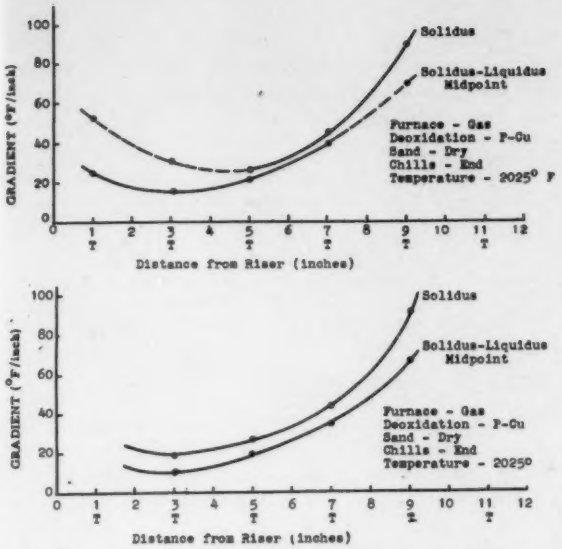


Fig. 15 — Thermal gradients for large and small risers. Top — 37 per cent larger; bottom — 44 per cent smaller.

of 1½, 1, ½ and ¼-in. The ¼-in. bar did not run satisfactorily, and the end of the ½-in. bar did not fill. Since this was a preliminary survey both thermocouple and leak rate data were obtained. As mentioned previously, this required a shift of about ¼-in. away from the centerline for pressure testing, and only the 2 in. and 1½-in. data are therefore considered representative (Figs. 16-18).

The 1½-in. bar showed a slightly lower leak rate than the 2-in. bar, particularly near the riser. This may be due to two reasons, 1) steeper gradient at the riser, and 2) the greater ease of plate like feeding compared to bar-like feeding (in plates, dendrites originate at only two walls compared to four in bars). These bars exhibit leak free regions when a thermal gradient of 60 F/in. is exceeded, in accord with the observations previously described for the other tests.

CONCLUSIONS

1) Pressure tight bars, 2 in. x 2 in. x 9 in. and shorter can be made by proper chilling techniques. This range can probably be extended by modification of the design of the tapered chills.
2) Preliminary experiments in varying riser volume indicated that the thermal gradient was affected for about 4-in. from the riser. Further experimentation employing insulated risers is indicated.
3) In the many different specimens which were tested, sound material was obtained when the thermal gradient exceeded 60 F/in., and unsound material resulted from lower gradients. The only exception to this rule occurred when high pouring temperatures were used, resulting in gassy castings.

RECOMMENDATIONS FOR FUTURE WORK

The data indicate that a thermal gradient of 60 F/in. is required for soundness in 2-in. square bar sections in a variety of mold modifications. The effectiveness of an end chill in establishing this gradi-

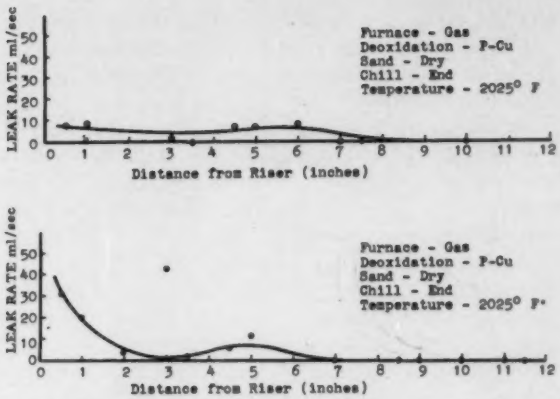


Fig. 16 — Bar thickness effect on leak rate. Top — 1½ in. bar; bottom — 2 in. bar.

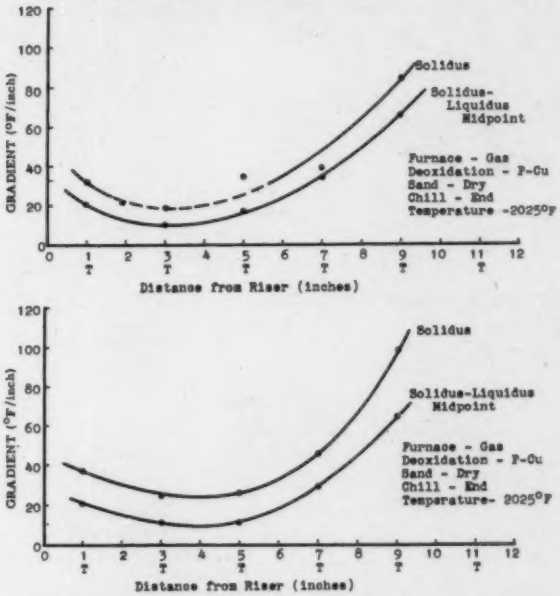


Fig. 17 — Thermal gradients in 2 in. (top) and 1½ in. (bottom) thick bars.

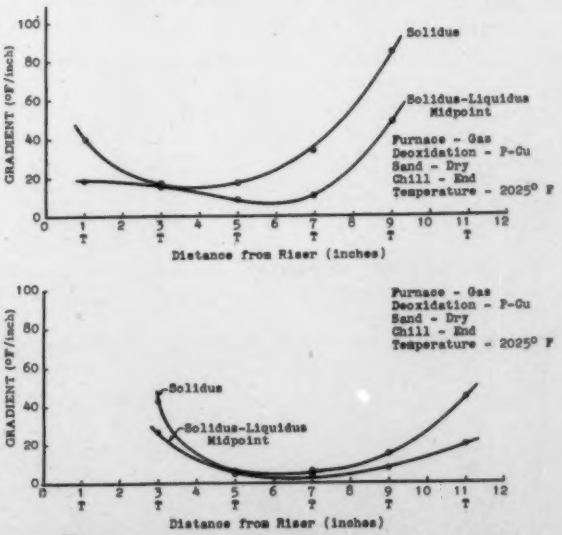


Fig. 18 — Thermal gradients in 1 in. (top) and ½-in. (bottom) thick bars.

ent is for about 4 in. along the bar. The risers employed thus far have provided this gradient for less than 1 in., and therefore unsound material is encountered near the riser.

Further work is indicated using insulated risers. A combination of risers plus wedge chills provides a gradient of 60 F/in. for slightly over three in. The combination of insulated risers with wedge chills should be of importance.

The determination of distances of riser and chill effectiveness for other sections will complete the data needed for mold design. Other bar cross-sections and several plate cross-sections should be investigated. Upon the completion of this work the application to typical castings should be investigated.

The attainment of pressure tight sections in these bars by the selection of the proper thermal gradients indicates possible application of the data to improvement of mechanical properties. It would be interesting to determine improvement in mechanical properties which should accompany the production of pressure tight sections.

REFERENCES

1. Richard A. Flinn and C. R. Mielke, "Effects of Foundry Variables upon Porosity of 85-5-5-5 Bronze," AFS TRANSACTIONS, *vol. 66*, pp. 391-396, 1958.
2. R. A. Flinn and P. K. Trojan, "Pressure Tightness in 85-5-5-5 Bronze," AFS TRANSACTIONS, *vol. 64*, pp. 339-343, 1956.
3. John Kura, et al, "Mechanical and Physical Properties of Three Low Shrinkage Copper Base Alloys," *Proceedings of the A.S.T.M., vol. 58*, 1958.

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METHOD OF OBTAINING CASTABILITY AND MAXIMUM FIELD SERVICE FROM CAST PRODUCTS

By J. W. Beckham

ABSTRACT

In recent years, foundrymen have become more and more aware of the need, as well as the benefits, of working directly with the designer in creating a cast product to best meet the requirements of both parties. To obtain this mutual goal, the casting design must make the most efficient use of the metal in field use and also have good castability in the foundry. The method of casting design outlined has proved to be a satisfactory engineering approach to obtaining these two desired end results — good castability combined with equal stress distribution on the casting while in field use.

INTRODUCTION

Casting designers and foundrymen are rapidly recognizing that each has basically the same problem in castings whether it is stress concentration due to field loads or strain concentration set up by the cooling of molten metal in a mold. This condition has been aptly covered by Caine[1] in his comparison of stress concentration factors in corner radii of *L* and *T* shapes as related to their thermal gradients. As casting designs usually get more complex than these basic shapes, the writer would like to present a workable method of combining maximum castability with minimum stress concentration in the design of any cast shape.

An interesting fact demonstrated by this method of design is that the designer who works out a casting free from areas of high stress concentration will, in nearly every case, also have a shape that can easily be cast. Since these two conditions are closely related, the need for close contact between design engineers and the foundrymen from the inception of a casting design becomes even more obvious. Too much importance cannot be placed on this point as we are all realizing more and more the importance of communication between designers and foundrymen in formulating the most advantageous design for both.

Some six years ago, the management of the foundry employing the writer, saw the need for close work with the customer on casting design, and set up a separate Casting Design Department specifically for this purpose. This, of course, is an ideal situation, as it enables the foundry to engineer the casting for both maximum castability and minimum stress concentration and deliver it to the customer as a finished package. It should be said here, however, that it is not possible to do this to every new casting design coming into the foundry.

DESIGN PROCEDURE

To start a casting design through such a procedure one must obtain from the customer as much of the following information as is available:

Quantity Involved. This naturally will determine to a great extent the amount of design development.

Time Available for Development. This also will determine whether or not much effort should be spent on design. It must be remembered, however, that in many cases the method described will give faster results than putting an unproved design in the field for lengthy tests.

Direction and Amount of Load. The direction of loading on the part and its approximate magnitude should be carefully determined.

Clearance Conditions. Knowing the casting's environment in the finished product and including necessary operating clearances.

Weight Limitation. This is an important factor in most designs because it has considerable bearing on the casting cost, as well as on freight costs per unit. A good casting design utilizes all its metal efficiently to give maximum strength at the least weight.

The answers to these questions will then serve as a guide in determining the extent of design work possible or feasible on any one part.

After obtaining the above mentioned information from the customer and determining the extent of work required on any particular job, it is then possible to proceed with the design. Usually, a certain amount of free hand sketching, done either with the customer's engineering staff at the time of the initial contact or later after the information has been received, will help formulate a rough picture of what is needed. At this stage of design it is beneficial, at times, to also make up some kind of wooden or clay model of the ideas suggested. This does not have to be an actual scale model done by a professional, but can merely be a means of working in three dimen-

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sions to help envision a shape difficult to put on paper.

Calculations Necessary

Some mathematical calculations are also needed in formulating the initial design. These, however, can easily become complex when one tries to figure stresses on such odd shapes as are inherent in products of the cast process. To help make this initial slope of the design easier, however, Caine[2] recently has worked out a set of tables which can be beneficial. These tables give design information such as moment of inertia, section modulus, and radius of gyration on various sizes of tubulars, Omega, U and C sections all of which can easily be cast.

This information can be helpful when used as a guide in selecting an appropriate size and shape for use in designing a cast part, especially if one does not often use the complex formulas needed for figuring these properties.

Next in line in this design program is the preparation of a drawing of the finalized design giving enough information from which to make a pattern. In most cases the sample can be made from a master pattern. This will cut down the initial pattern cost, and makes it possible to change the pattern with a minimum of expense if it is found necessary. The master pattern when made is then turned over to the foundry for producing samples. This gives them an opportunity to determine the best gating and finishing methods to be used on the casting.

The finished sample is then thoroughly checked by magnetic particle inspection for any flaws. Facilities are also available for taking radiographs of the sample parts with cobalt 60 or cesium. Part of the inspection also is a layout of the sample to check its functional dimensions.

If the sample passes these checks, it is then fully machined if not used as a rough casting, giving the machine shop an opportunity to check finish allowance, chucking lugs and all other points with which they are concerned.

The machined part is now ready to put in the assembly in which it functions and the design analyzed by use of experimental stress analysis. It is important to run this analysis on a fully machined casting, as many times a good casting design can be ruined by bad machining practices, the worst offender being sharp corners.

STRESS ANALYSIS

Two methods of stress analysis are most commonly used in checking the finished casting. The initial use of brittle lacquer will locate the areas of high stress on the part and their approximate magnitude. This test can either be run in the field under actual conditions or in a laboratory under simulated loadings. Tests made in a laboratory will give more accurate results, however, if field loading conditions can be reproduced.

Having located these areas of highest stress on the casting, it is possible to then apply electric strain gages in these locations to obtain more accurate readings of the strain.

A more detailed description of the procedures used in making the tests described above is given in a paper[3] presented at the 1955 American Foundrymens Society Convention by the writer.

The information obtained by such tests is invaluable in determining the final design of the casting. If this first design is shown to be highly stressed in any one area, the shape can be changed to eliminate this condition using the overall stress picture as indicated on the part by stress analysis.

Redesigned Brake Shoe

To demonstrate specifically how this procedure works, the following example of the redesign of a brake shoe is submitted. The casting shown on the left in Fig. 1 had been made in this design for a number of years. In an effort to reduce the weight of the axle assembly using this brake shoe, the manufacturer was considering the use of a lighter fabricated type as used on some of the competitive axles. To prevent such a change, it was necessary to produce a cast part weighing no more, and having equal or greater strength, than the fabricated brake shoe.

The first step in formulating the new design was a stress analysis check of the original brake shoes in the complete axle assembly. Figure 2 shows the test setup using one hydraulic ram to expand the shoes and another to simulate the circular wheel loading on the brake drum. This test pointed out the areas of highest stress and their relation to lower stressed areas on the brake shoe. At this time, some of the fabricated shoes were also evaluated by stress analysis in the same setup. With this information, it was then possible to reduce the weight of the original design 25 per cent, and eliminate the area of high stress on



Fig. 1 — Trailer brake shoe. Original design is on the left, and redesign is on the right.

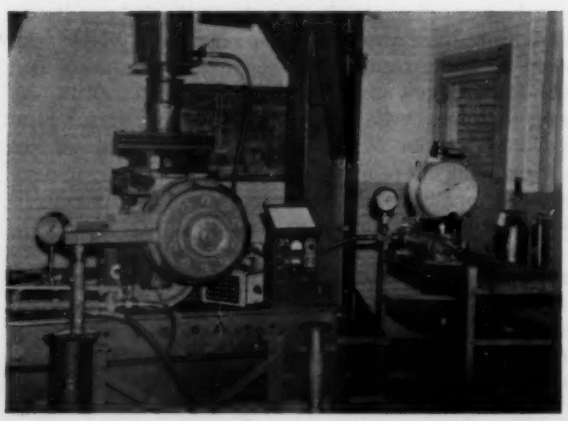
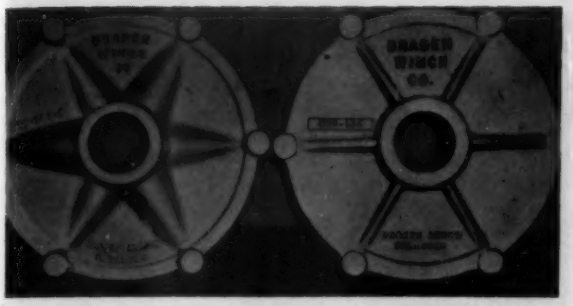


Fig. 2 — Laboratory setup for experimental stress analysis of brake shoes.





Figs. 3 and 4 — Winch housing end plate, top and side views. On the left of each figure is the redesign, and on the right is the original design.

the rib and in the notched section of the brake shoe's inner flange.

The new design is shown on the right in Fig. 1. Results of the stress analysis made on this redesigned shoe showed that in addition to reducing the weight 25 per cent, it was 1.94 times stronger than the original. This added strength was obtained merely by taking out the notch and rib on the inside flange and casting this surface with an uninterrupted curve. Small ears extending from both sides of this inner flange are drilled to serve the same purpose as the recess and rib used on the original design. Another advantage of the elimination of this high stressed area is the freedom from cracks and feeding problems formerly encountered there.

Winch Housing End Plate

Figures 3 and 4 show the top and side views of two castings used as a winch housing end plate. The original design on the right used opposed ribs across a flat plate to strengthen the hub. The junction of these opposed ribs across the plate created heavy sections in the casting that were not only difficult to feed but were slow to cool, thereby causing excessive cooling strains in the casting. Cracks on the outer edge of the casting were the natural outcome of this condition. The use of a corrugated plate section, as shown on the casting at the left in Figs. 3 and 4, eliminated the hot spots created by the opposed ribs solving the foundry problem.

In doing this it was also possible to reduce the weight 25 per cent, and increase the strength by distributing the concentrated stress that was created on the rib over a much wider area on the corrugation.



Fig. 5 — Plow sweep bracket. Original design is on the left, and redesign on the right.

Other Examples

Figure 5 shows a plow sweep bracket (left) which was failing in the strut (circled area). By a slight change in the angle of this strut and the use of larger fillets, the strength was increased 57 per cent with no noticeable weight increase (right).

Figures 6 and 7 show the top and bottom view of a trailer axle seat which, when redesigned, maintained equal strength with a weight reduction of 19 per cent and a nearly equal cost reduction. This also made a much easier job for the foundry.

Figures 8 and 9 show two views of a stalk cutter blade hub. The casting on the left was yielding in service on the ribs at small radius. Addition of a little height plus elimination of the high stress area on this rib doubled the castings strength with only 7 per cent weight increase (right).

Figures 10 and 11 are of a weldment and the composite weld conversion. This serves as a wheel arm support on a 15 ft mower. Four of these parts are used per unit—two right and two left. The steel casting was designed to permit its use either as a



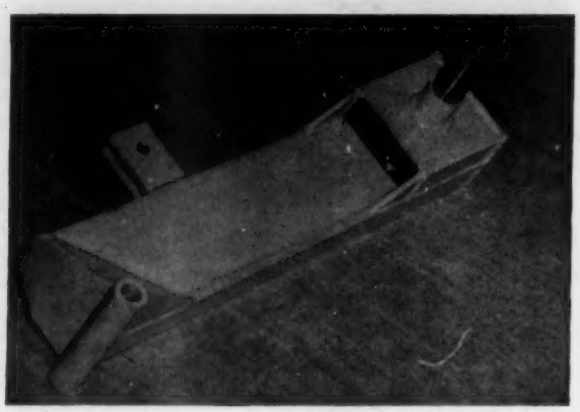


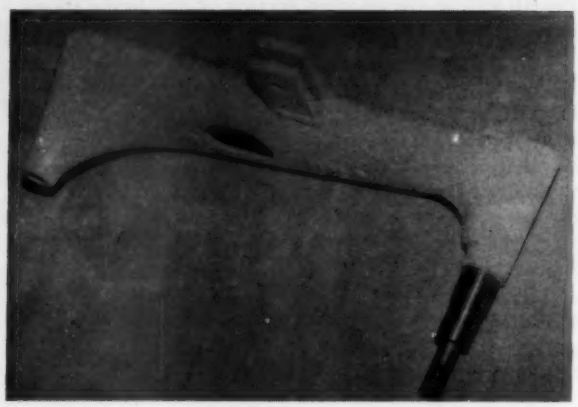
Figs. 6 and 7 — Top and bottom views of trailer axle seat. Original design is on the left, and redesign is on the right.





Figs. 8 and 9 — Top and side views of stalk cutter blade hub. Original design is on the left, and redesign is on the right.





Figs. 10 and 11 — Original fabrication (left) and composite weld (right) for a wheel arm support for a 15 ft mower.

right or left support. This is accomplished by welding the two ears on one side or the other. The spindle is turned from bar stock and also welded on. This design gave added strength, better appearance, at one third less cost.

CONCLUSIONS

The writer would like to emphasize the following points brought out in this paper.

1) Close contact between the foundry and the designer concerning any casting design or redesign from its inception will result in the maximum benefit to both parties.
2) A casting design free of unequally high stress concentrations in field use is, in nearly every case, one that can easily be cast in a foundry.
3) The use of experimental stress analysis in developing any casting design will give the designer the overall picture of stress relationship on the part. This, in turn, makes possible the design of a casting engineered for maximum field service, castability and minimum cost.

REFERENCES

1. J. B. Caine, "Interrelation Between Stress Concentration and Castability," A.F.S. TRANSACTIONS, *vol. 67*, 1959.
2. J. B. Caine, "Design Properties of Cast Section," Part 1 and Part 2, Report to Texas Foundries Inc., to be published in 1959.
3. Joe W. Beckham, "Use of Experimental Stress Analysis as a Casting Design Tool," AFS TRANSACTIONS, *vol. 63*, 1955.

PINHOLE OCCURRENCE IN MALLEABLE CASTINGS

bonding clays as cause

By D. R. Jones and R. E. Grim

ABSTRACT

Pinholes in malleable castings, it is the conclusion of the authors, are often caused by the evolution of water vapor from certain clay bonds in green sand molds when heated above 212 F by molten iron. Research on the problem was done after foundry experience revealed that pinholes occur less often when illite rather than other clay bonds is used. The amount and rate of water vapor evolution from each of four different clay bonds when heated above 212 F was recorded by laboratory methods.

The difference, it is believed, explains why the kind of clay bond used has such an important effect on the occurrence of pinholes. Theories on the formation of pinholes in steel castings are presented from the literature as possibly analagous to the mechanism in malleable iron. Practical foundry tests to support the laboratory investigation are also given.

INTRODUCTION

The occurrence of pinholes on or below the surface of heavy-sectioned malleable iron castings after annealing is, and has been, a costly problem to some malleable iron foundries. Some of these malleable foundries have successfully reduced the occurrence of pinholes by changing the clay bond. This paper explains why this change in bonding clay is effective in eliminating pinholes.

Users Recognition of Clay Effect on Pinholes

About four years ago at a meeting of malleable foundrymen the question of pinholes was raised. It was disclosed that several of the foundries represented had seldom if ever experienced pinholes, whereas other foundries in the same area producing similar castings were periodically plagued with pinholes. The fact that the foundries without pinhole problems each used an illite clay bond was not recognized as significant at that time.

One of the foundries (Foundry A) was experiencing a particularly heavy siege of pinholes. To determine whether their problem had any relation to conditions elsewhere, patterns and enough sand to make test molds were delivered to one of the foundries (Foundry B) that had reported no pinhole difficulties. Molds made with Foundry A's sand were poured off with metal from Foundry B. The castings shaken out of these molds contained pinholes. Similar molds made with Foundry B's sand produced castings with no pinholes.

The procedure was reversed. Sand from Foundry B was delivered to Foundry A. When metal from Foundry A's air furnace was poured into molds made with sand from Foundry B, no pinholes appeared on the castings. Since the sand and not the metal made the difference, Foundry A proceeded to duplicate Foundry B's sand. One ingredient at a time was changed with no benefit until the clay bond was changed to illite. This proved to be the critical item.

By changing only the clay bond, other foundries were also able to rid themselves of pinholes. At that time no one knew why the bond clay made a difference.

Thermal Analysis of Clays

A set of thermal curves was made at the Geology Dept. at the University of Illinois to determine the difference if any in gas evolution from the bonding clays in question when heated. Data indicated that the gas that creates pinholes could be caused by adsorbed water, which is released by some clays when heated to about 212 F or slightly above.

Apparatus and Procedure

The apparatus used to measure the rate and amount of evolution of water vapor when clays are heated is illustrated in Fig. 1.

The clay is heated in a glass or quartz tube enclosed in a small furnace. The tube is open at the ends, and a stream of nitrogen previously dried is passed across the hot clay under pressure. The nitrogen carries along the water vapor.

The nitrogen with the vapor then passes through a granular water absorbent, which picks up the water and causes a heat of wetting reaction. The amount of heat is dependent on the amount of moisture pickup. There is a differential thermo-couple junction at each end of the adsorbent and the temperature difference is plotted against the temperature of the clay.

The temperature difference gives a measure of the amount of water vapor evolved and, since it is plotted against furnace temperature, shows the rate of gas evolution as well as the amount. The depth of the dips of the curves indicates the amount of water vapor, and dip sharpness shows rate of gas evolution.

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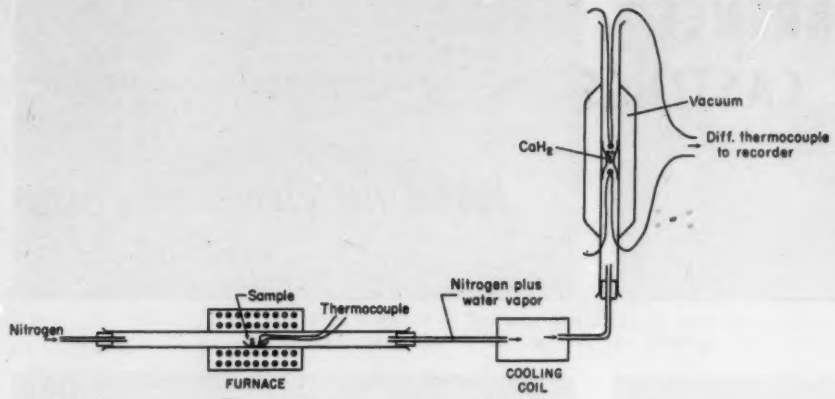


Fig. 1 — Schematic drawing of laboratory apparatus used to determine quantity of water vapor evolved, and the rate of evolution, from bonding clays.

The curves in Fig. 2 show the slight and gradual evolution of water vapor from illite, and the large expulsion of water vapor at low temperatures from the bentonites. This, the authors believe, is significant.

The samples tested were air dried. On air drying the bentonites retain considerable adsorbed water between unit layers of montmorillonite, their clay mineral compound, whereas fire clay and illite have only a little pore water and no adsorbed water similar to that in the bentonites.

In a green molding sand, the pore water in the illite and fire clay would be easily removed. In the bentonites, pore water would also be present, but there would be additional water between the unit layers which would be suddenly expelled when metal is poured into the mold. The adsorbed water is expelled abruptly at about 250 F.

Foundry Test

To prove the data, a test was made in a malleable iron foundry which had experienced pinholes on certain castings, and eliminated the pinholes by substituting illite for another bonding clay.

A pattern for one of these castings was selected and two molds were made. The cope and drag pattern surface of each mold was faced with southern bentonite bonded sand. One of the molds was torch

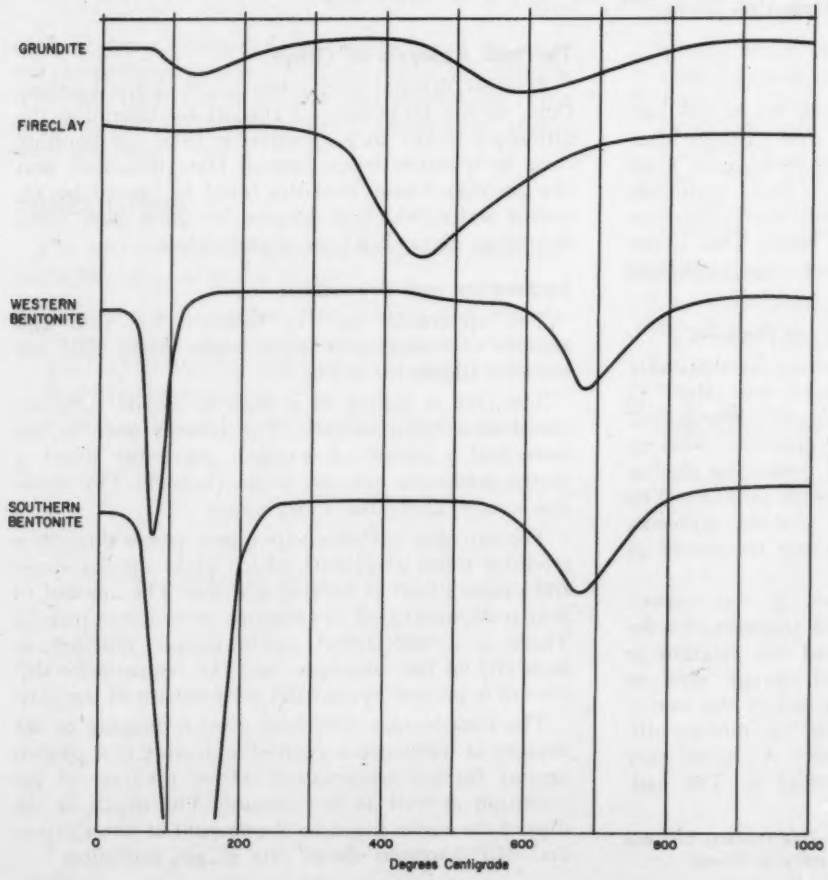


Fig. 2 — Thermal curves of bonding clays showing variations in water vapor evolution.

dried on both cope and drag surfaces. It was impossible to determine the temperature reached at the mold surfaces. A thermocouple placed ¼-in. back from the mold surface indicated a temperature of only 157 F during the torch drying. Since the temperature of the flame was over 1000 F it is safe to assume that all of the free water was removed.

It is reasonably certain, on the other hand, that none of the chemically combined water was removed, because when the torch was left in one place too long the sand surface would lose its bond and crumble indicating that the clay was calcined. The mold that was dried for this test and used did not reach the crumbly state.

The other mold was not dried. Both molds were poured off with iron from the same ladle within 10 min after being closed. The castings made in the flame dried mold, Fig. 3 (right), showed no evidence of pinholes after annealing, whereas the castings made in the undried mold produced pinholes (Fig. 3, left).

This work confirmed a test made by a large national concern that has made a study of pinholes in malleable iron and aluminum castings. This company oven-dried a bentonite-bonded mold prior to casting malleable iron, and was able to eliminate most of the pinholes. The incidence of pinholes is greater as the section thickness of the casting increases because of slower solidification of the iron. Also, pinholes are more apt to appear at reentrant angles of the casting where the adjacent mold surface is surrounded by iron to a greater extent than on straight surfaces.

The previously mentioned company makes malleable castings up to 5 in. in thickness, and their pinhole problem is consequently more acute than would be expected in more typical malleable casting production. They were able to eliminate pinholes on all but the heaviest sectioned castings when an illite bonded sand was used. When using other clay bonds they found it necessary to increase the seacoal addition up to 12 per cent as the section thickness of the casting increased.

It was found that whereas it required up to 12 per cent seacoal to react with the extra volume of water vapor given off by sand bonded with bentonite, the normal amount of seacoal present in the sand, say 4 per cent, was sufficient when illite was used. An

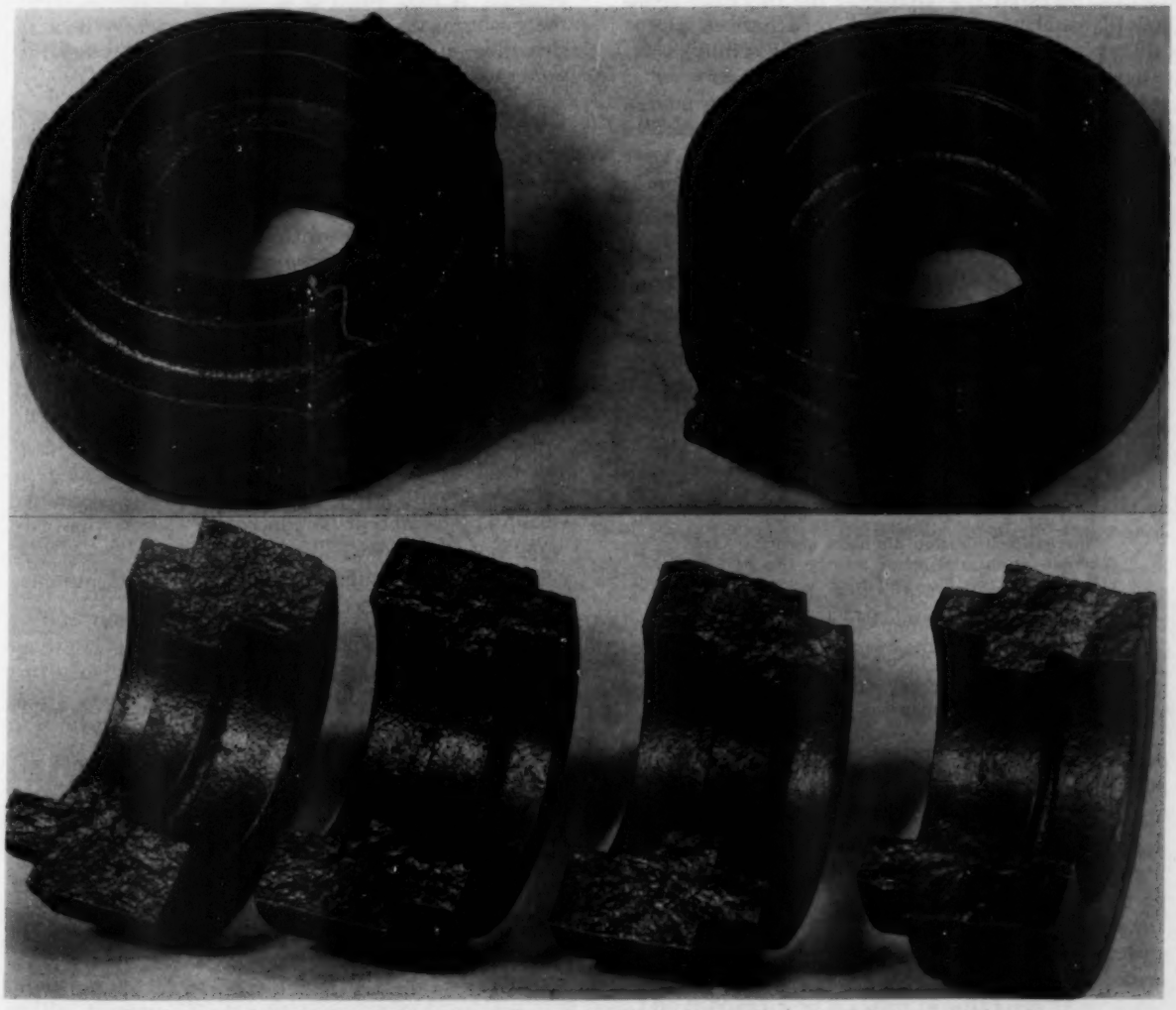


Fig. 3 — Complete and fractured casting with pinholes (left, top and bottom) made in undried molds, and castings free from pinholes (right, top and bottom) made in predried (flame dried) molds.

explanation for the usefulness of seacoal in this respect will be presented later. This foundry also found that molds made by the CO_2 process produced malleable castings free from pinholes.

This company concluded that pinholes are caused by entrapped gas in the metal. They determined that slightly more than half of this gas is present in the metal before it is poured into the mold. The rest of the gas, they determined, comes from green sand molds. The gas that is present in the metal before it enters the mold will not usually produce pinholes unless it is increased by gas from the mold.

Therefore, if water vapor from the molding sand is sufficiently low in volume such as in dried molds, CO_2 process molds, or illite bonded molds, pinholes will not usually occur. It also was determined that if hydrogen gas is removed from the metal before it is poured into the mold pinholes will not appear in the casting.

PINHOLE FORMATION IN STEEL CASTINGS THEORIES

An explanation for the appearance of pinholes in steel castings that is available in the literature seems logically analagous, and is presented here to give a fuller picture of the possibilities for pinholes occurring in malleable castings.

According to Sims and Zapffe,* when steel is poured into a green sand mold steam is generated. Part of the steam escapes through the facing sand, and some of it reacts with the hot steel thus:

$$H_2O + Fe \longrightarrow 2H + FeO$$

The FeO is found as a layer of scale on the surface because of its slow rate of diffusion. The hydrogen readily diffuses into the steel and raises the hydrogen content of the steel. The hydrogen then combines with the FeO to form a small bubble of water vapor, which grows when more hydrogen diffuses into the bubble.

> "When the growth of the advancing bubble slows down sufficiently because of depletion of hydrogen in the surrounding metal, the crystals seal off the passageway and a pin-hole remains.
>
> "Steels which develop pin-holes when cast into certain green sand molds may be free of pinholes when cast into well-dried molds."

According to Savage and Taylor,† pinholes in steel castings are caused by the reaction between dissolved carbon and oxygen from the iron oxide that is found at the metal-sand interface. Their contention is that carbon rather than hydrogen combines with oxygen and forms carbon monoxide with sufficient pressure to form gas bubbles in liquid steel.

Seacoal Effect

Strangely enough, in malleable foundries, seacoal in the sand, although a heavy gas producer, has a tendency to decrease rather than increase pinholes. An explanation for this is that the carbon in seacoal has a greater affinity for oxygen than does iron; therefore, water vapor in a green sand mold reacts with the seacoal to form CO rather than with the iron to form FeO.

The FeO is the basis for the reaction with hydrogen gas to form pinholes, as suggested by Sims and Zapffe; and FeO also has to be present to combine with the dissolved carbon in the metal to form carbon monoxide pinholes, as suggested by Savage and Taylor. Seacoal would prevent either of these taking place if used in sufficient quantity.

The explanation for the action of illite clay, it is believed, is that when it is used to bond the sand the evolution of water vapor is sufficiently slow so that most of the steam can be expelled through the sand rather than react with the iron to form FeO.

CONCLUSION

Pinholes may occur in certain malleable castings when the metal is cast in green sand molds. This possibility is reduced when the amount of water vapor available for reaction with the hot metal is minimized. It has been demonstrated that certain clay bonds used in green sand molding emit a smaller volume of water vapor and at a slower rate when heated than other clay bonds. In many cases pinholes are avoided by using these clays.

When water is removed from the mold entirely by heating prior to pouring metal into it, the absence of water vapor for a reaction insures a pinhole free casting in most cases. Further, when seacoal is used in sufficient quantity in green sand molds, the possibility of pinhole formation is reduced no matter what clay bond is used.

Therefore, a reduction in occurrence of pinholes in malleable castings can be obtained by:

1. Pre-drying the mold.
2. Increasing the seacoal content.
3. Removing hydrogen gas from metal prior to pouring.
4. Selecting a clay bond with a low water vapor evolution.

*C. E. Sims and C. A. Zapffe, "The Mechanism of Pin-Hole Formation," AFS TRANSACTIONS, *vol. 49*, p. 255, 1941.

†R. E. Savage and H. F. Taylor, "A Thermodynamic Study of Pinhole Formation in Steel Castings," AFS TRANSACTIONS, *vol. 58*, p. 393, 1950.

INDUSTRIAL ENGINEERING FOR THE SMALL FOUNDRY

a practical approach

By F. E. Noggle

ABSTRACT

Webster defines *practical* as given or disposed to action rather than to speculation. Therefore, this is limited to an exploration of a work measurement plan that will lend itself to immediate application, while the broader theoretical aspects of the subject will be largely ignored. Within the limits of practicability a further limitation will be observed, that is, all recommendations and conclusions will be made with a view to practical application in the small, as opposed to the large, foundry. For the purpose of this discussion, the small foundry is defined as one which has approximately 50 to 100 employees.

INTRODUCTION

The term *industrial engineering* covers a wide variety of activities in the management of a business. It includes planning the flow of work through the plant, time and production studies, planning plant facilities to increase efficiency, setting up wage schedules, conducting safety campaigns, determining product cost and cost control and estimating, to name some of the more common types of activity. However, the ultimate general goal of all these activities is to control and/or lower costs.

It is this general goal which will be kept in mind in this discussion, so that for the present purpose industrial engineering shall be defined as fact gathering, fact analysis, system development and application aimed at controlling or lowering costs.

One further question that presents itself is—wherein do industrial engineering plans for the large foundry and the small foundry differ? It is important to realize that the difference lies not in their respective needs for such plans, as it is just as important to the small foundry as to the large foundry that costs be controlled at their practical minimum level. Rather, the difference lies in the availability of personnel to carry on the work of industrial engineering.

In a large foundry, a typical industrial engineering department may consist of a time-study department of five or six standards men, a work simplification or methods department, an estimating department and a production control and planning department. These all report to the industrial engineer, who in turn reports to the works manager. Obviously, departmentalization on this scale is out of the question for the smaller foundry. A plan of practical industrial engineering for the smaller foundry must, therefore, be premised upon the employment of minimum personnel with specific individuals assigned to multiple duties in maintaining the system.

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PRESENTATION METHOD

Keeping the goal of practical application in mind, the plan must be presented in the most concrete terms possible. For this purpose, let us postulate a "typical" small foundry, based upon knowledge and experience with smaller foundries as they actually exist. In its general outline, it will be found that this typical small foundry presents the following picture.

Fewer than 100 are employed in this foundry. The company is getting along, but many times finds it difficult to obtain orders. Often orders will be lost to competitors by margins of a few cents per casting. The profit and loss statement does not show a good margin of profit, in fact losses occur periodically. Personnel relations are fair, but employees in general are not entirely happy. They ask for larger raises than business conditions seem to warrant. The scrap picture is not bad, but definitely could be improved.

Matters come to a head after several apparently good orders are lost to competitors by narrow margins. At this point the manager decides to find out what can be done about the situation. He may more or less consciously formulate the problem into the following questions:

1) How can we lower the production costs?
2) How can we pay more money to labor and still remain in a competitive position pricewise?
3) Can we somehow improve the estimating practices to attract more business, without lowering our profit?

FACT-FINDING

Report and Recommendations

Since this foundry is without an industrial engineering department, it is necessary to call in a qualified industrial engineer to make a survey of existing conditions, and to recommend a plan which attempts to realize the goals of cost control and of cost reduction which are implicit in the three questions above.

The engineer who is called in will know from experience that the work ahead can be roughly divided into three phases. The first of these is a survey de-

signed to ascertain all the facts which are needed to clearly define the specific problems, and to define the areas in which standards of performance (which are basically the means of controlling costs) should be set up. The outcome of this survey will be a report in which problems are clearly defined and a specific system of cost control is outlined.

The second phase is setting up the system which has been evolved to meet this company's problems. It includes the gathering of standard data upon which performance standards will be based, and the training of company personnel who will be responsible for maintaining the cost control system after it has been installed.

The third phase of activity is maintenance of the system, including a method of performance reports to management. This third phase of the system is carried on by company personnel who have undergone the training of phase two.

What, now, are likely to be the contents of the industrial engineer's report after he has completed his fact-finding survey? Since our hypothetical foundry is "typical," the engineer's report is likely to include the following points under the headings of Wage Payment, Estimating Practices and Production Control.

Wage Payment

1) About 30 per cent of all labor put in is covered by incentives.
2) The average productivity of the employees who are on incentive is only approximately 90 per cent of the nationally accepted norm, although their earnings are 150 per cent to 200 per cent of the operator's base rate as it exists in this foundry.
3) The type of incentive plan in use is a piecework system. Piece prices are determined by comparison and/or a direct time study.
4) Production recording is far from accurate, especially in the matter of recording delays and other time allowances.
5) Counts of pieces produced are fairly accurate, but earnings are greatly affected by guessing at the length of delays.
6) Distribution of wages is causing labor relations problems. This is shown in Fig. 1, where it may be seen that actual earnings of less skilled occupations are equal to and even higher than the earnings of those in jobs requiring more skills. This situation exists because piecework prices have been established for operations such as squeezer molding and blower coremaking. The molding of loose work, floor molding and similar activities have been left on daywork.
7) The present system is so shot through with uncertainties and guesswork that it does not actually provide management and supervision with a means of controlling labor costs. Much needless waste is not even guessed at until it is demonstrated by a poor profit and loss statement.

The engineer's recommendations under this heading are:

1) Replace the present piecework system with a measured time system based on standard data.
2) Extend the incentive coverage from the present 30 per cent to at least 80 per cent of all labor using the measured time system, for the purpose of gaining better control of labor costs, gaining better distribution of wages to eliminate grievances and labor relations problems and to lower labor costs through higher productivity.
3) Establish a timekeeping function for the purpose of accurately recording delays and other off-standard time. This same timekeeping function will also perform a certain portion of the payroll calculations.
4) Develop a system of reporting vital information to supervision and management for the purpose of readily detecting waste and controlling labor costs.

Estimating Practices

Under this heading the engineer points out:

1) Estimates are made by comparison with similar castings. Actual past performance is here considered the substantiating factor.
2) Overhead costs are applied on a department-wide basis. For example, the same overhead is applied to core blowing and bench coremaking.

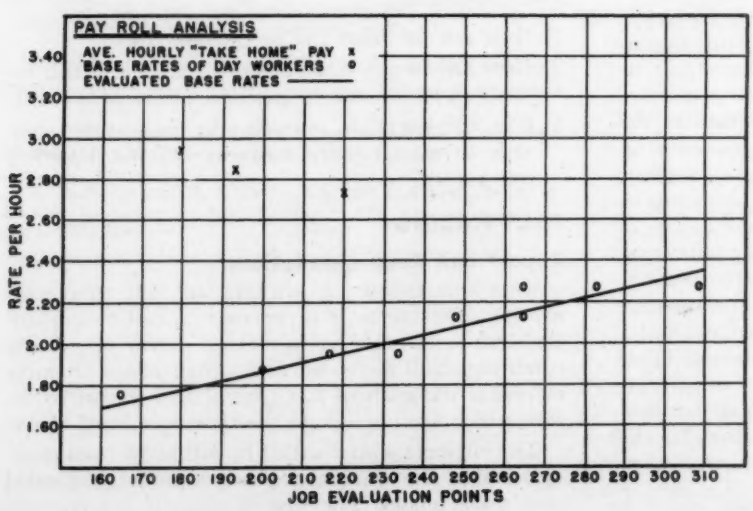


Fig. 1 — Distribution of wages causing labor relations problems. Actual earnings of less skilled occupations are equal to and even higher than those jobs requiring more skills. This is because some operations have piecework prices established, while others do not.

3) Cleaning costs are calculated at a fixed rate per pound for castings.

In making his recommendations the engineer will aim at correcting certain serious costing errors which result from present practices. For example, not only can comparisons with similar castings be misleading, but actual past records of performance on those castings may lead to unbelievably high labor costs with resultant noncompetitive pricing. At first glance this may appear a paradox, since what can be more accurate than an actual record of production? However, the fact is that at best such records show only what was done and not what could or should have been done in production.

At worst, the record may not even show what was done. It may not show what difficulties were encountered, whether the sand was too wet or too dry, that the cores were not ready and the molder waited until they were delivered, or 101 other things which make jobs take longer than they should. Thus, the estimator may find himself in the position of attempting to use an unknown quantity as the basis for comparison in his own important task of arriving at prices that will bring business into the plant.

Case History

The degree of error in estimates and quotations based on past performance can be shocking when it is finally brought to light. To illustrate this, consider the case of an actual company which is on record. Over a period of a year overall performance had been accurate within 5 per cent of estimated time on operations. However, standard data for incentive application had been developed, and it was decided to take a specific job and find out how the estimate and the new standard compared. The estimate, based on past performance on various jobs, was given as 141 hr.

However, the new performance standard allowed for only 97 hr, or only 65 per cent of the estimate. Needlessly, the superintendent ridiculed the lower figure, since he felt he could depend upon past experience. However, when this job was put on incentive based on the new standard, it was actually carried out in only 84 hr. Moreover, performance continued at approximately the same greatly improved rate of productivity wherever the new standards and incentives were applied. Similarly unrealistic estimating is exceedingly common wherever the foundry depends upon past performance as a measuring stick.

Obviously, if past performance was poor, it will only result in continuance of needlessly high cost when it is followed as a model. At the same time, past performance times may have been based upon methods or equipment that have now been replaced by more efficient means which are not, however, taken into consideration, and therefore do not actually bring about the cost savings for which they were intended. Thus, we find that the engineer's report contains these recommendations under this heading:

1) Collect costs by cost centers.
2) Establish a normal capacity for each cost center in terms of measured time.
3) Calculate a costing rate per unit of time for each cost center involved.
4) Develop a system of reporting variances from the standard for management for the purpose of controlling manufacturing costs.

Production Control

The final heading of the engineer's report shows these facts:

1) Delivery dates are frequently missed because of lack of sufficient information regarding progress of production.
2) Customers' telephone inquiries regarding items on production are difficult to answer because production information is not centralized, but is scattered through various departments.
3) Frequent errors in production records are caused because records are kept in various places.
4) Losses frequently occur because of overrun on orders.
5) Losses also occur because of lack of coordination, especially between core room and molding departments. These losses consist of lost molding times when cores are not ready on time, and of broken cores caused by storing and re-handling when made too far in advance.

The engineer's recommendations under this heading are:

1) Centralize in one office all records pertaining to production.
2) Improve coordination of production, again by centralizing sufficient information in one place.
3) Develop an integrated data processing system which will reduce errors and at the same time reduce clerical work.
4) Tie in workers' incentive payments with quantities on order to prevent unauthorized overruns.
5) Install some type of visual record showing status of orders so that pertinent information regarding progress is readily available.

SETTING UP THE COST CONTROL SYSTEM

Having ascertained and analyzed in his report problems which are involved, the industrial engineer is now ready to enter the second phase of his work. This phase is the development and installation of a system based on his recommendations for overcoming the problems.

Basic to his thinking must be two important facts. First, this smaller foundry can no more afford to ignore a situation that prevents control of costs, leads to noncompetitive pricing and poorly coordinated, wasteful production methods, than a larger foundry can. On the other hand, unlike the large foundry, the smaller company cannot afford to hire a staff of highly skilled engineers to maintain the new cost control system. This means that the outside industrial engineer must now train a group of selected company personnel so that they will be able to carry out the proposed system without further outside help.

The training program will involve three groups of personnel:

1) Those who will make studies and develop work measurement standards.

2) Plant foremen who will have direct control of labor costs.
3) A selected group of workers who will be taught to understand the workings of the new incentive plans.

Training of Standards Men

Personnel who will be trained to develop and maintain the cost control system must be picked with care. Certain psychological tests of aptitude and interests are available which can assist in selecting qualified people. This test should be given to enough persons so that there can be a selection of individuals who are to receive time-study training. At the end of the training session a final group will be selected who are to do the actual development work.

It may well be that only one or two qualified time-study men will be needed to maintain the system after it has been developed. If more men were trained and used in development, it is not necessary for them to remain as white elephants in the standards departments, since they will be highly qualified for supervisory positions in the organization, such as in production control or as supervisors, foremen or estimators.

To educate the selected group in problems of time study and methods of developing standard data, about six weeks of intensive training are required. In this time the group will learn to read a stop watch accurately, to performance rate operators, as well as learn how to establish allowances for off-standard time properly and to analyze methods. They will learn the basic problems of standard data development, as well as something about work simplification. While this course of training will not turn out accomplished industrial engineers, it will place the men involved in a position to start actual development of standard data under the guidance of an instructor.

Supervisory Training

The main purpose of supervisory training is to teach personnel how to use the labor cost control tool which will later be placed in their hands, since the main purpose of the incentive system will be to lower and control labor costs not simply to pay more money to workers. These supervisors should learn enough about time study and standard data to gain the understanding which will result in confidence in the new system as it is being developed. They need a working knowledge of standards so that they can understand the effect of methods and methods changes on standards, and so that they can readily answer questions asked by workers.

In the smaller foundry it is advisable to have supervisors know the principles of work simplification. They can gain practical experience in this phase of the program during the training courses under the guidance of the industrial engineer. Since the standards department of a small foundry is seldom in a position to initiate and carry out work simplification projects, this aspect of supervisory training becomes quite important, and much can be accomplished through the men who are on the spot where actual production is in operation.

Worker-Training

The training course for workers is aimed at giving them an insight to standard data, how it is developed and how job standards are factored from data. These workers should learn enough about time study to understand that performance rating can be reasonably accurate when the time-study man has had proper training. Such insight and general understanding is necessary, since it brings about faith and confidence in operation standards and furnishes background which enables workers to discuss their problems intelligently.

Selecting and Planning the Project

At the outset of the program all personnel involved, including supervisors, foremen and workers, should be made fully aware of what the project will be, what management plans to do and why it intends to do it. It should be stressed that management wants both to bring about more equitable distribution of wages, and to provide an incentive opportunity to more workers. It should be spelled out that this will result in better relations through eliminating the causes of grievances. Mention should also be made of the need for lower labor costs and for more efficient control of these costs in order to improve the company's competitive position.

The decision as to where to begin in the actual development of standard data must be based on careful consideration as to which department or operation should be studied first. In making this decision, it must be remembered that the time-study men involved will be inexperienced. Therefore, an operation should be chosen where the elements of the operation follow a fairly standard sequence, and where the majority of these elements are of sufficient duration so that reading the stop watch will not be difficult.

It is also important to select an operation where substantial savings are likely to result, whether because of the large number of workers involved, or because the operation is known to be a bottleneck and increased production will benefit other operations as well.

Getting Started

Having trained time-study men, and having selected a logical department or operation as a starting point, the next step is the actual work of developing standard data. It is possible at this point to go far astray from what is practical, by attempting to go to the extremes that are represented by ideal projection. It is far more important in getting started to aim at definite improvement, even though room for further improvement may still remain.

In the matter of methods, for example, the time required to carry out a complete methods project may result in long delays due to an insistence upon having only the best methods before developing data and initiating incentives. Savings which could be accrued from incentive application would be lost during the prolonged time the methods project was in progress.

It is far more practical to adopt the view that work simplification and methods analysis is a project

separate from incentives, and should continue year after year. Then, when methods changes are made standards can also be revised to obtain the savings which result from the further methods improvements.

However, minor methods changes will need to be made before the development of standard data is undertaken. For example, inconsistencies between individual operators must be eliminated so that each person is performing the operating in practically the same manner. If this is not done it becomes impossible to correlate data from time studies into standard data charts.

For example, in bench coremaking the placement of core racks may vary at different benches. One coremaker may only have to turn around to place his cores on a rack, while another may have to walk several feet to do this. In this case appropriate steps must be taken to place racks in the same relative position to each bench before standard data can be developed.

Similar inconsistencies between the methods used by individuals may also be found in operations such as riddling sand, or within the walking distances required for obtaining facing sand or cores in a slinger molding department. Such methods improvements can be undertaken after one study, and without undue delay.

Inadequate Information

Another major pitfall which must be avoided in the development of standard data lies in an inadequate descriptive write-up of the operation, the department, the methods and materials used. Unless this descriptive record is as complete as possible in all details, it may prove impossible to maintain the cost control system efficiently after it has been installed.

This is true because changes in methods, materials, layout, equipment or work sequence can naturally change what the standard should be so that the old standard of performance is no longer realistic in terms of the changes which have come about. However, with an adequate descriptive write-up, it is possible at any time in the future to check back to determine if changes affecting the standard have been made.

For example, there is on record a company which had developed standard data for core room operations. During the course of about ten years, average performance increased until it had reached 200 per cent of standard. Obviously, something had changed during those ten years, but proof of change could not be made because standard data did not contain a description of the methods, the materials or the equipment. It did not even contain a sketch of the layout of the department as it existed at the time the standard data were developed.

Had these things been included in the original descriptive write-up, the company would have had available substantiating proof of change and could have corrected their standards accordingly. They had gone on for years paying largely unearned wages which greatly and needlessly increased their costs.

More fortunate was the company on record which, when it found the performance of the squeezer molding crew unreasonably high, readily determined from the descriptive write-up that the work elements of "carry out mold" and "return to machine" were performed on a sand floor. Concrete floors had been installed, along with other improvements which were also readily detectable, which made the old performance data obsolete. It was a relatively simple matter to correct standards to take advantage of the savings that should have been affected by the improvements.

What is required in the descriptive write-up of a set of standard data? It should cover in minute detail these items:

1) Conditions under which application is to be made, such as individual or group incentive.
2) Tools and equipment used, including a list of machines and their serial numbers, sand formulas, grinding wheel make and identification numbers, surface speed of grinding wheels, horsepower of motors, a list of small tools used, etc.
3) Description of each work element, and the "break point" (point at which the element ends and the watch is read in making time studies).
4) A description of the general process, telling where raw materials are stored, how they are delivered, what is done with finished pieces, who takes them away, etc.
5) A sketch of the relative position of the department to other departments and to the plant as a whole.
6) A sketch showing the individual work stations within the department, giving all the appropriate dimensions.
7) A sketch of one work station (if they are all laid out alike) within the department, giving all appropriate dimensions.
8) A description of the development of the data, showing which elements have been combined in each of the standard data charts together with the standard time for each element.
9) Instructions on how to factor production standards, including sketches explaining classifications.
10) Examples of factor cards used for the operation.
11) Description of all allowances included in the standard data.
12) Description of what constitutes delays and how they are to be paid.

A descriptive write-up covering the above points, provided it is carefully and adequately written, can prevent many of the difficulties encountered over a period of time because of "creeping changes" in methods. In general, it must be kept in mind that the twin pitfalls of standard data development are on the one hand over-complication, and on the other hand, over-simplification. Data should not be complicated to the point where the cost of establishing a production standard is too high. However, if data remain too simple there is danger that earnings will not be consistent with actual effort.

Setting Up Cost Control

It must never be forgotten that all of the activity involved in developing standard data and incentives is ultimately aimed at the primary goal of lowering (wherever possible) and controlling labor costs. Just

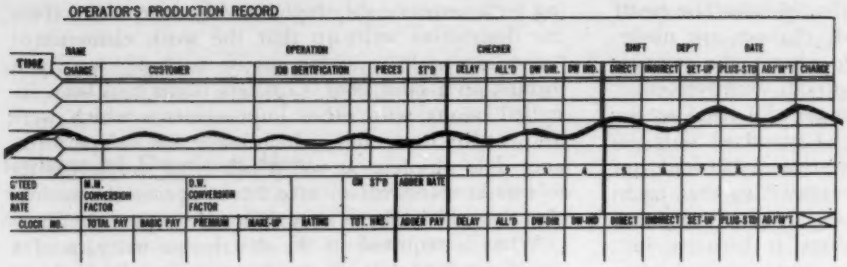


Fig. 2 — Daily production record form which is designed with cost control in mind as well as payroll.

as accurate measurement is the heart of accurate cost establishment, so effective reporting of conditions which affect costs is the heart of continuing cost control.

To be classed as effective, a reporting system must present conditions to management and supervision in a manner which not only accurately reflects conditions, but reflects them accurately before too much time has elapsed, before wasteful situations on tendencies have been allowed to go on long enough to do real harm. In addition, reports must be presented in a manner which is easily understandable and does not require too much time to analyze. If such a reporting system is not installed, no cost control system can be considered practical.

Reporting System

The effectiveness of a reporting system depends in large part upon the creation of paper forms for recording. Figure 2 shows a daily production record form which is designed with cost control in mind, as well as payroll. At first glance this form may appear complicated. However, it is actually easy to use, and is designed to segregate important key cost items. Thus, the first division is between standard and non-standard labor costs. This provides the basis for determining what portion of the payroll is on standard, and, therefore, controlled or controllable.

Further subdivisions go into direct, indirect and other categories for the purpose of determining what is production and what is waste on the payroll. Finally, all operators' daily production records for a given department or cost center are then summarized on a cost control summary (Fig. 3). This is primarily a work sheet designed for the purpose of collecting all labor costs for a given cost center and presenting them in a summarized form. The summary appears at the lower right hand corner of the sheet, where we find the pertinent labor cost information which supervision needs to control labor costs.

The pertinent items here include all waste items. These are (1) overtime premium, (2) make-up pay, (3) delay pay and (4) plus standard pay. Generally, the total of waste costs should not exceed 5 per cent of direct costs, and experience with a given cost center may establish an even lower norm. After that, at any time that the waste cost exceeds either 5 per cent or the established norm, supervision should immediately analyze the report to determine the major cause of waste, and then to bring costs back into line.

The summary sheet (Fig. 3) next lists the rating and the per cent of time on standard of the operators for the period covered by the report. The rating should consistently remain well over 100 per cent, and time on standard should never fall below 85 per cent. When either of those figures are violated, costs may soar to an alarming extent.

The next item listed in the cost control summary (Fig. 3) is the per cent of capacity at which the cost center concerned operated for the period reported. This item is not always controllable by supervision, but it does affect production costs through the distribution of overhead costs. In establishing a standard cost for the cost center, overhead costs are calculated so as to be absorbed at some pre-determined percentage of capacity. If and when actual production drops below this percentage of capacity, overhead costs are not fully absorbed and losses occur.

The next items on the control summary are direct cost/direct hour and total cost/direct hour. The direct cost is calculated by dividing the direct dollars by the direct (salable) hours produced, and reflects the average hourly base rate of all operators in that cost center. This could be considered an ideal cost/

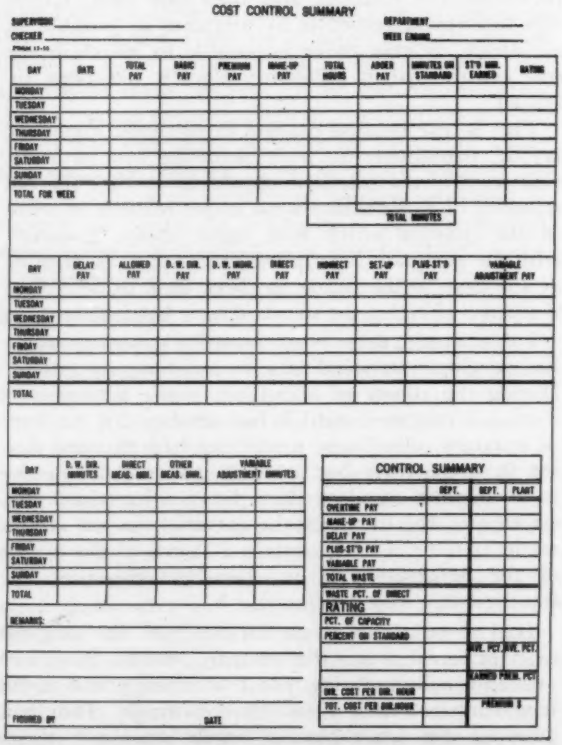


Fig. 3 — Cost control summary record. This is primarily a work sheet designed for the purpose of collecting all labor costs for a given cost center and presenting them in summarized form.

hour, since it is not possible for labor costs to be lower than the average base rate. Total cost/direct hour is found by dividing the total pay for the cost center by the direct salable hours produced. This figure reflects the final results of efforts to control labor costs.

Reporting with the view to controlling labor costs is most effective when quick comparisions can be made with the results of previous weeks. For this reason it is advisable to transfer the cost control figures we have explained above to a cost control record (Fig. 4), which shows progress for a full year. This affords a graphic record of the progress each supervisor is making toward controlling his labor costs.

Estimating and Pricing

At this point, having developed standard data for all direct operations and most of the indirect operations, and having instituted the necessary cost control report system, the next step is to develop a realistic estimating procedure.

The estimating form shown in Fig. 5 was designed for, and is used by, a small non-ferrous foundry. This form serves as a check list to remind the estimator of each item of cost which must be considered in calculating the selling price. Basic to its formulation is the use of standard cost conversion factors for each cost center in the plant. Without entering into detail of how this cost conversion is made, it should again be pointed out that it is based upon the standard data achieved through accurate measurement and analysis when establishing performance standards.

Production Control

Experience has proved that at this point of cost control development, all the pertinent information necessary to practical production control systems has been established and recorded. Developing an effective production control system, therefore, is not a matter of obtaining new or different data, but simply a procedure of consolidating and integrating data already available. Invariably this results in eliminating wasteful duplication of effort and record keeping, and at the same time bring essential data together in forms which are more convenient to handle and more accessible when required for use. The possibility of error is held to a minimum.

The chief purpose of a production control system may be divided into three practical, imperative goals. They are:

1) Establishing a promised delivery date on the basis of actual production conditions, rather than of wishful thinking.
2) Meeting that delivery date.
3) Reducing production costs.

Needless to say, delivery dates quite often are farcical in common practice due to lack of production control. It seems almost unbelievable that this condition should ever be allowed to exist, in view of the fact that to the average buyer of castings, service rates second only to quality while price rates third, behind service. Needless to say, the establishment and meeting of firm delivery dates can only result in greatly improved customer satisfaction with obvious bearing on sales volume.

In analyzing the information required for production control, it will be found that two categories of information are involved. These are:

1) Information necessary to produce the order.
2) Information relating to the progress of production.

The information necessary to produce the order must appear on shop orders. A considerable amount of clerical work which is usually required to produce a shop order can be saved, when it is considered that about 90 per cent of the information involved remains the same for any order received for a given pattern or part number. Essentially, the only information for a given pattern or part which varies from one order to the next consists of items such as order number, quantity on order, delivery dates required and promised.

Basic Order Forms

If modern duplicating equipment is used to produce an adequate supply of basic order forms for specific jobs to which they might apply, it only remains to add the specific information regarding quantities and deliveries to make the order complete. Figure 6 reproduces a permanent duplicating master order form which is re-issued every time a new order is received. The exact information appearing on it may be varied from foundry to foundry as required, but basically it will have the same information.

In ordinary practice, much of this information will be scattered in various offices or departments of the plant, with much of it never appearing on the shop order. The advantage of having it centralized on the shop order is that this practice eliminates not only the necessity of looking up information in other records, but actually eliminates the need for keeping other records at all.

For example, the order delivered to the pattern storage department is not only an order to send the pattern to the shop, but also instructions on where to locate the equipment, thus eliminating the need to search in pattern storage files for this information. Again, by returning the mold order with the pattern, the equipment can be stored properly without referring to any other records. In the same manner, shipping instructions on the order also eliminate the need for special record keeping in the shipping department, with consequent loss of time in keeping and checking needless records.

Information relating to the progress of production can also be simplified through centralization. Figure 7 reproduces a production control record card which is maintained in the office of the person responsible for the control of production. It provides, in one central location, the answers to all questions, either from the customer or management regarding the status of the order. This card, which is the heart of the production control system, can be designed for mechanical posting of data, but the smaller foundry will usually find that manual entry and calculation produces satisfactory results.

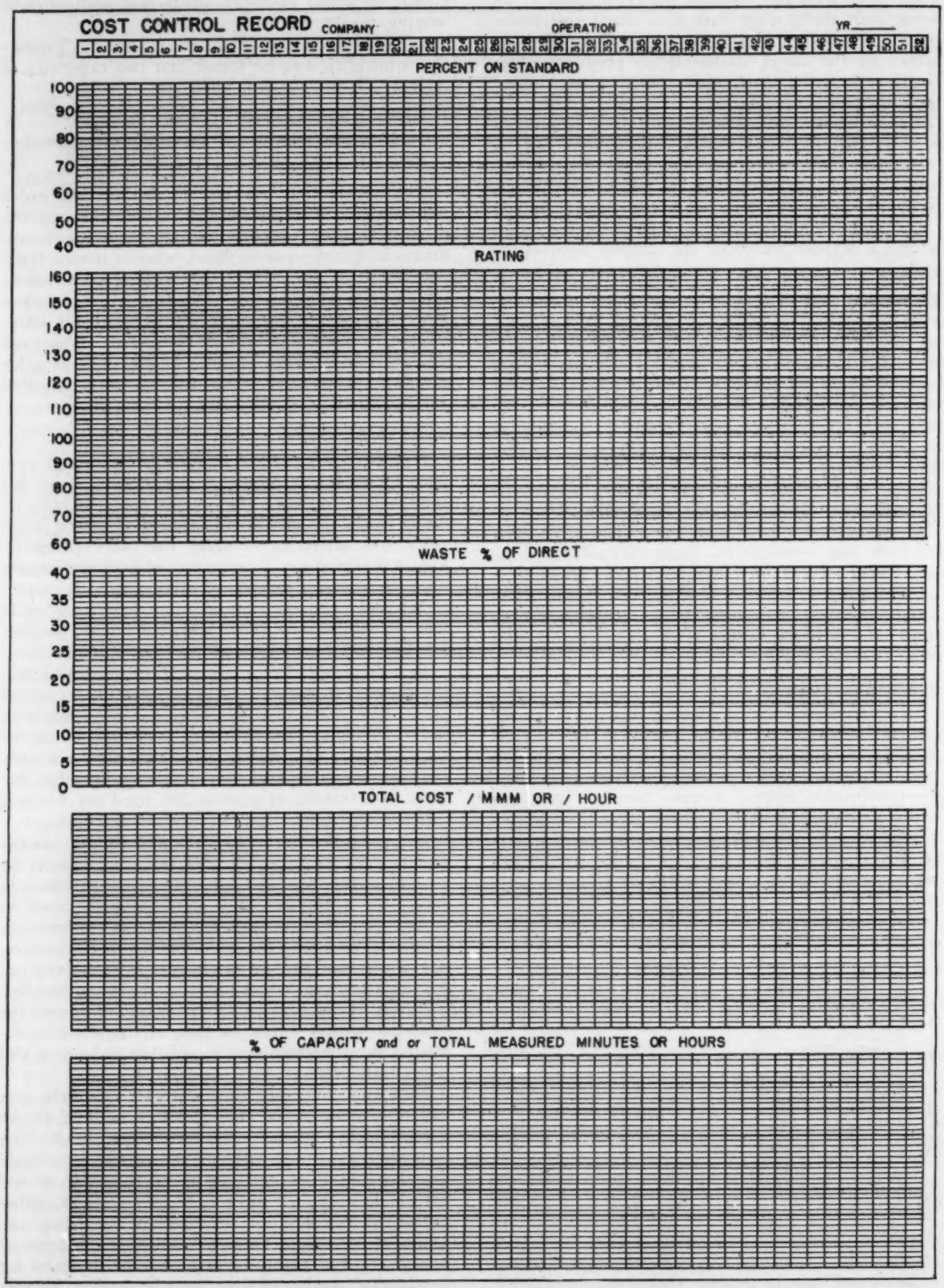


Fig. 4 — Cost control figures are transferred to the cost control record which shows progress for a full year. This affords a record of progress each supervisor is making toward controlling his department's labor costs.

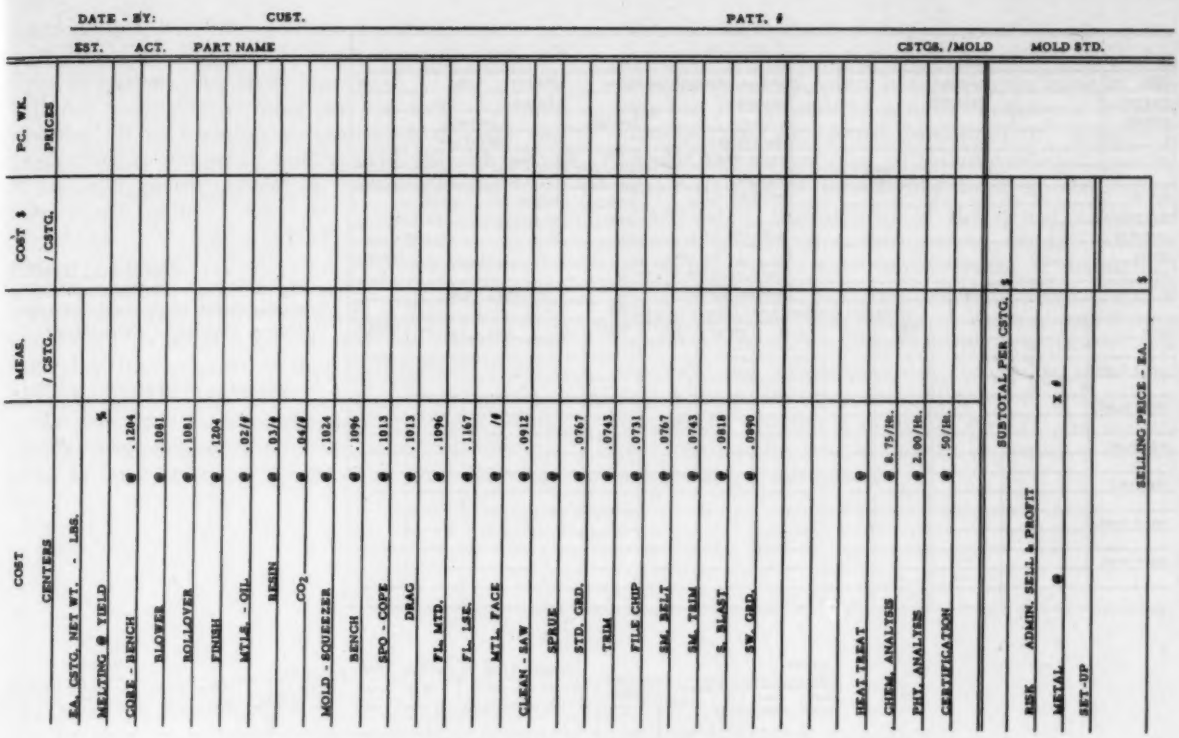


Fig. 5 — Estimating form designed and used by a small nonferrous foundry. This form serves as a check list to remind the estimator of each item of cost which must be considered in calculating selling price.

A system of colored flags attached to the production control record cards serves as a visual means of detecting any order still on file which is overdue. These colored flags, designating the month and week in which shipment is promised, also serve to establish that orders will be scheduled in the proper sequence.

Maintaining and Carrying Out the Cost Control System

The size of the plant, and to a certain extent the quantity of new and repeat orders received weekly, will determine the number of people required to maintain the cost control system which has now been developed. In the small foundry it is usually necessary for one person to double in more than one capacity, or even to assume responsibility for all of the various industrial engineering activities. However, regardless of the number of people involved, the various functions of the work remain the same for all practical purposes. It should be kept in mind that, at this point outside industrial engineering help is no longer utilized and that the work is carried on by plant personnel.

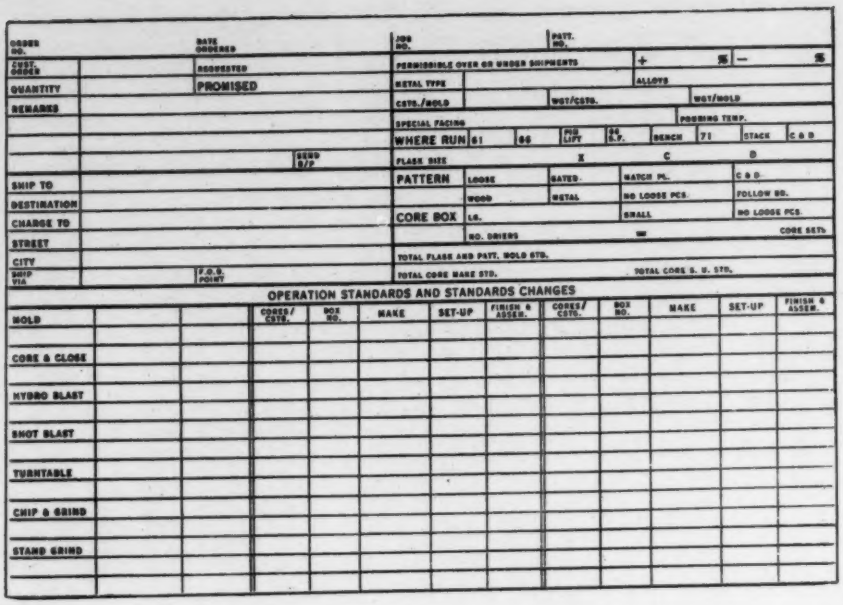
If the cost control system is not effectively maintained, all of the work of measurement, standard development and personnel training that has gone before will be wasted, since it is only when applied that the system will serve to control costs. Various factors are involved in maintaining the system effectively.

First, there is the maintenance of the incentive system. This must be done carefully and accurately, but it becomes almost a clerical function since the data on which it is based have already been developed. Foremen, particularly, must be alert to improvements in methods, so that they will be reported and considered in changing operation standards. Unless this is done the incentive system will fall far short of its purpose of lowering labor costs, since the savings which are supposed to come with the methods improvement will not be realized.

It is wise to make a periodical audit of the system to establish that standards are being kept up-to-date. The descriptive standard data write-up of methods in use in each department must be constantly scrutinized, and kept meticulously up to date, to avoid unrecorded creeping changes which can come over the years. If this is done, the incentive system will continue to serve its purpose of cost reduction, and savings, which are the result of methods improvements, will not be thrown away in the form of unearned wages.

In maintaining the estimating system, the actual job of estimating is best done by the standards man or standards department. The reason for this is that day-to-day use of standard data in setting up operation standards has provided the standards man with a knowledge of standard practices so necessary to accurate estimating. Also, the standards man should compare his estimates with actual operation standards which are set for incentive purposes after the order is received.

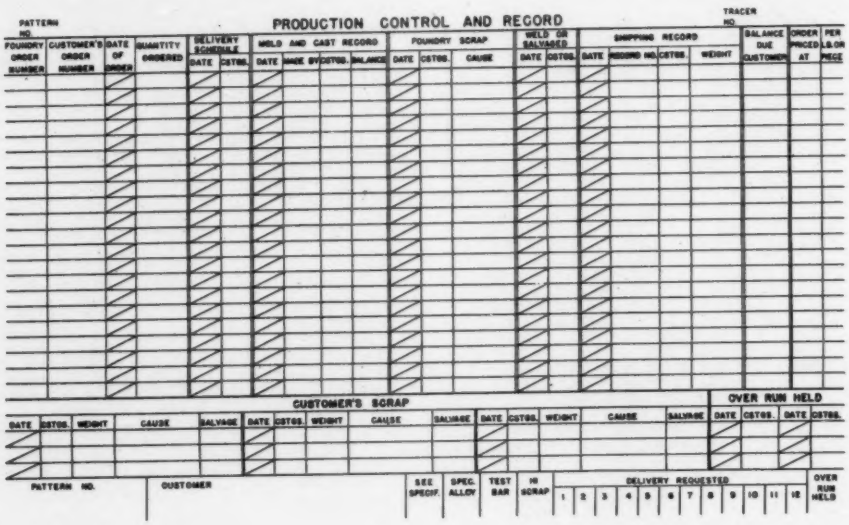
The maintenance of the production control system is essentially simple. Only if the type of production involved in the plant undergoes major changes, such as going from a short run job shop to long pro-


ORDER NO. | DATE ORDERED | JOB NO. | PATT. NO.
CUST. ORDER | REQUESTED | PERMISSIBLE OVER OR UNDER SHIPMENTS + % – %
QUANTITY | PROMISED | METAL TYPE | ALLOYS
REMARKS | CSTS./MOLD | WGT/CSTG. | WGT/MOLD
SPECIAL FACING | POURING TEMP.
WHERE RUN 61 66 MID LIFT 84 B.Y. BENCH 71 STACK C & D
SEND B/P | FLASK SIZE X C D
SHIP TO | PATTERN LOOSE GATED MATCH PL. C & D
DESTINATION | WOOD METAL NO LOOSE PCS. FOLLOW BD.
CHARGE TO | CORE BOX LG. SMALL NO LOOSE PCS.
STREET | NO. DRIERS CORE SETS
CITY | TOTAL FLASK AND PATT. MOLD STD.
SHIP VIA | F.O.B. POINT | TOTAL CORE MAKE STD. TOTAL CORE S. U. STD.

OPERATION STANDARDS AND STANDARDS CHANGES

			CORES/ CSTG.	BOX NO.	MAKE	SET-UP	FINISH & ASSEM.	CORES/ CSTG.	BOX NO.	MAKE	SET-UP	FINISH & ASSEM.
MOLD												
CORE & CLOSE												
HYDRO BLAST												
SHOT BLAST												
TURNTABLE												
CHIP & GRIND												
STAND GRIND												

Fig. 6 — A permanent duplicating master order form re-issued every time a new order is received.


PRODUCTION CONTROL AND RECORD

PATTERN NO. | TRACER NO.

FOUNDRY ORDER NUMBER	CUSTOMER'S ORDER NUMBER	DATE OF ORDER	QUANTITY ORDERED	DELIVERY SCHEDULE DATE	DELIVERY SCHEDULE CSTGS.	MOLD AND CAST RECORD DATE	MOLD AND CAST RECORD MADE BY	MOLD AND CAST RECORD CSTGS.	MOLD AND CAST RECORD BALANCE	FOUNDRY SCRAP DATE	FOUNDRY SCRAP CSTGS.	FOUNDRY SCRAP CAUSE	WELD OR SALVAGED DATE	WELD OR SALVAGED CSTGS.	SHIPPING RECORD DATE	SHIPPING RECORD RECEIVED NO.	SHIPPING RECORD CSTGS.	SHIPPING RECORD WEIGHT	BALANCE DUE CUSTOMER	ORDER PRICED AT	PER LB. OR PIECE

CUSTOMER'S SCRAP

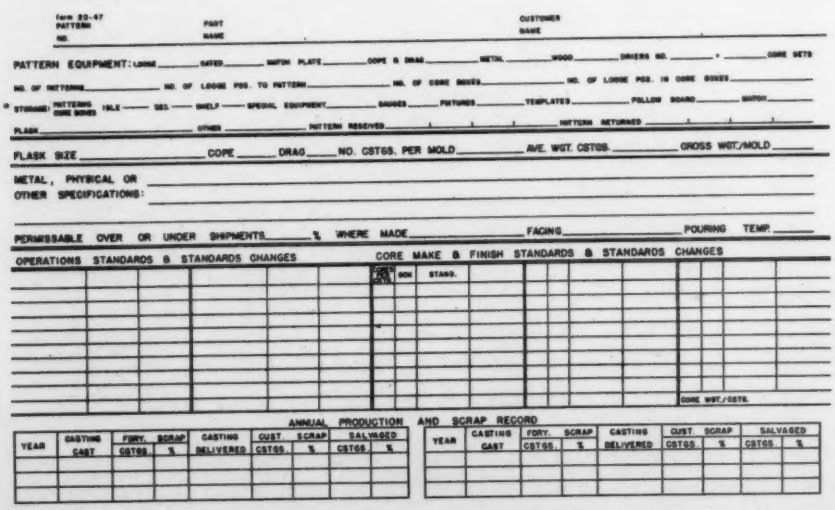
DATE	CSTGS.	WEIGHT	CAUSE	SALVAGE	DATE	CSTGS.	WEIGHT	CAUSE	SALVAGE	DATE	CSTGS.	WEIGHT	CAUSE	SALVAGE

OVER RUN HELD

DATE	CSTGS.	DATE	CSTGS.

PATTERN NO. | CUSTOMER | SEE SPECIF. | SPEC. ALLOY | TEST BAR | NO SCRAP | DELIVERY REQUESTED 1 2 3 4 5 6 7 8 9 10 11 12 | OVER RUN HELD

Fig. 7a — One side of a production control card which is maintained in the office of the person responsible for the control of production.


form 20-47 | PATTERN NO. | PART NAME | CUSTOMER NAME

PATTERN EQUIPMENT: LOOSE ____ GATED ____ MATCH PLATE ____ COPE & DRAG ____ METAL ____ WOOD ____ DRIERS NO. ____ * ____ CORE SETS

NO. OF PATTERNS ____ NO. OF LOOSE PCS. TO PATTERN ____ NO. OF CORE BOXES ____ NO. OF LOOSE PCS. IN CORE BOXES ____

* STORAGE: PATTERNS ____ CORE BOXES ____ FILE ____ BES ____ SHELF ____ SPECIAL EQUIPMENT ____ GAGGES ____ FIXTURES ____ TEMPLATES ____ FOLLOW BOARD ____ MATCH ____

FLASH ____ OTHER ____ PATTERN RECEIVED ____ / ____ / ____ PATTERN RETURNED ____ / ____ / ____

FLASK SIZE ____ COPE ____ DRAG ____ NO. CSTGS. PER MOLD ____ AVE. WGT. CSTGS. ____ GROSS WGT./MOLD ____

METAL, PHYSICAL OR OTHER SPECIFICATIONS: ____

PERMISSABLE OVER OR UNDER SHIPMENTS ____ % WHERE MADE ____ FACING ____ POURING TEMP. ____

OPERATIONS STANDARDS & STANDARDS CHANGES

CORE MAKE & FINISH STANDARDS & STANDARDS CHANGES

CORES PER MOLD	BOX	STAND.						

CORE WGT./CSTG.

ANNUAL PRODUCTION AND SCRAP RECORD

YEAR	CASTING CAST	FORY. SCRAP CSTGS.	FORY. SCRAP %	CASTING DELIVERED	CUST. SCRAP CSTGS.	CUST. SCRAP %	SALVAGED CSTGS.	SALVAGED %

YEAR	CASTING CAST	FORY. SCRAP CSTGS.	FORY. SCRAP %	CASTING DELIVERED	CUST. SCRAP CSTGS.	CUST. SCRAP %	SALVAGED CSTGS.	SALVAGED %

Fig. 7b — Reverse side of card shown in Fig. 7a.

duction runs, will the production control system require revisions. Since most small foundries are job shop operations, the daily operation of the system will not vary greatly. Once set up, with a scheme whereby all information pertaining to delivery and production is simplified and centralized, this phase of the cost control system is more likely than any other to run itself.

CONCLUSIONS

Since the small foundry exists in a highly competitive field, it is obvious that its need for effective cost control is just as great as that of the large foundry with hundreds of employees.

For practical purposes the basic problems of industrial engineering remain the same, whether applied to large or to small companies. The difference lies initially in the fact that the small foundry is not equipped with an experienced industrial engineering department. This makes it advisable to bring in outside help on a consulting basis for the purpose of ascertaining facts and actual conditions, of devising a specific cost control plan, of training selected foundry personnel and of formulating basic standard data.

Thereafter, maintenance of the system which has been developed is carried on by trained foundry personnel. Such a cost control system, if maintained scrupulously, will not wear out with the passage of a few years and the coming of changed conditions. It will adapt itself and constantly keep pace to new methods and situations, with the result that costs will remain under control, improved methods or equipment will result in actual savings, and pricing will continually remain highly competitive on the most realistic basis possible.

HOW TO AVOID SAND SEGREGATION

Based on work done jointly by Committee 8-F of the AFS Sand Div. and the Grading & Fineness Committee of the National Industrial Sand Assoc.

ABSTRACT

The problem of sand segregation was studied, with particular reference to the loading and unloading. Sands colored to indicate fineness were used in the tests. Recommendations based on the results of these tests are given.

INTRODUCTION

Segregation of foundry sands is not a new problem with foundrymen. It has been a constant problem in foundries and with sand producers. However, the problem has become more evident because of vast improvements in sand technology and a greater emphasis on sand control. This added emphasis on sand technology and quality control of castings had led to a greater realization of the importance of sand distribution and grading in foundry operations. Sand segregation problems become more important, since variation in grain distribution ultimately leads to variation in casting quality.

The Joint Committee undertook the study of sand segregation to learn where segregation asserts itself to the greatest extent, and to investigate possible means of reducing it to a minimum.

Sand segregates every time it is moved from one place to another regardless of the method of transportation—truck, rail, wheelbarrow, screw conveyor, belt conveyor, bucket elevators or any other known method. Segregation is most extensive when sand is placed in storage or removed from storage, both at the point of production and at the point of consumption.

One of the participating companies gathered experimental data to demonstrate segregation and indicate one possible method for its control. The work involved the study of a dry sand being loaded and unloaded from a storage silo. The sieve analysis of the sand is shown in the basic test column Table 1.

For the purposes of visually demonstrating segregation the base sand was divided into three fractions: that which was retained on a U.S. Series Sieve No. 40 and coarser was dyed red; from 50 to 100, no dye; and 140 and finer, dyed blue. The three sand fractions were thoroughly mixed and loaded into the experimental square silo made of plate glass with a plexiglass bottom (Fig. 1).

Four experiments were run, with the sand being loaded as indicated in Fig. 2 and subsequently withdrawn. The *X* indicates the point at which the sand was loaded into the top of the silo, and the *O* indicates the discharge point on the bottom.

TEST PROCEDURE

In the first experiment the sand was introduced at the center of the top of the silo through a 1-in. diameter opening. The sand was discharged from the bottom of the tin through sixteen ⅜-in. diameter openings. In each of the other three experiments, the sand was introduced through 1-in. diameter inlets and discharged through 1-in. diameter outlets, as shown in 2, 3 and 4 of Fig. 2.

Colored motion pictures were taken while the silo was being emptied. The dyed sand clearly showed the segregation which takes place when dry sand is handled in each of the four ways. Segregation of sand particles always occurs in the same manner. The coarsest particles always roll to the outside and the finer particles remain in the center Fig. 3.

A complete sieve analysis for tests 1 and 4 is shown in Tables 1 and 2. The numbers at the heads of the columns locate the point at which the top level of the sand was located at the time each sample was taken. The white marks on right-hand silo brace in Fig. 1 are numbered from zero to nine, starting at the top, graduated 6-in. apart.

In the four tests run, the same well-mixed sand sample with an AFS Grain fineness number of 69.71 was used. In test 1, the AFS Grain fineness number

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Members of the AFS Grading, Fineness and Distribution Committee (8-F) who conducted this work are, *Chairman* T. W. Seaton, American Silica Sand Co., Inc., Ottawa, Ill.; *Vice-Chairman* J. G. Smillie, John Deere & Co., Materials Engineering Dept., Moline, Ill.; R. W. Bennett, Walter Gerlinger, Inc., Milwaukee; W. D. Chadwick, Manley Sand Co., Rockton, Ill.; L. F. Dennie, Technical Service, Detroit; George DiSylvestro, American Colloid Co., Skokie, Ill.; H. E. Donnocker, Ottawa Silica Co., Ottawa, Ill.; E. M. Durstine, Keener Sand and Clay Co., Columbus, Ohio; H. S. Fagan, The Fahralloy Co., Harvey, Ill.; F. P. Goettman, Standard Sand Co., Grand Haven, Mich.; L. D. Marinelli, Kensington Steel Co., Chicago; R. E. Morey, Metallurgy Div., Naval Research Laboratory, Washington, D. C.; George Nestor, National Malleable & Steel Castings Co., Cleveland; J. S. Schumacher, The Hill & Griffith Co., Cincinnati; J. A. Schumann, Carpenter Bros., Inc., Milwaukee; E. W. Smith, Foundry Materials, Chicago; N. J. Stickney, Sand Products Corp., Cleveland; T. A. Tarquinio, American Manganese Steel Div., Chicago; L. E. Taylor, Ottawa Silica Co., Ottawa, Ill.; Stanton Walker, National Industrial Sand Association, Washington, D. C.; S. A. Wick, New Jersey Silica Sand Co., Milville, N. J.



Fig. 1 — Dry sand loading and unloading was studied in an experimental square silo made of glass. The white marks on the right-hand silo brace are numbered from zero to nine starting at the top and 6-in. apart.

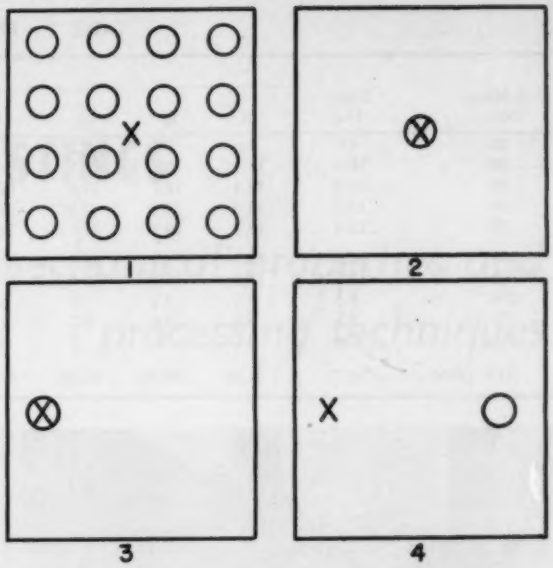


Fig. 2 — Four experiments were conducted with sand being loaded as indicated. The *X* indicates the point at which the sand was loaded into the top of the silo, and the 0 indicates the discharge point on the bottom of the silo.

of samples drawn from the storage bin varied from a low of 66.19 to a high of 72.32. In test 2 this spread ranged from 61.74 to 75.61. Test 3 showed a spread from 61.38 to 84.91, and test 4 varied from 51.98 to 81.65. The marked superiority of the first technique is evident.

These experiments demonstrated that the practice of using several or a large number of openings when withdrawing sand from storage is one of the best developed so far to minimize segregation. This practice should be encouraged as much as possible throughout the industry. The variation in sieve tests taken at various stages of unloading the test silo clearly indicate that segregation had been kept to a minimum.

It is safe to assume that these variations will occur in the sand when withdrawn from large storage areas. These measurements show that sand castings would be made under varied conditions in spite of the fact that the base analysis of the sand is satisfactory. Casting finish and other variables will vary rather substantially.

Sand producers and foundrymen should make every possible effort to draw sand from storage using as many openings at the base of their bins or silos as possible. These openings should be distributed uniformly over the complete base of the silo or bin. In addition, the orifice-type discharge is being used successfully, and should also be considered.

RECOMMENDATIONS

The following recommendations are offered to producers and consumers as methods of further minimizing segregation.

Unloading

Wheelbarrows, endloaders and grab buckets. When unloading sand from cars, work across the car perpendicular to the track in order to get as complete a cross-section of the sand in the car as possible. This

TABLE 1 — RESULTS OF TEST 1

U. S. Sieve No.	Basic Test	Retained, %								
		1	2	3	4	5	6	7	8	9
20	1.0	0.7	1.0	0.9	1.2	0.9	1.0	1.0	1.0	1.0
30	11.4	11.4	10.7	10.7	12.3	11.5	12.5	12.6	10.3	10.3
40	14.3	15.2	13.9	13.7	15.6	14.4	15.9	15.1	12.1	12.5
50	11.4	11.3	11.1	10.5	11.0	10.0	10.6	10.0	13.2	15.3
70	14.9	14.8	14.5	15.1	15.0	14.3	13.8	13.9	15.2	17.4
100	14.9	14.9	15.4	15.9	15.8	14.2	13.6	13.0	15.6	15.7
140	18.5	17.7	19.0	19.0	17.7	19.4	18.9	18.9	19.0	17.0
200	9.5	9.6	10.3	9.7	8.5	10.3	10.1	10.5	9.7	7.4
270	3.5	3.8	3.4	3.9	2.4	4.3	3.1	4.3	3.3	2.9
Pan	0.6	0.6	0.7	0.6	0.5	0.7	0.5	0.7	0.6	0.5
AFS GFN	69.71	69.80	71.60	70.40	66.19	72.32	68.77	72.00	70.57	66.38

TABLE 2 — RESULTS OF TEST 4

U. S. Sieve No.	Basic Test	Retained, %										
		0	¼	½	1	2	3	4	5	6	7	8
20	1.0	1.6	1.6	1.7	1.6	0.5	0.7	1.1	1.0	1.1	0.7	1.7
30	11.4	17.1	15.2	15.5	15.7	7.0	8.6	12.5	12.2	13.6	9.7	18.3
40	14.3	19.4	18.7	17.4	17.5	9.4	11.6	14.5	15.4	15.9	14.4	18.7
50	11.4	14.5	15.0	13.8	13.3	10.2	11.4	11.5	12.6	11.7	12.8	11.9
70	14.9	17.7	16.8	17.7	14.9	15.0	14.5	14.8	14.8	14.4	15.9	13.8
100	14.9	13.5	14.2	14.6	14.3	16.4	15.9	14.0	13.7	13.7	15.1	11.8
140	18.5	11.8	13.7	13.9	15.3	22.3	20.8	17.5	17.5	17.3	18.4	13.9
200	9.5	3.7	4.2	4.7	6.1	12.3	11.7	9.3	9.1	8.7	9.6	7.0
270	3.5	0.6	0.6	0.7	1.2	5.9	4.0	4.1	3.1	3.2	2.9	2.6
Pan	0.6	0.1	T	T	0.1	1.0	0.8	0.7	0.6	0.4	0.5	0.3
AFS GFN	69.71	51.98	53.93	54.96	57.87	81.65	75.79	69.58	67.43	66.15	69.11	59.16



Fig. 3 — Segregation of sand particles always occurs in the same manner. The coarsest particles always roll to the outside, and the finer particles remain in the center.

will assist in reblending the sand and counteract segregation as it is placed into storage.

Bottom dump equipment. Segregation is nearly impossible to control at the point of discharge, and reblending must be accomplished from this point on to storage.

Conveying and Storing

Belts. Distance of fall onto and from the belt should be as short as possible. Wider belts running at slower speeds are better than narrow belts at high speeds for reducing peaking of sand on the belt. The use of baffles and cones at the point of discharge helps distribute sand uniformly throughout the storage area.

Bucket elevators. The use of baffles at the discharge point is important.

Pneumatic conveying. Information on this method is too limited at this time to make any recommendations.

Gravity flow. Baffles should be used to retard the speed of sand flow and minimize the distance of free fall.

Screw Conveyors. Segregation is kept to a minimum with this method of conveying.

Storing. When discharging into storage, endeavor to distribute sand uniformly throughout the storage area.

CONCLUSIONS

The following conclusions were drawn as a result of this study:

1) Sands of high clay content will segregate less than those which are nearly or entirely clay free.
2) Dry sands show a much greater tendency toward segregation than damp sands.
3) Sands of wide grain-size distribution will segregate more than those of narrow distribution. This in no way condemns the use of broadly distributed sands, but rather is intended to point out where segregation occurs to the greatest extent.
4) Sand tends to segregate every time it is handled.

This study has shown the segregation effect of sand on its distribution, and has indicated the sensitivity of dry, low clay containing sands to segregation. The importance of controlling segregation cannot be overemphasized whether it be at the point of production or the point of consumption.

Recommendations have been made by this committee for methods of keeping segregation to a minimum, and it is hoped that these recommendations will be useful to foundrymen and sand producers everywhere.

HIGH STRENGTH STEEL CASTINGS

mechanical properties and processing techniques

By K. D. Holmes, J. Zotos and P. J. Ahearn

ABSTRACT

Mechanical property evaluation was made on high strength test coupons and castings manufactured under various melting and heat treating conditions. Properties of arc furnace and induction melted material were compared, and the effectiveness of certain heat treatments were assessed. Best combination of strength, ductility and toughness was obtained in thin sections of low sulfur and phosphorus steel cast from arc furnace heats.

INTRODUCTION

In recent years the desire for lightweight equipment has sparked considerable interest in the development of high strength cast steels having ductility and toughness comparable to Ordnance steels now employed at lower strength levels. At the present time cast steels in the 150,000 psi yield strength range are quite practical especially in thin sections, while higher strengths than that quoted should be considered as developmental. Every effort is being made to increase the practical yield strength level to the 200,000 psi range, while still maintaining ductility and toughness in the final product.

It is recognized that the inherent cleanliness and chemical composition of steel as poured from the ladle has a considerable effect on the subsequent performance of that steel during mechanical testing. Accordingly, great emphasis has been placed on the development of arc furnace and induction melting practices that produce clean steels of the requisite chemical analysis.

In addition, heat treatment variables have been studied to maximize the formation of low temperature transformation products during the hardening treatment. A third area of investigation, the solidification of high strength steels, is being conducted under contract with a private institution. Results[1] of this work are being incorporated into current Watertown Arsenal practice for manufacturing high strength steel castings.

This paper outlines the current high strength steel manufacturing procedure, and presents a philosophy of manufacture consonant with the attainment of high strength and ductility in the finished product. The mechanical property results submitted for examination are in the strength range between 140,000 and 270,000 tensile strength, and are representative of various size test blocks and actual development castings.

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PROCEDURE

Refractories and Raw Materials

In the melting and casting of high strength steels the constituents of the furnace charge and the refractories employed in and around the furnace must be selected with considerable care, for many potential problems can be eliminated at the very start by proper selection. The furnace, ladle and pouring pit refractories are magnesia brick or magnesia ramming mix. Magnesia possesses a desirable combination of refractory properties such as high melting point, chemical inertness to basic slags and good spalling and erosion resistance which minimizes the presence of mechanically entrapped inclusions in the steel.

The constituents of the furnace charge are somewhat dependent upon the type of melting equipment employed. Induction furnace charges consist of a low carbon steel bar stock with a specified, maximum sulfur and phosphorus of 0.025 per cent, but in general the percentage is in the 0.015 – 0.020 per cent range. The pig iron for the carbon addition and the alloys are also purchased in accordance with this maximum tramp element requirement. The charge for an arc furnace heat is always first quality heavy scrap. The same alloys as those employed in the induction furnaces are used along with slag making materials such as limestone, fluorspar and iron oxide.

All inputs are dry and relatively free of rust, scale and oil. This same precaution is also taken with melters tools, such a skimmers and rabbles, that come in contact with the bath. Great emphasis is placed on the necessity for clean, moisture free inputs which minimize absorption of hydrogen by the steel. Thus, the mechanical properties of the castings produced are not degraded by the presence of hydrogen.[2]

Melting Practice

The induction practice is primarily a remelting and alloying operation; extensive refining of the steel is impractical. The bar stock is melted with the inert alloys molybdenum and nickel. Pig iron and ferrochromium are then added, and a mild carbon boil is obtained which tends to flush out any hydrogen absorbed during the meltdown. The bath is deoxidized with ferromanganese and ferrosilican additions, superheated and tapped; final deoxidation with aluminum is conducted in the metal stream.

On the other hand, the arc furnace practice is designed to take advantage of every practical means known to reduce the sulfur and phosporus content to extremely low levels, and to produce a steel low in gas content and free of nonmetallic inclusions. This is accomplished using a double slag basic practice[3] similar to that employed in the production of steels for forging. The highly basic oxidizing slag employed during the boil period reduces the phosphorus and hydrogen contents to low levels. This slag is removed, and deoxidation of the bath is achieved by the addition of carbon, ferromanganese and ferrosilicon.

The reducing slag is then developed to eliminate sulfur, and to maintain static conditions in the bath that allow accurate chemical control so that final analysis will be to specifications. Final deoxidation is conducted with aluminum in the ladle, and pouring temperature is accurately controlled through the use of high temperature thermocouples.

Pouring Practice

The metal is teemed from a bottom-pour ladle into suitable skin-dried green sand molds that have been gated and risered for complete soundness. Figure 1 shows the pattern, rigging and drum cradle casting. The molding mix is an 80 AFS sand mulled with 4¼ per cent Western bentonite, 3.5-4 per cent water and 1 per cent cereal. Suitable insulation, usually rice hulls, is placed on the riser surface, and the castings are allowed to cool in the molds to at least 600 F before shakeout.

Heat Treatment

The heat treatment is designed to maximize the development of low temperature transformation products in the steel. This is accomplished through a high temperature homogenize and normalize, followed by an austenitize, quench and double temper to the desired strength level. Certain precautions are taken to insure that the above objective is met. For example:

1) High temperature scale formed during the homogenize and normalize is removed by shot blasting prior to the quench, for this scale could reduce quench severity.
2) Carbon content of the steel is as low as possible in order to reduce susceptibility to quench cracking, and also to obtain optimum combination of strength and ductility.[4]
3) The austentizing temperature is within the standard temperature range for the subject steel, but before quenching the furnace temperature is reduced to a point approximately 50 F above the Ar_3 transition temperature.[5] This procedure reduces the amount of heat that is ordinarily removed during the quench, and thus the effective quench severity is increased without causing drastic thermal gradients that lead to quench cracks in the casting.
4) The quenching medium is usually water. When necessary, an interrupted quenching procedure, first in water and then in oil, is employed in preference to a straight oil quench.
5) Hardenability of the steel is more than adequate for the section size being treated. Retained austenite, usually associated with excess hardenability, is eliminated through the double temper treatment.
6) The tempering temperature is selected to avoid any embrittlement range, and the steel is often quenched from the temper for the same reason.

It is realized that the benefit derived through some of these precautions may be marginal, but because of the developmental nature of the program such marginal improvements are justified in view of the desire for maximization of mechanical properties.

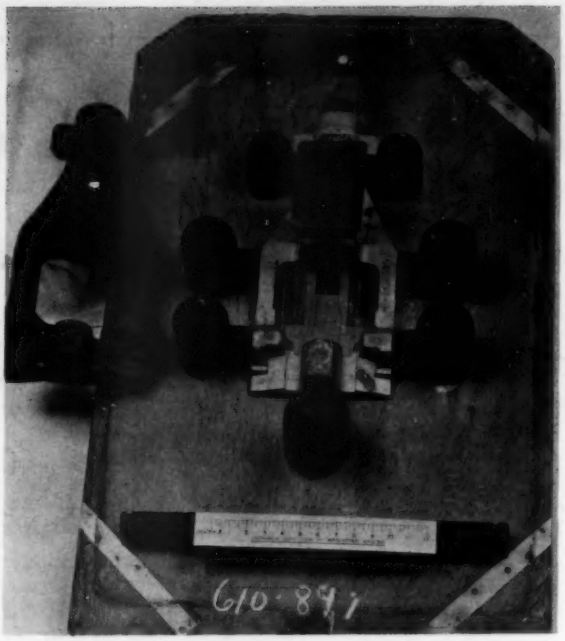


Fig. 1 — Pattern, rigging and drum cradle casting.

Testing

A sketch of the coupon block for mechanical testing is shown in Fig. 2. Design of coupon allows the researcher to test in thin sections (fins), and also at the midwall and center of a massive section. Thus, he can study mass or section size effect in relation to mechanical properties. Results of coupon testing is always confirmed by tests conducted on actual shape castings whenever such is involved.

The mechanical property evaluation is made on 0.357 in. diameter threaded tensile bars and 0.394 in. square V-notch Charpy bars that have been removed from suitable casting locations.

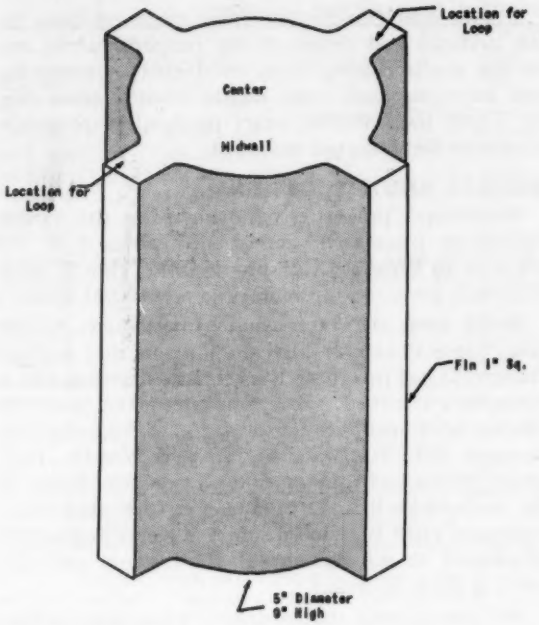


Fig. 2 — Sketch of coupon block.

The steels generally employed do not have a well defined ductile to brittle transition over a narrow temperature band, rather it is spread out over a considerable temperature range. The lower part of the transition occurs at between –20 and –40 F. At these testing temperatures, fibrosity of the Charpy fracture should be between 10-15 per cent. In the preparation of these test bars final grinding to finish dimensions is carefully controlled, for high strength steels will develop grinding cracks. This danger is easily eliminated by maintaining adequate coolant flow during this operation.

Mechanical test information is often supplemented by metallographic inspection for structure and non-metallic inclusions, and macroexamination for overall quality, cleanliness and soundness. Such additional information can be most valuable in predicting steel performance prior to mechanical testing, or can be used to ascertain reasons for premature failure of specific test bars.

Steel Chemistry

The data presented herein are confined to two basic analyses, the first one being a chromium, nickel, molybdenum steel that has been used extensively in the manufacture of heavy-walled tubes and massive castings. The basic analysis of this steel was modified slightly with respect to carbon for the high strength program. The second analysis, a S.A.E. 4340 modified with 1.5 per cent silicon, was selected for its ability to resist temper embrittlement. The exact chemistries for each heat of the program appear in Table 1.

TABLE 1 — HIGH STRENGTH STEEL COMPOSITIONS

Cr-Ni-Mo Steels	Compositions, %									
	C	Mn	Si	S	P	Ni	Cr	Mo	Al	V
Arc Furnace										
Arc - 1	0.305	0.83	0.30	0.009	0.009	2.35	0.76	0.84	0.045	
Arc - 2	0.245	0.87	0.42	0.008	0.009	2.48	0.74	0.44	0.055	
Arc - 3	0.295	0.82	0.52	0.009	0.009	2.48	0.76	0.42	0.050	
Induction Furnace										
Ind - 1	0.38	0.66	0.30	0.027	0.013	2.31	0.87	0.43	0.041	0.15
Ind - 2	0.33	0.64	0.34	—	—	2.12	0.89	0.43	—	0.125
Silicon Modified Induction Furnace										
Ind - 3	0.37	1.46	1.64	0.022	0.017	1.30	—	0.31	0.020	

These two analyses were employed in the manufacture of finned coupons from both arc furnace and induction melts. In addition, the chromium, nickel, molybdenum analysis was used for the production of high strength drum cradle castings.

The specific processing procedure for each type of casting is given in the schedules appearing as Tables 2 and 3. Table 2 presents necessary processing information pertinent to finned coupons cast from the chromium, nickel, molybdenum arc furnace steel, while Table 3 gives the same information for induction steels of the chromium, nickel, molybdenum and silicon modified analysis cast into finned molds as well as procedure for the cradle castings.

Comparison of these tables will point out certain planned differences in practice designed to allow the investigator an opportunity for correlation of these variations with resulting mechanical properties. For example, steels (Table 2) were manufactured in the arc furnace, and then cast and heat treated in such a way as to eliminate any tendency for quench cracking by employing a slack quench. It is believed the major variation is represented by the interrupted quench and the double slag basic melting and refining practice. As has been mentioned, this practice yields clean steel unusually low in sulfur and phosphorus content.

On the other hand, induction melted steels (Table

TABLE 2 — PROCESSING SEQUENCE FOR ARC FURNACE Cr-Ni-Mo CASTINGS

Heat No.	Casting Type	Treatment Schedule
		Low Strength Treatment *Normalize:* 1900 F, 16 hr; air cool (remove gates and risers, shot blast, weld quenching loops). *Harden:* 1650 F, 6 hr; furnace cool to 1500 F, 2 hr; H_2O quench, 5½ min; air, 10 min; oil, 15 min. *Temper:* 600 F, 2 hr; raise to 1130 F, 6 hr; oil quench, 15 min. *Soak:* 600 F, 12 hr.
Cr-Ni-Mo Steels Arc - 1 Arc - 2 Arc - 3	Finned Coupons	
		or High Strength Treatment *Homogenize:* 2200 F, 5 hr; air cool. *Normalize:* 1750 F, 3 hr; air cool. *Anneal:* 1225 F, 4 hr; furnace cool (remove gates and risers, shot blast, weld quenching loops). *Harden:* 1600 F, 3 hr; furnace cool to 1350 F, 2 hr; H_2O quench, 1½ min; air, 5 min; oil, 1 min. *Double Temper:* 400 F, 6 hr; air cool.

TABLE 3 — PROCESSING SEQUENCE FOR INDUCTION FURNACE HIGH STRENGTH CASTINGS

Heat No.	Casting Type	Treatment Schedule
Cr-Ni-Mo Steels		
Ind - 1	Finned Coupons	*Homogenize:* 2200 F, 8 hr; air cool. *Normalize:* 1750 F, 8 hr; air cool. *Anneal:* 1225 F, 12 hr; furnace cool (remove gates and risers, shot blast, weld quenching loops). *Harden:* 1600 F, 4 hr; furnace cool to 1325 F, 2 hr; H_2O quench cold. or *Double Temper* at 300 F, 16 hr; H_2O quench from temper.
Ind - 2	Drum Cradles	*Homogenize:* 2200 F, 5 hr; air cool. *Normalize:* 1750 F, 3 hr; air cool. *Anneal:* 1225 F, 4 hr; furnace cool (remove gates and risers — shot blast). *Harden:* 1600 F, 3 hr; furnace cool to 1500 F, 2 hr; oil quench. *Double Temper:* 775 F, 4 hr; water quench. or *Double Temper:* 825 F, 4 hr; water quench. or *Double Temper:* 1175 F, 4 hr; water quench. or *Double Temper:* 1200 F, 4 hr; water quench.
Silicon Modified Steel		
Ind - 3	Finned Coupons	*Homogenize:* 2200 F, 8 hr; air cool. *Normalize:* 1750 F, 8 hr; air cool. *Anneal:* 1225 F, 12 hr, furnace cool (remove gates and risers, shot blast, weld quenching loops). *Harden:* 1600 F, 4 hr; furnace cool to 1325 F, 2 hr; H_2O quench. *Double Temper:* 600 F, 16 hr; H_2O quench from temper. or *Double Temper* at 300 F, 16 hr; H_2O quench from temper.

3) were cast and heat treated to optimize the development of maximum properties through the heat treatment cycle. Coupons were water quenched cold so that massive sections would have the opportunity to exhibit properties comparable to fins. The cradle castings were processed in exactly the same way as the coupons, except that an oil quench was required in view of the complex casting geometry and the thin sections involved.

In all cases tempering temperatures were adjusted somewhat in conformance with objectives. Arc furnace coupons were tempered to tensile strength levels of 135,000 and 225,000 psi. Direct comparison could be made of deterioration in ductility, etc., with increased strength. In addition, high strength coupons from the arc furnace practice could be compared with similar coupons (Table 3) from the induction practice. Tempering temperature for the silicon modified steel was as low as possible in order to obtain maximum strength, while tempering treatments for the cradles were varied to obtain properties over a range of strength levels.

Tensile and Charpy bars were removed from the fin, midwall and center of the coupons, while tests on the cradle castings were conducted by sectioning and removing bars from highly stressed areas (Fig. 3). These tests disclose exact mechanical properties of cradles that were not sectioned.

RESULTS AND DISCUSSIONS

Mechanical properties resulting from the various processing procedures appear in Tables 4 - 8. Table 4 is an outgrowth of procedural Table 2, while Tables 5 - 8 are complementary to procedural Table 3.

Steels from the arc furnace (mechanical properties, Table 4) exhibit high ductility in thin sections. At the 135,000 psi strength level little deterioration in properties is noted when tests are taken from the thicker sections. This is not the case with the high strength 225,000 psi tensile strength results. Here, there is a marked difference in properties between the fin and midwall and, in addition, the yield-tensile strengths ratio is reduced. Such a trend is generally attributed to microstructural constituents associated with a slack quench.[6]

We can surmise that a drastic water quench from the austenitizing temperature would have improved midwall properties. The desirable high strength fin properties are reproducible for the results are from three different arc furnace heats, and each result presented represents the average of four tests. It is inferred that thin-walled castings would have proper-

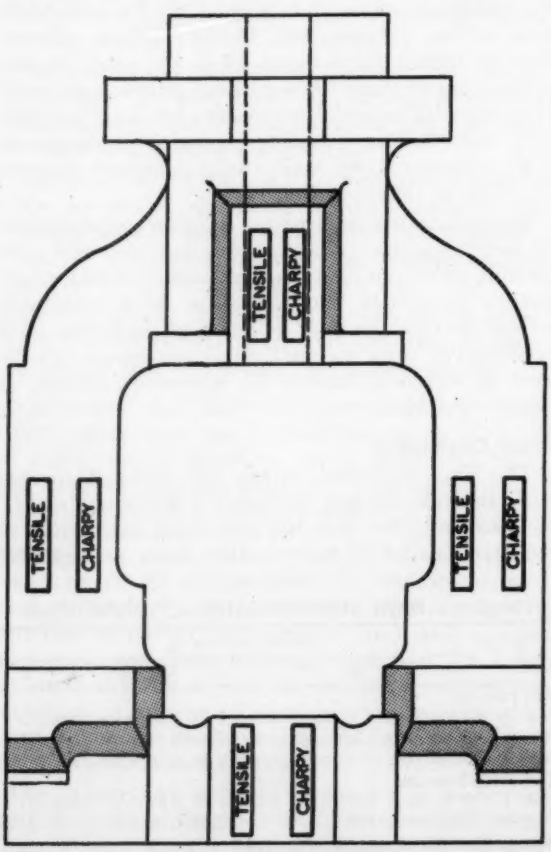


Fig. 3 — Test locations for drum cradle.

TABLE 4 — MECHANICAL PROPERTIES OF Cr-Ni-Mo ARC FURNACE STEELS

Heat No.	Y.S.—0.1%, psi	Y.S.—0.2%, psi	T.S., psi	0.1% Y.S. T.S.	Elong., %	R. A., %	Impact at —20 F, ft-lb	Fibrosity, %
Coupons Heat Treated to 130,000 T.S. Tempered at 1130 F.								
Arc-1-Fin	124,625		139,750	0.89	17.3	44.2	44.0	100
Arc-1-Mid	122,000	123,875	142,500	0.87	15.6	28.2	29.0	100
Arc-2-Fin	114,325		131,000	0.81	20.0	53.8	54.0	100
Arc-2-Mid	114,000	115,375	131,750	0.87	16.4	35.5	37.6	100
Arc-3-Fin	117,625		136,375	0.86	19.3	51.8	49.8	100
Arc-3-Mid	119,500	121,000	137,725	0.87	15.0	31.0	39.2	100
Coupons Heat Treated to 225,000 T.S. Tempered at 400 F.								
Arc-1-Fin	173,875	186,375	229,250	0.76	10.2	38.2	17.4	30
Arc-1-Mid	124,375	143,750	197,875	0.63	1.2	2.4	12.9	14
Arc-2-Fin	171,875	183,625	216,500	0.80	12.0	33.8	13.8	25
Arc-2-Mid	126,250	144,125	193,250	0.65	3.8	9.2	14.3	12
Arc-3-Fin	175,250	188,250	230,500	0.74	8.0	33.6	15.0	24
Arc-3-Mid	123,375	143,500	193,250	0.64	2.3	4.6	11.9	11

TABLE 5 — MECHANICAL PROPERTIES OF Cr-Ni-Mo INDUCTION FURNACE STEEL (Heat No. Ind — 1, Coupons Tempered at 600 F)

Position	Y.S.—0.1%, psi	Y.S.—0.2%, psi	T.S., psi	0.1% Y.S. T.S.	Elong., %	R. A., %
Tensile Testing						
Fin	200,500	204,000	239,000	0.84	6.4	11.4
Fin	202,500	205,000	239,500	0.84	5.0	8.7
Mid-Wall	207,500	212,000	249,000	0.83	5.0	11.4
Mid-Wall	210,000	215,500	251,500	0.84	5.0	1.6
Center	203,500	211,000	230,000	0.88	0.7	1.6
Center	195,500	203,500	217,500	0.90	0.7	3.8

Charpy Impact (Ft-Lb) Data for Transition Curves

Position Temp. C(F)	Fin	Mid-Wall	Center
−40(−40.0)	9.2	6.4	6.4
	10.3		5.7
−20(−4)	7.8		6.4
	9.7		6.2
0(+32)	10.0	6.2	6.7
	11.8		6.7
+20(+68)	9.7	9.5	7.5
		9.5	7.5
+40(+104)	10.6	7.5	8.9
			9.5

ties identical to the fin properties, that is, a tensile strength of 225,000 psi, a 0.1 per cent yield strength of 175,000 psi, 10 per cent elongation, 35 per cent reduction in area and a −40 F impact resistance of 15 ft-lb.

The over-all outstanding combination of mechanical properties in the high strength fins is in large measure due to the inherent cleanliness of the arc furnace steel, and, in particular, to its low sulfur and phosphorus content < 0.010 per cent.[7,8]

Table 5 presents mechanical properties of induction melted steel of the same analysis as the arc furnace steel. The only chemical difference between the two is that the induction melt had a greater carbon content (0.09 per cent C) and also had higher sulfur and phosphorus content (approximately 0.007 per cent). Comparison of the high strength arc furnace fin data to the induction melt data indicates the general superiority of the arc furnace steel. This difference may be attributed to the sulfur and phosphorus differences or to the higher carbon content of the induction material.

Unfortunately, in order to make comparisons at equivalent strength levels, the induction melted steel was tempered at 600 F, and it can be inferred that the use of this tempering temperature caused some embrittlement.[9] This seems to be confirmed by data of Table 6 where the induction steel was tempered at a lower temperature (300 F) to yield a higher strength (270,000 psi) and increased ductility (20 per cent reduction in area).

The 500-600 F embrittlement can be overcome by silicon additions.[10] Table 7 gives results for coupons with the silicon addition and a 600 F temper. Properties are comparable to those of the chromium, nickel and molybdenum analysis in Table 6.

All of the induction steels developed differences in mechanical properties from the fin to midwall to center. In view of the drastic water quench employed, it seems unlikely that microstructural constituents usually associated with slack quenching could account for these differences. The yield-tensile strengths ratio remains fairly constant and indicates that the quench was effective.[6] Possible reasons for the difference between midwall and center properties have been discussed elsewhere.[11]

Interpretation of overall impact results is quite difficult, but some observations can be made. First, toughness decreases with increased strength; also, it decreases in going from the outside to the inside of the castings. On the other hand, the toughness required as insurance against complete brittle failure

TABLE 6 — MECHANICAL PROPERTIES OF Cr-Ni-Mo INDUCTION FURNACE STEEL (Heat No. Ind. — 1. Coupons Tempered at 300 F)

Position	Y.S.—0.1%, psi	Y.S.—0.2%, psi	T.S., psi	0.1% Y.S. T.S.	Elong., %	R. A., %
Tensile Testing						
Fin	194,500	209,000	274,000	0.72	9.3	22.6
Fin	195,500	209,500	280,000	0.70	8.6	19.1
Mid-Wall	200,000	214,500	283,500	0.70	5.7	9.3
Mid-Wall	197,000	211,500	283,500	0.70	5.0	8.2
Center	186,000	202,000	240,000	0.77	1.4	3.8
Center	182,000	198,000	236,500	0.77	2.9	5.5

Charpy Impact Data for Transition Curves

Position Temp. C(F)	Fin	Mid-Wall	Center
−40(−40.0)	7.0	4.2	4.2
	6.2	3.4	3.4
−20(−4)	7.0	5.4	5.2
	7.3	5.4	5.4
0(+32)	6.7	6.0	4.4
	7.3	5.4	4.7
+20(+68)	8.4	5.7	6.2
		6.2	5.4
+40(+104)	8.6	7.0	6.0
		6.7	5.0

TABLE 7 — MECHANICAL PROPERTIES OF SILICON MODIFIED INDUCTION FURNACE STEEL
(Heat No. Ind. — 3. Coupons Tempered at 600 F)

Position	Y.S.—0.1%, psi	Y.S.—0.2%, psi	T.S., psi	0.1% Y.S. T.S.	Elong., %	R. A., %
Tensile Testing						
Fin	216,500	228,500	270,000	0.80	6.4	16.6
Fin	218,500	231,500	272,000	0.81	5.7	13.5
Mid-Wall	213,000	226,000	266,500	0.80	4.3	9.8
Mid-Wall	217,000	230,000	268,500	0.80	2.1	6.0
Center	*	*	193,500		0.0	0.0
Center	204,000**	*	203,000		0.0	0.0

*Specimen broke before requested offsets obtained.
**Yield point by divider method.

Charpy Impact Data for Transition Curves

Temp. C(F)	Position Fin	Mid-Wall	Center
−40(−40.0)	8.6	5.7	3.2
	6.4	6.0	7.0
−20(−4)	9.5	6.4	6.7
	9.2	2.8	5.7
0(+32)	9.7	7.5	7.0
	11.8	7.3	7.3
+20(+68)	10.0	6.4	6.7
		7.3	8.6
+40(+104)	10.6	7.3	7.0
		9.2	7.8

(flow and fracture stress coincide) must increase with increased yield strength, for ability to absorb elastic energy also increases with the yield strength.

Even though the steels exhibited reasonable percentages of fibrosity in the Charpy test, the above reasoning leads to a conservative approach in the application of high strength steels. Performance tests of actual castings should be required, and then a practical toughness should be specified that is representative of successful test castings.

It is interesting to compare results of these coupon experiments with similar forging data taken in the transverse direction.[6] At low strength levels the equivalence of properties is well known. Comparison of the limited data available indicated the same equivalence seems to hold at high strength also.

The high strength program is not limited to coupon block castings, but also includes the manufacture of useful high strength castings. The data now presented on the drum cradle casting are typical of this. Table 8 shows mechanical properties resulting from various tempering treatments and the cradle manufacturing procedure outlined in Table 3. High tempering temperatures, in the 1200 F range, resulted in a strength level of 160,000 psi with a ductility of 35 per cent reduction in area and a −20 F impact energy of 29 ft-lb, while the processing technique with a low tempering temperature (775 F) yielded a strength of 215,000 psi, ductility of 15 per cent reduction in area and −20 F impact energy of 8.5 ft-lb. The higher strength properties represent a reasonable compromise between ductility, toughness and strength. Since tests were conducted on bars removed from actual castings, it is assumed that mechanical properties of castings submitted for service test are as shown in Table 8.

TABLE 8 — MECHANICAL PROPERTIES OF DRUM CRADLES
(Heat No. Ind. — 2. Average of 3 tests)

Y.S. −0.1%, psi	Y.S. −0.2%, psi	T.S., psi	0.1% Y.S. T.S.	El., %	R.A., %	−20 F Impact ft-lb	Fibrosity, %
A. Casting Tempered at 775 F							
201,333	204,333	217,667	0.92	5.7	15.0	8.5	18
B. Casting Tempered at 825 F							
186,333	189,000	197,000	0.94	3.5	13.0	10.5	23
C. Casting Tempered at 1175 F							
150,750	150,083	157,233	0.96	13.6	33.9	29.7	100
D. Casting Tempered at 1200 F							
152,667	152,500	161,133	0.94	13.3	35.0	28.1	100

SUMMARY

Cast test coupons were manufactured, heat treated and tested from arc furnace and induction melted steels. Data presented show the mechanical properties obtainable through various heat treating and testing procedures. The superiority of arc furnace practice is high lighted by presentation of data from coupon fins wherein a strength level of 225,000 psi is obtained with 35 per cent reduction in area and 15 ft-lb impact at −40 F. Thin-walled, high strength castings were manufactured, tested for mechanical properties and submitted for service testing. Small section size castings are quite feasible, but as section increases mechanical properties are lowered.

REFERENCES

1. *Investigation of Solidification of High Strength Steel Castings Under Simulated Production Conditions,* M.I.T., July 1, 1958, Contract Nr. DA-19-020-505-ORD-4511.
2. "Delayed Failure and Hydrogen Embrittlement in Steel," R. P. Frohmberg, W. J. Barnett and A. R. Troiano. *A.S.M. Transactions,* pp. 892-923, 1955.
3. "Optimum Operating Procedure for Basic Electric Steel Furnace for Desulphurization and Dephosphorization," *Journal of Steel Castings Research,* No. 8, pp. 12-16, May 1957.
4. "Tensile and Impact Properties of Low Carbon Martensite," C. C. Busby, M. F. Hawkes and H. W. Paxton, *A.S.M. Transactions, vol. 47,* pp. 135-156, 1955.
5. "Stepped Austenitizing Treatment for 4340 Steel," E. P. Klier, Volker Weiss and George Sachs, *A.I.M.E. Transactions, vol. 209,* p. 424-425.
6. "Effect of Silicon on Transverse Properties and on Retained Austenite Content of High Strength Steels," John Vajda, John J. Hauser and Cyril Wells, *A.S.M. Transactions, vol. 49,* pp. 517-549. 1957.
7. "Effect of Phosphorus and Sulfur Content on Ductility and Toughness of Cast Low Alloy Steels," J. Zotos, *Electric Furnace Steel Proceedings, A.I.M.E., vol. 15,* 1958.
8. "Low Temperature Impact Properties of Cast Steel," *Journal of the Iron and Steel Institute, vol. 187,* Part II, pp. 104-120, Oct. 1957.
9. "Embrittlement of Alloy Steel at High Strength Levels," L. J. Klinger, W. J. Barnett, R. P. Frohmberg and A. R. Troiano, *A.S.M. Transactions, vol. 46,* pp. 1557-1598, 1954.
10. "Some Effects of Silicon on the Mechanical Properties of High Strength Steels," C. H. Shih, B. L. Averbach and Morris Cohen, *A.S.M. Transactions, vol. 48,* pp. 86-116, 1956.
11. "Mass Effect on Castings Tensile Properties, a rationalization," G. W. Form, P. J. Ahearn and J. F. Wallace, AFS TRANSACTIONS, *vol. 67,* p. 65, 1959.

CORE BOXES FOR SHELL CORES

By J. E. Stock

ABSTRACT

The use of gray iron shell core boxes is discussed, showing reasons why the author's company uses this metal in preference to aluminum. Some of the provisions necessary when using gray iron are given as well as the use record of some of the boxes.

INTRODUCTION

There seems to be little said about core boxes for shell cores that finds its way into printed form to be published for the benefit of those interested in shell cores.

Probably due to the fact that throughout the foundry industry, each individual working on the shell sand method for cores is prone to think things out as he goes along, and as in most cases use the trial and error method of arriving at the solution of the problem or problems.

When we talk normally in terms of a ten year period of active interest in shell core and shell mold processes, we sometimes think that in ten years any intelligent person should know all there is to know about it. However, that is possibly true with many, but it has been found that in the foundry business at the end of a ten year period it is not uncommon to become confused about many issues. This is evident when you read the trade journals and find so many different opinions voiced on any one process or method.

These opinions are based on certain facts, and, therefore, are true when you know all the conditions from which the facts were obtained.

The one outstanding fact is that each one of us here is a little bit different than each other in more than one respect. The same applies to our respective foundries, no two are exactly alike even though the basic fundamentals of foundry practice are still observed.

SHELL CORE EQUIPMENT

Equipment for making shell cores is a subject that is important in many ways. There are those who wish to redesign the foundry to fit the equipment, and there are those who wish to have the equipment designed to fit the foundry. Both have their place depending upon what is best suited to the particular application.

Before you can decide what you want in the type of core box you must first decide what type of machine you are going to use. Of course, you do not need a machine to make shell cores, they also can be made on a bench, simply heat the boxes and take them apart by hand.

METAL USED

When the decision is made to provide a core box for making shell cores it will have decided what type of metal is to be used to fit the needs. Shell core boxes can be made of several metals, iron, steel, bronze, aluminum and possibly some other combinations.

At the author's company gray iron is used for all shell core boxes as well as shell mold patterns, and both are made by the "cast to size" method. The reason is simple, we have the gray iron and what we think is a satisfactory and most economical method of making the core boxes. We are also well satisfied with the results obtained from the use of gray iron core boxes. We have used cast aluminum shell core boxes with varying results.

The main objection to aluminum was the co-efficient of expansion characteristics when subjected to temperatures up to 600 F differ somewhat from those obtained by use of gray iron at the same temperature. If we were to say that the rate of heat transfer in aluminum core boxes is twice as rapid as that of gray iron core boxes then the dissipation rate should follow the same line.

This has an appreciable effect on the cycle time on a given machine to produce a quality shell core whether it is to be drained hollow, or made solid.

ALUMINUM CORE BOXES

Experience has shown that when using aluminum core boxes on several shell core machines, the machines can be operated on shorter cycles than when using gray iron boxes on the same machine, and at higher temperatures. However, on multiple station machines where there are six different sizes and shapes of cores being made at the same time and temperature, not all of the cores were cured exactly the same. Some cores because of their design were

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good, others having several right angle sections were subject to insufficient cure and peel back occurred.

Warpage was encountered in the aluminum shell core boxes after the first 30 days they were on the machines.

The use of aluminum shell core boxes has been discontinued at the author's company, and gray iron boxes are being used. To begin with the core boxes were made solid. Then, in order to provide for better heat transfer to all sections of the core cavity the back was hollowed out, and metal wall sections of a thickness to insure uniform temperature to all sections were provided. In some cases it was necessary to provide for copper inserts at some sections to provide an excess of heat to a given area to properly cure.

CORE BOX EXPANSION PROVISION

It is necessary and of great importance in making the patterns for the core boxes that, in addition to providing for normal shrinkage of the metal, consideration be given to the normal expansion of the metal box when heated, otherwise a core that is larger than desired will be obtained.

Shell core boxes can be designed with a minimum of draft, and in some cases no draft. Proper rigging of shell core boxes is important, especially when used in high production. Proper breaking of shell core boxes can eliminate a lot of lost time and scrap cores.

Venting shell core boxes can present quite a problem in some instances. Wherever possible slotted or screen blow vents should be avoided. A practical vent that works well is to drill a hole in the area to be vented and insert a square pin to provide vent area on four sides. This type of vent cannot fall out as do some of the common blow vents, due to the contraction and expansion occurring when boxes are subjected to variations in temperature.

One of the most desirable methods of venting shell core boxes has been accomplished by milling slots in the parting of the boxes adjacent to the area to be vented. These slots may be of a depth that is suitable to the type of core sand used. They should allow for the air to escape and not the sand. The slots do not necessarily have to be milled to the outside of the box at the parting line, but should be approximately ½-in. outward from the cavity, and a hole of small diameter could be drilled from the slot through the box to the outside. This is a matter of preference.

For removal of cores from the boxes the core boxes should be well treated with a release agent before use, and in many cases the use of an ejection system is necessary.

CORE BOX RIGGING

Proper rigging of core boxes to provide efficient operations for blowing, curing and ejection of cores is of the utmost importance regardless of whether it is for high or low production rates.

Analysis of the gray iron used in the cast to size core boxes at the author's company is as follows:

Carbon.....320 - 340	6 to 9 mm chill depth
Silicon .....220 - 240	on a wedge type speci-
Sulfur 0.15	men 1⅜-in. high by
Phosphorus . .0.20	½-in. base.
Manganese . . .0.65	

All gray iron core boxes are subject to a HT5 stress relief before machining. To date the author's company has approximately 300 gray iron core boxes producing shell cores that have been in use over the past four years. These core boxes have been subjected to heat variations from ambient or room temperature up to 600 F.

Some are in production for 2 or 3 hr, and others have operated for 24 hr for six days without allowing to cool.

Based on past experience all shell core boxes and/or patterns required in the future will be made of gray iron.

TIN EFFECT ON STRUCTURE AND PROPERTIES OF FLAKE AND NODULAR GRAPHITE CAST IRONS

By E. C. Ellwood

ABSTRACT

The paper reviews previous work on the effect of tin on flake and nodular graphite cast irons. New results on the effect of tin on the promotion of pearlite matrices in two nodular irons are presented. The tin effect on the mechanical properties of such irons is discussed.

INTRODUCTION

Over the past few years, work which has been published shows that tin has a marked effect on the structure and properties of cast iron. In particular, it affects the stability of eutectoid cementite and promotes the formation of pearlite in the matrix of cast iron. This latter effect is not accompanied by any great risk of the formation of proeutectoid cementite, so that by tin additions a completely pearlitic matrix may be obtained without risk of the formation of hard spots.

PREVIOUS WORK

Tin in Flake Graphite Cast Irons

De Sy and Foulon[1] carried out a limited investigation on the effect of tin on the matrix of flake graphite iron as a corollary on their investigation of tin in nodular iron. The iron chosen contained nearly 4 per cent carbon, and approximately 2 per cent silicon, in a substantially phosphorus-free base. Without treatment, the iron solidified with coarse graphite flakes surrounded by sheaths of ferrite, which comprised about 20 per cent of the volume of the matrix.

With 0.020 per cent residual tin, traces of free ferrite remained, and with 0.048 per cent tin, a completely pearlitic matrix was obtained. This work established that tin is a pearlite promoter in gray iron. However, before its use for this purpose could be established, information on the mechanical properties of iron containing small amounts of tin was required. In an investigation sponsored by the Tin Research Institute at Battelle Memorial Institute, Davis, Krause and Lownie[2] studied the effect of tin in cast iron in rather more detail, paying particular attention to its effect on the mechanical properties.

Their experiments were carried out on an automotive-type hypoeutectic, and a general-purpose hypereutectic gray iron. The Brinell hardness for both irons increased with increasing amounts of tin up to 0.5 per cent. The tensile strength in 1.2-in. diameter bars increased with tin content up to about 0.2 per cent, and levelled off as more tin was added. The transverse strength of the two irons was affected differently by tin.

The transverse strength of the general-purpose irons was markedly increased by tin contents of 0.1 and 0.2 per cent, while that of the automotive-type iron was unaffected. The deflection in the transverse test, and the impact strength of unmachined 1.2-in. diameter bars, increased slightly when about 0.05 per cent tin was added, and then fell off with increasing amounts of tin.

The effect of tin upon the stress-strain relationship in the two types of cast iron was also studied. It was found that the modulus of elasticity for the automotive-type iron was increased by about 5 per cent. For the soft hypoeutectic iron, 0.08 per cent of tin reduced the modulus of elasticity slightly (but not consistently), and increased the tensile strength by about 16 per cent. In both cases, the effect on the stress-strain curve was small from a practical point of view, but it is interesting that tin was beneficial rather than harmful.

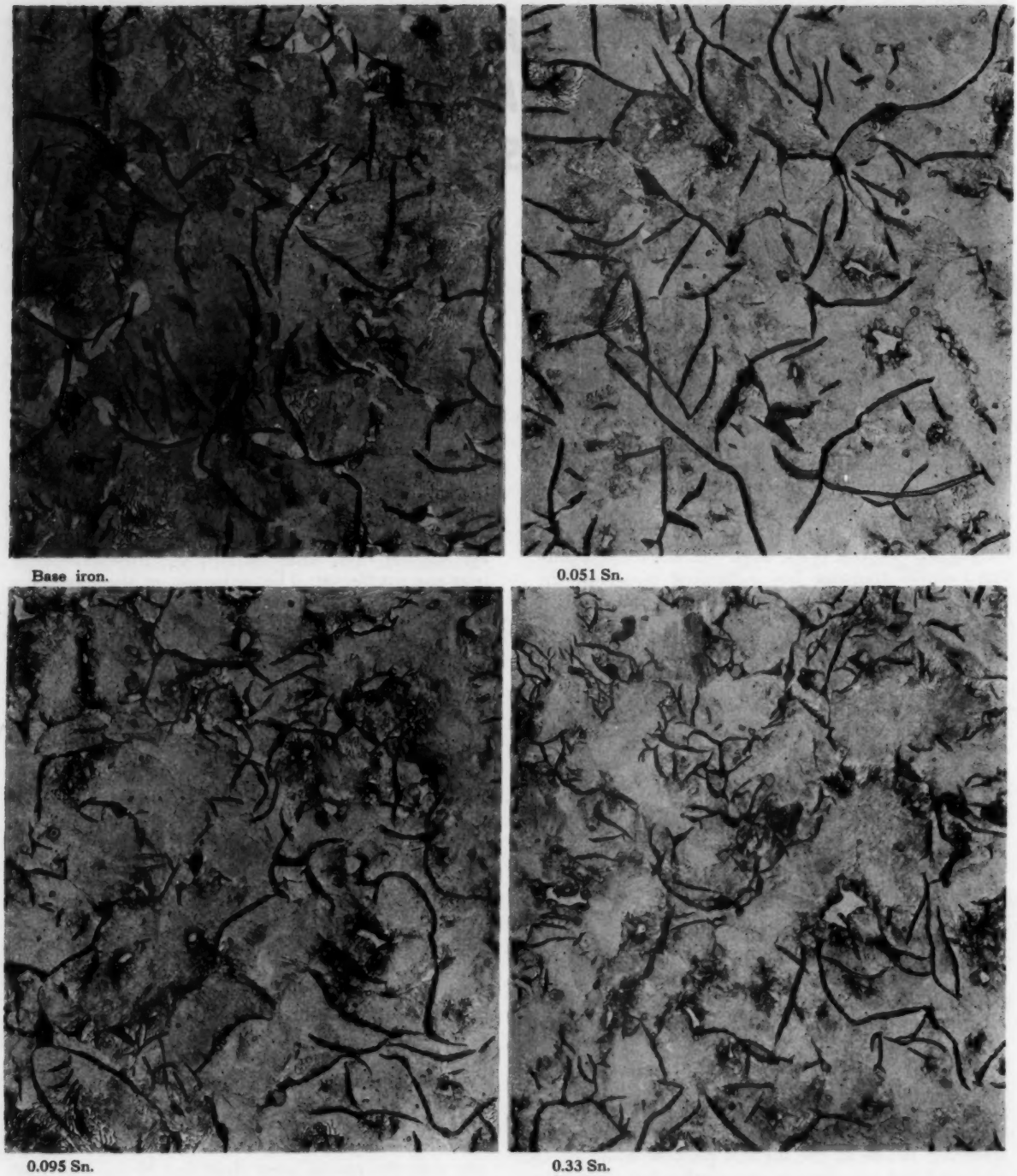
Impact Strength Test

The effect of tin on the impact strength of the irons was investigated in more detail. Test bars of 1.5, 1.125 and 0.75 in. diameter were machined from cast bars of 2.0, 1.2 and 0.875 in. diameter, respectively, and were tested for impact strength in the unnotched condition in a Charpy-type test. The impact strength increased slightly with small additions of tin. In both types of iron, and in the three sizes of bars tested, the impact strength of the iron containing 0.1 per cent tin was no lower than that of the base iron.

At the same time, tin had no deleterious effect on the impact strength until after massive ferrite was substantially eliminated from the microstructure. Therefore, tin was shown to be embrittling only in amounts exceeding those required to obtain the desired microstructure.

This work showed that for the automotive hypoeutectic iron, the percentage of pearlite in the matrix increased from about 80 per cent with no tin, to almost 100 per cent with 0.051 per cent tin (Fig. 1). With the hypereutectic iron, about 0.2 per cent tin was required to obtain a substantially pearlitic ma-

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Base iron. 0.051 Sn.

0.095 Sn. 0.33 Sn.

Fig. 1 — Microstructures of hypoeutectic cast iron in 1.2 in. diameter bars. Base iron composition — 3.33 C; 2.08 Si; 0.64 Mn; 0.108 P; 0.096 S. 250×.

trix, but in this case the base iron contained about 55 per cent ferrite (Fig. 2). These results were obtained on test bars 1.2 in. in diameter, but it was shown that similar amounts of ferrite and pearlite were obtained in test bars 0.875 in. and 2.0 in. in diameter for the same tin content.

It was also shown that tin additions considerably in excess of the optimum amounts necessary to produce a fully-pearlitic matrix did not result in the formation of massive carbides in the microstructure.

The effect of tin upon the amount of pearlite in gray iron is of practical significance because of the known wear resistance of fully-pearlitic iron. Pearlitic structures may also be obtained by the addition of conventional alloying elements such as vanadium or chromium, but these additions have the undesirable feature that small excesses cause massive carbides to appear in the microstructure. In contrast to this, rea-

sonable excess of tin does not promote the formation of massive carbides.

As a complement to the laboratory investigation, the effect of tin on the microstructure and mechanical properties of gray iron was also determined in several commercial foundries. Additions of tin had about the same effect on mechanical properties and microstructure of commercial cast iron melted in cupolas as was obtained in cast iron melted and tested in the laboratory.

Tin in Nodular Cast Iron

In their original patent application, Millis, Gagnebin and Pilling[3] listed tin, lead, antimony, bismuth, arsenic, selenium and tellurium as subversive elements which interfere with the production of nodular graphite structures. Tin was singled out as being particularly detrimental, and it was recommended that it should not be allowed to exceed 0.05 per cent.

Morrogh[4] investigated the effects of some of the



Base iron.



0.058 Sn.



0.092 Sn.



0.33 Sn.

Fig. 2 — Microstructures of hypereutectic cast iron in 1.2 in. diameter bars. Base iron composition — 3.70 C; 2.45 Si; 0.65 Mn; 0.134 P; 0.101 S. 250×.

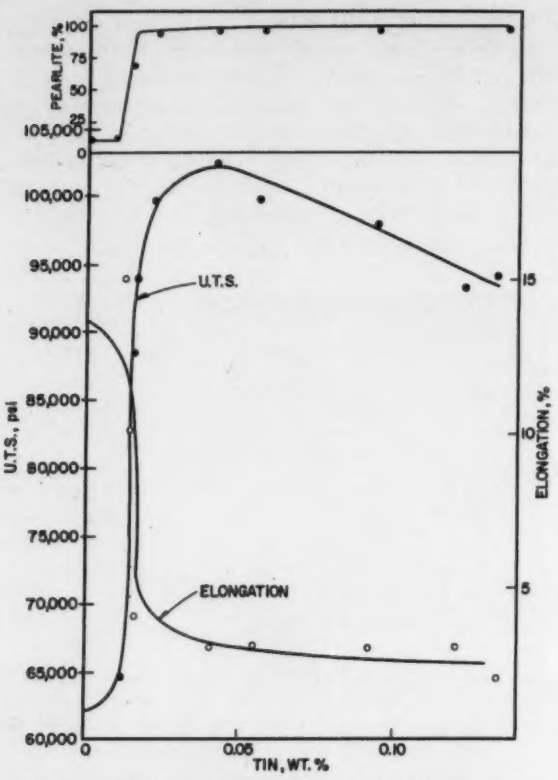


Fig. 3 — Tin effect on the structure and properties of nodular iron.[4]

so-called subversive elements on the structure and properties of nodular irons, and found that up to 0.13 per cent tin had no effect on the shape of the graphite nodule. However, addition of tin had a profound effect on the stability and occurrence of pearlite. An iron, the matrix of which was substantially ferrite without tin, may be transformed completely to pearlite by the addition of relatively small quantities of tin. This modification of the structure is naturally accompanied by considerable changes in mechanical properties.

Tin Effect on Properties

These are shown in Fig. 3, from which it will be seen that increasing the tin from zero to 0.041 per cent changed the matrix from 90 per cent ferrite to 100 per cent pearlite. Corresponding with this is an increase in tensile strength and hardness from 62,500 to 102,000 psi, and from 158 to 253 Brinell hardness, respectively, with a decrease in elongation from 15 to 3 per cent. Where maximum ductility is required, this change is undesirable, but where high-strength, wear-resisting engineering castings with reasonable shock resistance are required, 100,000 psi tensile with 3 per cent elongation is quite attractive.

Further additions of tin above that required for a completely pearlitic structure cause slight deterioration in tensile properties and some increase in hardness. However, it appears that up to three times the optimum amount of tin can be added without serious effects. Morrogh did not report massive cementite in the microstructure of iron when three times the tin required for a completely pearlitic structure was added. He clearly showed that the separation of the graphite as nodules was not affected by tin contents up to the maximum studied (i.e., 0.13 per cent).

While Morrogh's results were obtained incidentally to an extended research on the effect of several elements on the formation of nodular graphite in irons with a substantially pearlitic matrix, de Sy and Foulon[1] adopted a different approach. Although they did not wish to minimize the range of applications for which it is necessary to demand maximum ductility, they thought that the general tendency had been to exaggerate in this regard.

They contended, on the other hand, that the applications of pearlitic nodular graphite cast irons did not seem to have received sufficient attention, and considered that pearlitic nodular iron constituted a structural material of great interest, especially where resistance to wear was required. As a consequence, their investigation was aimed specifically at the attainment of nodular iron free from ferrite or cementite.

Manganese Content

In the first place, they investigated the possibility of obtaining the desired structure by variations in the manganese content. In the particular iron used, they found that free ferrite was still obtained with 1.33 per cent, and free cementite was produced by 1.47 per cent of manganese. Thus, while it was undoubtedly possible to get the desired results, the useful range of manganese content appeared to be inconveniently narrower than would be desired in usual foundry practice.

Since manganese is the least effective of the recognized carbide-formers, they concluded that the difficulty in obtaining the exact transition structure between pearlite and ferrite on the one hand, and pearlite and cementite on the other, would increase if stronger carbide-forming elements than manganese were used. Consequently, they turned their attention to nickel, copper and tin.

Under their experimental conditions, they found that the percentage of ferrite diminished with increase in nickel content, and became stable at around 10-20 per cent with about 2.5 per cent of nickel. Further additions of nickel did not decrease the volume of ferrite appreciably, and the matrix became self-hardening.

With copper, an addition of 0.8 per cent gave a completely pearlitic matrix and corresponded with maximum tensile strength. Larger additions caused a fairly sharp decrease in strength, which was associated with a degeneration in the surface regularity of the graphite nodules. No massive cementite was formed. With 3.2 per cent of copper, the graphite nodules were no longer recognizable as such and considerable loss in strength resulted.

The results from tin obtained by these authors were in good agreement with those of Morrogh. They found that additions of 0.03 per cent were sufficient to ensure a pearlitic matrix, and give an increase in tensile strength from 59,800 to 96,000 psi with a corresponding decrease in elongation from 13 to 3 per cent. Further increase in tin content did not produce

any free carbide in the microstructure, but tensile properties decreased and became rather more erratic than those obtained by Morrogh.

The promise of these results induced the Tin Research Institute to sponsor an investigation at Battelle Memorial Institute on the effect of tin on the microstructure and more important mechanical properties of nodular iron. The results of this investigation carried out by Williams, Keyser and Lownie[5] are reported herein.

EXPERIMENTAL

After preliminary experiments on small melts, two compositions of nodular cast iron, shown in Table 1, were selected for study. These compositions gave base irons with significantly different amounts of pearlite in the microstructure. Base iron *A* contained 71 per cent pearlite, and base iron *B* 58 per cent pearlite in a 1-in. Y-block.

A 600-lb melt of each of the two compositions was made in an induction furnace from low-sulfur and low-tin stock. Parts of these melts were drawn off and alloyed with 0.00, 0.04, 0.036, 0.050, 0.13 or 0.37 per cent of tin, each portion being treated subsequently with nickel-magnesium nodulizing agent, cerium and an inoculating addition of ferrosilicon in the normal manner. The treated melts were used to prepare A.S.T.M. standard Y-block castings measuring 0.5, 1.0 or 3.0 in. in the test section.

Each Y-block casting was sectioned as described in A.S.T.M. specification A.396-56-T. The sections were used for metallographic and mechanical-test specimens. The proportions of the micro-constituents were determined accurately on a number of samples, which were subsequently used as standards from which the proportion of ferrite in the remaining samples was determined by comparison. Brinell hardnesses were determined on the top inside surface of sections cut from Y-blocks. All other mechanical tests were made on duplicate tensile specimens as described in A.S.T.M. A.396-56-T. No heat treatment was carried out before testing.

Tin Effect on the Iron Matrix

Microstructures obtained when each of the base irons was alloyed with various amounts of tin, and cast into Y-blocks of different sections, are described in Table 2. With no addition of tin, base iron *A* contained 67.5 per cent to 71.5 per cent pearlite. With about 0.05 per cent of tin, the pearlite in this iron increased to about 95 to 97 per cent of the matrix. In base iron *B*, matrices containing 57.5 to 63 per cent pearlite were changed to contain 90 to 95 per cent of pearlite with the same tin addition.

Above about 0.05 per cent of tin, additional amounts had less effect than at lower levels of tin. The removal of the last traces of ferrite generally required about 0.1 to 0.2 per cent tin. Removal of the last traces of ferrite was more difficult in base iron *B*, which contained higher initial quantities of ferrite, than in base iron *A* (Figs. 4 and 5).

The effects of tin in both irons were relatively insensitive to section thickness. The castings 0.5 and 1.0-in. thick behaved in about the same way, and the 3.0-in. castings required only a little little more tin to make them completely pearlitic. The Y-block 0.5-in. thick contained more massive cementite, but this occurrence bore no relation to the amount of tin used. The cementite in the thin sections was probably the result of insufficient inoculation during the manufacture of the base iron.

The inoculating addition was calculated on the basis of the 1.0-in. Y-blocks, and appears to have been

TABLE 1 — CHEMICAL COMPOSITION OF TWO BASE NODULAR IRONS ALLOYED WITH TIN

Element	Base Iron A (71 per cent of Pearlite)[a], %	Base Iron B (58 per cent of Pearlite)[a], %
Carbon	3.52	3.42
Silicon	2.37	2.58
Manganese	0.51	0.42
Phosphorus	0.023	0.023
Sulfur	0.017	0.017
Nickel	0.42	0.42
Tin	0.003	0.003
Cerium	0.006[b]	0.006[b]
Magnesium	0.05[b]	0.05[b]
Carbon Equivalent (T.C. + ⅓ Si + ⅓ P)	4.31	4.28

[a] Measured in unalloyed 1-in. Y-block.
[b] Cerium and magnesium contents were estimated from recoveries obtained in preliminary heats.

TABLE 2 — MICROCONSTITUENTS IN MATRIX OF TWO DIFFERENT NODULAR IRONS ALLOYED WITH TIN IN CAST SECTIONS 0.5, 1.0 AND 3.0 IN. THICK

Base composition, %	Tin content, %	Microconstituents in 0.5-in. section, % of matrix			Microconstituents in 1.0 in. section, % of matrix		Microconstituents in 3.0 in. section, % of matrix	
		Pearlite	Ferrite	Massive Cementite	Pearlite	Ferrite	Pearlite	Ferrite
				Base Iron A				
3.52 C	0.003	69.0	24.0	7.0	71.0	29.0	67.5	32.5
2.37 Si	0.014	77.0	15.5	7.5	73.0	27.0	72.5	27.5
0.51 Mn	0.036	90.5	6.7	2.8	94.0	6.0	85.0	15.0
	0.05	96.0	4.0	0.0	97.0	3.0	95.0	5.0
	0.13	100.0	0.0	0.0	100.0	0.0	98.0	2.0
	0.37	95.0	0.0	5.0	100.0	0.0	100.0	0.0
				Base Iron B				
3.42 C	0.003	62.5	31.7	5.2	57.5	42.5	63.0	27.0
2.58 Si	0.014	72.5	27.5	0.0	64.0	36.0	61.0	29.0
0.42 Mn	0.036	88.4	11.6	0.0	86.0	14.0	75.0	25.0
	0.05	95.0	3.0	2.0	90.0	10.0	90.0	10.0
	0.13	98.0	3.0	0.0	98.0	2.0	96.0	4.0
	0.37	98.0	0.0	2.0	100.0	0.0	100.0	0.0

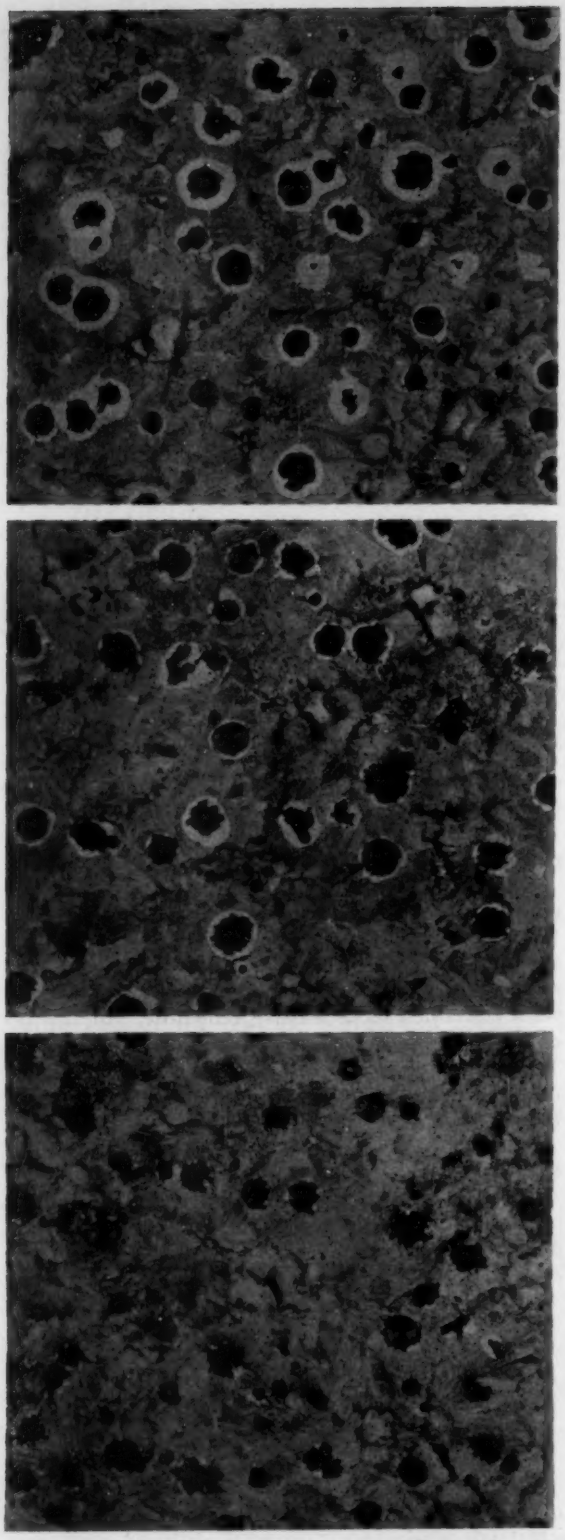


Fig. 4 — Microstructures of nodular cast iron *A*. Base iron composition — 3.52 C; 2.37 Si; 0.51 Mn; 0.023 P; 0.017 S; 0.42 Ni. 100×. Top: Base iron — 0.003 Sn, 29 per cent ferrite; center: 0.036 Sn, 6 per cent ferrite; bottom: 0.05 Sn, 3 per cent ferrite.

insufficient to prevent chill in the smaller blocks. It is particularly significant that in the 0.5-in. sections, which were found to produce massive cementite when only the residual tin content of the base iron was present, even large additions of tin did not increase this tendency. Tin, therefore, was shown to be highly effective in increasing the amount of pearlite in the matrix without causing the formation of massive cementite.

The minimum amount of tin required to produce a completely pearlitic matrix for each iron and section size is plotted graphically in Fig. 6.

Tin Effects on Mechanical Properties

The major mechanical properties of the two nodular irons containing various amounts of tin are given in Table 3. For a given thickness of casting and a given base iron, the Brinell hardness increased with increasing amounts of tin. The increasing hardness was appreciable with additions of up to 0.05 per cent of tin which are required to give a substantially pearlitic matrix. With additions beyond this amount, the hardness increased only gradually.

The effect of tin was about the same for all three sizes of sections studied, although the hardness at a particular level of tin increased as the size of the section decreased. This effect on section size applied equally to the base irons without tin additions and was normal in cast iron.

The results of tensile tests at various levels of tin showed that the tensile strength of the nodular irons under investigation increased to a maximum with increasing amounts of tin, and then decreased as more tin was added. Only when the amount of tin added exceeded the minimum required to obtain a pearlitic matrix was the tensile strength of the iron less than optimum.

Stress-strain data for nodular irons alloyed with tin were determined on tensile specimens cut from 1.0-in. Y-blocks. Tensile, modulus of elasticity and yield strengths were determined only for the base composition, and for the levels of tin nearest the minimum amount required to produce a completely pearlitic matrix. The results obtained are shown in Table 4, from which it will be seen that with amounts of tin sufficient to produce the required matrix, significant increases in the tensile modulus of elasticity and yield strengths of the iron were produced.

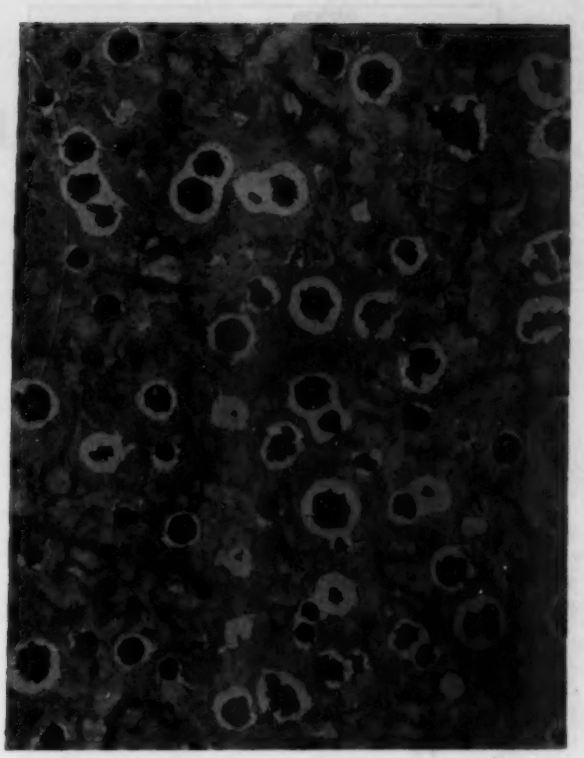
Tin Effect on Formation of Spheroidal Graphite

Tin is widely reported to be a subversive element which interferes with the formation of spheroidal graphite. Cerium is widely used to neutralize any subversive effects of small amounts of elements normally found as residuals in pig iron and high-quality scrap. It is important to recognize that in this investigation as little as 0.006 per cent of residual cerium was completely effective in neutralizing the potential subversive effect of as much as 0.37 per cent of tin.

From this it is apparent that useful amounts of tin may be added to nodular iron, with no more attention to its subversive effects than is required for any nodular iron. In addition, all the amounts of tin used in this investigation had no obvious effect on



Base iron, 0.003 Sn, 42 per cent ferrite.



0.036 Sn, 14 per cent ferrite.

Fig. 5 — Microstructures of nodular iron *B* in 1-in. section. Base iron composition — 3.42 C; 2.58 Si; 0.42 Mn; 0.023 P; 0.0175 S; 0.42 Ni. 100×.

TABLE 3 — TIN ADDITIONS EFFECTS ON MAJOR MECHANICAL PROPERTIES AND MICROSTRUCTURE OF NODULAR IRONS CAST INTO DIFFERENT SIZE Y-BLOCKS

Section Thickness, in.	Composition, %				Mechanical Properties			Microstructure		
	T.C.	Si	Mn	Sn	Tensile Strength, psi	Elongation[a] %	Bhn	Pearlite, %	Ferrite, %	Cementite, %
					Base Iron A					
3	3.52	2.37	0.51	0.003	97,500	8.1	220	68	32	—
				0.014	99,750	8.1	226	72	28	—
				0.13	106,750	4.0	258	98	2	—
				0.37	89,250	1.2	280	100	—	—
					Base Iron B					
	3.42	2.54	0.42	0.003	98,000	8.4	220	63	27	—
				0.13	103,750	2.4	265	96	4	—
				0.37	84,750	1.6	287	100	—	—
					Base Iron A					
1	3.52	2.37	0.51	0.003	99,250	6.8	224	71	29	—
				0.05	115,250	4.6	260	97	3	—
				0.13	114,500	3.3	267	100	—	—
				0.37	96,250	1.3	296	100	—	—
					Base Iron B					
	3.42	2.54	0.42	0.003	96,250	7.2	216	58	42	—
				0.13	103,500	2.2	278	98	2	—
				0.37	88,500	1.2	297	100	—	—
					Base Iron A					
½	3.52	2.37	0.51	0.003	99,000	5.5	249	69	24	7
				0.05	124,500	6.0	280	96	4	—
				0.13	126,000	5.0	282	100	0	—
					Base Iron B					
	3.42	2.57	0.42	0.003	97,500	6.5	250	62	33	5
				0.13	125,000	3.5	282	97	3	—
				0.37	88,000	1.0	317	98	0	2

[a] Elongation in 1 in. for specimens ½-in. thick. Elongation in 2 in. for specimens 1 and 3 in. thick.

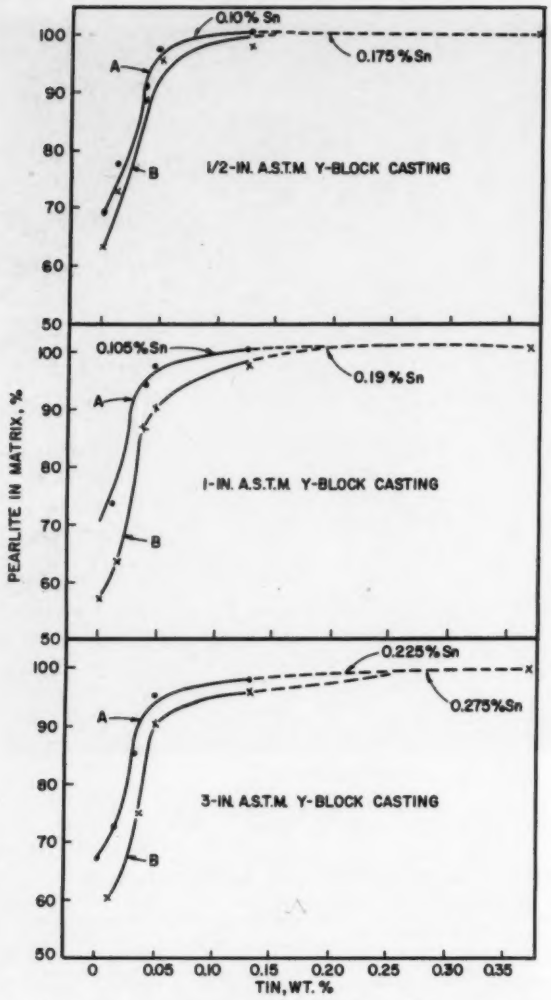


Fig. 6 — Tin content effect on matrices of two nodular irons cast in three different sized A.S.T.M. Y-Block castings. Line A composition — 3.52 C; 2.37 Si; 0.51 Mn. Line B composition — 3.42 C; 2.58 Si; 0.42 Mn.

TABLE 4 — EFFECT OF TIN ON THE TENSILE MODULUS OF ELASTICITY AND THE YIELD STRENGTH OF TWO NODULAR IRONS CAST INTO 1-IN. Y-BLOCKS

Base Iron Composition, %	Tin Content, %	Pearlite in Matrix, %	Tensile Modulus of Elasticity, psi	Yield Strength (0.2 per cent Offset), psi
		Base Iron A		
3.52 C 2.37 Si 0.51 Mn	Residual (0.003) 0.13	71 100	22,700,000 24,900,000	56,500 67,750
		Base Iron B		
3.42 C 2.54 Si 0.42 Mn	Residual (0.003) 0.13	58 98	24,400,000 26,200,000	57,750 85,500

the handling or foundry properties of the irons. Thus, the material containing tin handled, poured and behaved in about the same manner as the base iron to which tin had not been added.

SUMMARY

Tin had a marked effect in eliminating ferrite from the matrix of both gray and nodular irons. At the same time, tin, in amounts in excess of that required to promote a completely pearlitic matrix, did not lead to the formation of massive cementite. It has been shown that tin in amounts sufficient to give a substantially pearlitic matrix in gray iron, did not have any detrimental effect on the mechanical properties. Recent work showed the same to be true of nodular iron. Tin had no effect on the form of the graphite nodules in castings up to a 3-in. section.

REFERENCES

1. A. de Sy and J. Foulon, Twenty-first International Foundry Congress, Sept. 1954, *Metallurgia Italiana, 47*, (9), p. 421, 1955.
2. J. A. Davis, D. E. Krause, and H. W. Lownie, Jr., AFS TRANSACTIONS, *65*, p. 590, 1957.
3. K. D. Millis, A. P. Gagnebin, and N. B. Pillings; U.S. Patent No. 2,485,760.
4. H. Morrogh, *J. Research and Dev., B.C.I.R.A.*, April 1952.
5. R. C. Williams, N. H. Keyser and H. W. Lownie, Jr., Private communication.

CYCLIC PERMANENT MOLD OPERATION

some thermal aspects

By C. L. Goodwin and H. Y. Hunsicker

ABSTRACT

An experimental program was undertaken to determine the influence of mold section thickness, casting section thickness and operating rate on mold temperatures and time required for solidification of aluminum castings in permanent molds. These relationships were examined under dynamic conditions by repetitively casting rectangular aluminum alloy plates in coated cast iron molds of uniform section. The molds were cooled only by normal convection, conduction and radiation to the atmosphere.

Data developed in the investigation demonstrate certain fundamental principles and provide information which can be applied to the design and operation of permanent molds. It was found that mold temperatures, casting solidification rates and permissible speeds of operation are influenced considerably more by casting thickness than by mold thickness. However, operating characteristics can be changed significantly by varying mold thickness. For any selected production rate mold temperatures will be lower and solidification more rapid in thick molds than in thin molds. For a specified solidification rate, higher production is possible in thicker molds. These advantages may be partially offset, however, by handling and operating difficulties associated with thicker molds.

INTRODUCTION

Developments in the field of mechanized and precisely timed permanent mold equipment have increased the need for knowledge of the thermal aspects of the permanent mold process. Permanent molds are normally operated on a cyclic basis so that thermal relations of interest are dynamic in character. Knowledge of temperature conditions conducive to optimum production efficiency probably lags well behind the advances which have been made in mechanical equipment.

In a complete thermal analysis of permanent mold operation, consideration would have to be given to casting geometry, mold design, mold material, methods of cooling or heating, mold coating composition and thickness and length of operating cycle. The interrelation of these factors makes it difficult to rationalize the overall effects of changes in one variable, and has led to interesting discussions among permanent mold foundrymen. For example, all agree that a thin mold heats more rapidly than a thick mold when a casting is poured.

However, the thin mold cools more rapidly after the casting is removed from the mold. These observations lead to the conclusion that cyclic temperature fluctuations of thin molds exceed those of thick molds. It is difficult, however, to predict whether, with the same conditions of operation, a thick or thin mold will maintain a higher mean temperature. Similarly, the effect of mold temperature fluctuations on solidification of castings cannot be readily analyzed.

This difficulty of analytically determining the casting or mold geometry effects on permanent mold operation made a direct experimental approach to the problem necessary. An investigation embracing all possible variables would be prohibitively long. The scope of the present program thus was limited to determining, for castings and molds of simple shape, the influence of mold thickness, casting thickness and mold operating rate on mold temperatures and casting solidification times. While heat transfer was the fundamental consideration, it was obvious that factors of a metallurgical nature also were involved.

EQUIPMENT AND PROCEDURE

The experimental mold equipment consisted essentially of five sets of rectangular cast iron plates and five cavity spacers. The general arrangement of a typical plate and spacer combination is shown in Fig. 1. The spacers, which completely defined the size of the casting cavity, were interchangeable and could be mounted between any one of the five sets of mold plates.

Dimensions of the spacers and mold plates permitted the production of 0.25, 0.50, 0.75 and 1.50 in. thick castings in molds with walls 0.5, 1.0, 1.5, 2.0 and 3.0 in. in thickness. All the cavity spacers, and all the castings, were approximately 15 in. long and 6¼ in. high. The molds were mounted in an opening device actuated by compressed air. Metal was poured into the mold cavity through the central portion of the open upper edge.

An 0.004 in. thick refractory mold coating was ap-

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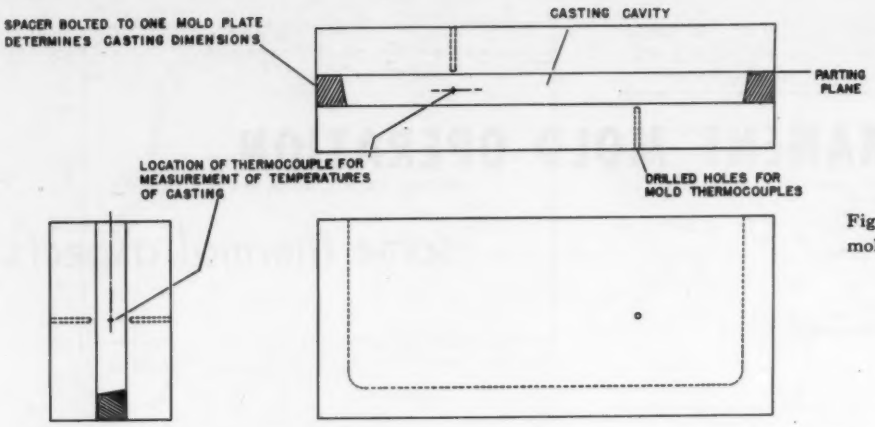


Fig. 1 — Sketch of experimental mold equipment.

plied to all cavity surfaces. Since small variations in coating thickness have a large influence on the chilling capacity of permanent molds, it was necessary to adjust this variable carefully. A magnetic thickness gage was used for controlling the coating thickness.

Temperatures in each of the mold plates were obtained with 26 gage chromel-alumel thermocouples located 1/8-in. from the mold-casting interface. Temperatures in the casting were observed by allowing similar thermocouples to solidify in the metal. Note in Fig. 1 the thermocouple in the casting was located adjacent to one of the mold thermocouples.

Thermal Considerations

All thermocouples were located so that the recorded temperatures would not be influenced by localized thermal conditions near the edges of the casting or adjacent to the metal entrance. The potentials developed by the thermocouples were recorded directly as temperatures on a high speed recorder. Thermocouples in the two mold plates were connected in parallel, and thus an average temperature for the two mold halves was recorded.

It was recognized that the alloy used in the investigation would have an influence on thermal conditions. Alloy 356 was selected because its solidification characteristics were considered representative of many alloys used in production.

The experimental procedure consisted of mounting a particular mold-cavity combination in the opening device, and then pouring castings at arbitrarily selected but fixed time intervals. Cyclic casting was continued until the system attained a state of dynamic thermal equilibrium as judged by the repetitive sequence of mold temperatures appearing on the recorder. When equilibrium had been established, careful records of mold temperature variations were obtained.

Simultaneously, a cooling curve was taken in the casting. Proceeding in this manner, each combination of mold and casting was brought to equilibrium for four or five different operating cycle times. Equilibrium was established on the longest cycle first and then on progressively shorter cycles until a limiting rate of operation was recorded, or until a limiting rate could be determined by a short extrapolation. The molds were cooled only by normal convection, conduction and radiation to the atmosphere.

The complete operating cycle for all of the experimental casting runs consisted of the following distinct phases:

1) The interval between pouring and opening of the mold for removal of the casting.
2) An interval after removal of the casting during which the mold remained open.
3) An interval during which the mold was closed prior to pouring.

The duration of the first interval was equivalent to the time needed for complete solidification of the casting as determined from the eutectic thermal arrest. The balance of the total cycle time was then arbitarily divided equally between the second and third phases of the operating cycle.

MOLD TEMPERATURES AND CASTINGS SOLIDIFICATION TIMES

Experimental data obtained in the foregoing manner consisted of a series of mold temperatures vs. time curves, and a corresponding series of cooling curves for the casting. Typical curves are shown in Fig. 2 for the combination of the 1.0 in. thick mold and 0.5 in. thick casting. Curves are included for four cycle times. The repetitive sequence of mold temperatures shown is typical for systems that have attained a state of dynamic thermal equilibrium.

Such a state is established when the mold temperature rises sufficiently so continuous heat dissipation from the mold balances the intermittent heat input from the casting. The small breaks apparent in the cooling portion of the mold temperature curves are the result of the slight drop in cooling rate coincident with closing the mold.

Cooling curves for the castings are shown just above the mold temperature curves. It is evident that solidification was retarded by the higher mold temperatures which developed when castings were poured with greater frequency. The shortest permissible cycle time, or the maximum permissible rate of operation, was reached when the mold became so hot that almost the entire cycle was needed to complete solidification in the casting. Bar graphs included in the lower portion of Fig. 2 show the lengths of the various portions of the operating cycle.

The mold temperature sequence for different op-

erating rates and for different mold-casting combinations may be compared most readily on the basis of the maximum and minimum temperatures recorded during an operating cycle. To facilitate this comparison, mold temperature data were assembled in curves of the form shown in Fig. 3. Curves are included in this figure for nine mold-casting combinations.

For any particular combination of mold and casting, the two curves which start in the lower right portion of the graph and converge toward the upper left represent the maximum and minimum mold temperatures developed by casting plates at various rates. These temperatures are actually plotted as functions of the total cycle time corresponding to each casting rate. The curve which starts in the lower left corner of each graph and intersects the minimum mold temperature curve at its upper extremity defines the time required to completely solidify the casting for each value of minimum mold temperature.

At the casting rate corresponding to the point of intersection of the minimum mild temperature curve and the solidification time curve, practically the entire cycle is required for solidification of the casting. More rapid rates of casting are not possible, and thus this time value defines the maximum permissible rate of operation.

It is evident from Fig. 3 that mold temperatures, casting solidification times and the maximum permissible speeds of operation are influenced considerably more by variations in casting thickness than by variations in mold thickness.

MOLD DIMENSIONS EFFECTS AT SPECIFIC OPERATING RATES

A number of mechanical operations is generally involved in the production of permanent mold castings. These include pouring the casting, removing and cooling cores, opening the mold, removing castings and reassembling mold parts. Duration of the operating cycle in many cases is determined more by the time required for these mechanical operations than by time needed to solidify the casting. This is particularly true with molds designed for manual operation.

In such cases, the mold designer should have some knowledge of the relationship between mold thickness and operating mold temperature. This is important because mold temperatures influence solidification rates and thus the metallurgical quality of castings. Mold temperatures also affect mold life.

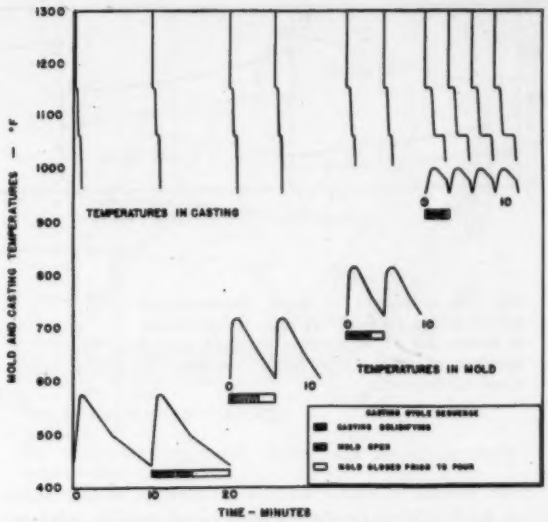


Fig. 2 — Cyclic temperature variations when casting 0.5 in. thick castings in a 1.0 in. thick mold.

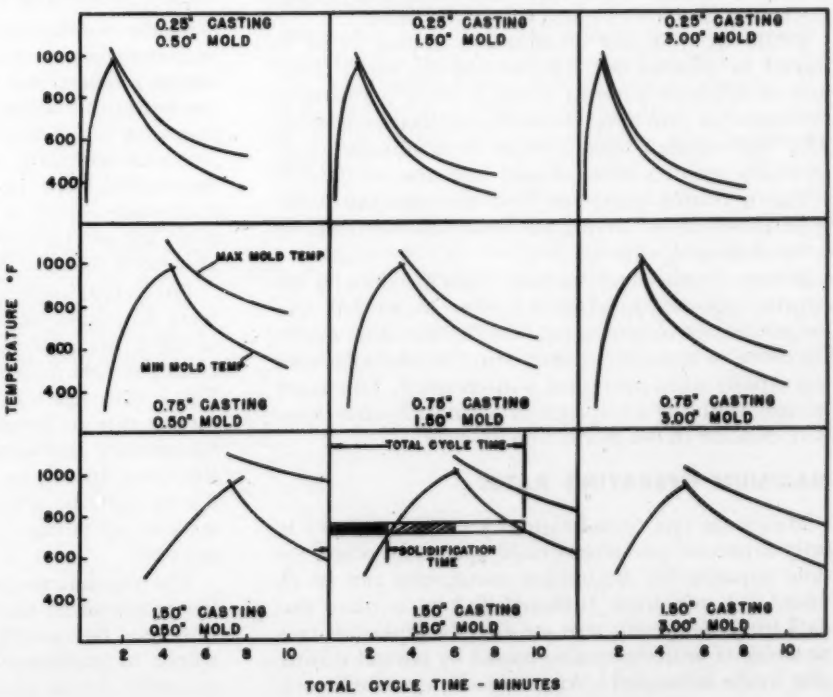


Fig. 3 — Curves showing effect of total cycle time on mold temperatures and casting solidification times. Curve starting in lower left of each graph shows time to solidify casting for any specified minimum mold temperature. Relationships are shown for the nine different mold-casting combinations.

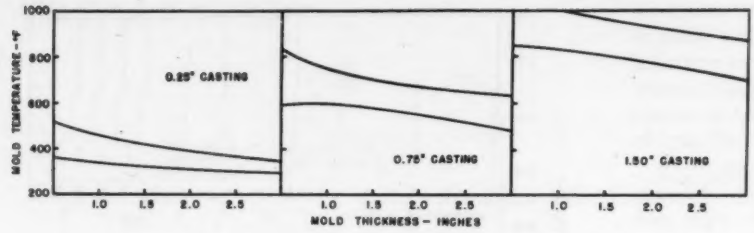


Fig. 4 — Mold thickness effect on maximum and minimum temperatures of molds operated on an 8 min cycle. Relationship is is shown for three values of casting thickness.

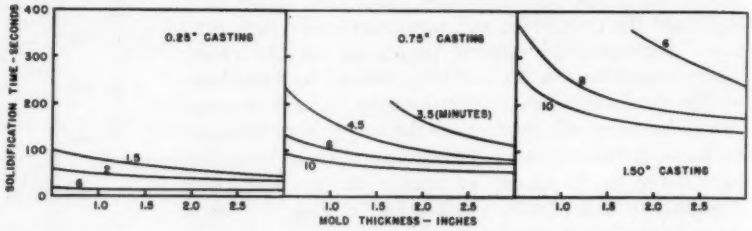


Fig. 5 — Influence of mold thickness on solidification time of castings. Relationship is shown for three casting sections and a number of total cycle times (figures affixed to curves).

The effects of mold section thickness on the maximum and minimum temperatures of molds operated at a specific rate are shown, for three values of casting thickness, in Fig. 4. It is evident that thicker molds generally operate at lower temperatures than thin molds. A change in mold thickness from 0.5 to 3.0 in. may lower the operating temperatures by as much as 250 F (121 C).

The influence of mold thickness on solidification of castings is shown in Fig. 5. This influence, of course, is a consequence of variations in mold temperature. It is apparent that when the casting rate is constant the solidification rate increases with increasing mold thickness. For example, on a 4.5 min cycle, a 0.75 in. thick casting solidifies in 165 sec in a 1 in. mold, compared with 100 sec in a 2 in. mold. This effect is more marked with thicker castings and at the higher production rates. A higher solidification rate is usually desirable because of the finer structure and higher mechanical properties which will be obtained.

While the forgoing relationships appear valid, it should be pointed out that varying the mold thickness in different parts of a single mold is relatively ineffective in providing thermal gradients for feeding. The high thermal conductivity of aluminum alloys generally requires more drastic measures to develop adequate thermal gradients. Such measures can usually be provided by varying the mold coating thickness or by chilling locally.

Because of their greater heat capacity, thick molds require a greater number of casting cycles and take longer to reach operating equilibrium than thin molds. By virtue of the same characteristic, thick molds cool less rapidly when operation is interrupted. This latter feature would be advantageous when short shut-down periods occur in the production foundry.

MAXIMUM OPERATING RATES

Maximum rate considerations are of importance in fully automatic permanent mold operations where the time required for mechanical movements can be reduced to a minimum. It should be kept in mind that data from the present tests are directly applicable only to molds of uniform section cooled by normal dissipation to the atmosphere. Automatically operated molds may be cooled in this manner, or may be equipped with means for accelerated cooling.

It was pointed out previously that the maximum permissible operating rate is realized when the mold becomes so hot that almost the entire casting cycle is needed for solidification of the casting. Limiting rates determined from the intercepts of minimum mold temperature curves with casting solidification curves (Fig. 3) are shown in relation to both mold thickness and casting thickness in the diametric projection of Fig. 6. The figure shows that casting thickness is the most influential factor determining the ultimate casting rate. Mold thickness variations are of some consequence, however, and it will be observed that the increase from 0.5 to 3.0 in. provided an average increase in maximum operating rates of approximately 40 per cent.

MOLD OPERATION WITHIN SPECIFIED LIMITS

At the maximum permissible rate, mold temperatures will be in excess of 950 F (510 C). Since these higher temperatures may accelerate deterioration of the molds, it may be desirable to specify a maximum operating rate which will hold mold temperatures below some arbitrarily selected value. These rates can be determined from the maximum mold temperature vs. cycle time curves of Fig. 3. Production rates with a mold temperature limitation are, of course, considerably lower than the theoretical maximum rates.

Production rates which are possible with an arbitrary limitation on the maximum mold temperature of 800 F are plotted both against mold section and casting section thickness in the diametric projection of Fig. 7. Once again it will be noted that any particular casting may be produced at a faster rate within this temperature limitation by increasing the mold section thickness. Increasing the mold thickness from 0.5 to 3.0 in. permits, depending upon the specific casting section, operating rate increases of from 50 to 100 per cent.

The maximum operating rate may be defined by the requirement that a specific minimum solidification time be assured. This might be desired, for example, to produce a casting of requisite metallurgical structure. At the maximum permissible rate of opera-

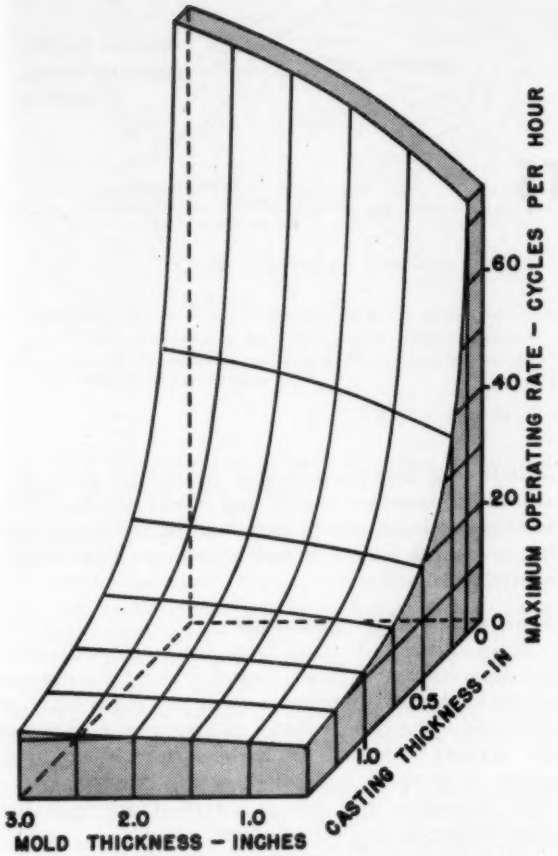


Fig. 6 — Diametric projection illustrating relationship between mold thickness, casting thickness and maximum permissible operating rate. Molds cooled only by normal dissipation.

tion, solidification is slow and may be accelerated only by operating more slowly or changing the conditions of mold cooling. Operating rates necessary to assure specified solidification times also may be determined from the mold temperature and solidification time curves of Fig. 3.

PRODUCTION CASTINGS

Commercial castings are more complex than the simple plates employed in this experimental program. The solidification time for production castings is usually not established by the thickness of the predominant areas of the casting but by the size of bosses, ribs or risers. Nevertheless, the basic principles illustrated by this investigation for simple castings should be applicable to more complicated parts. Probably the major differences in production work arise from use of accelerated cooling with forced air or water.

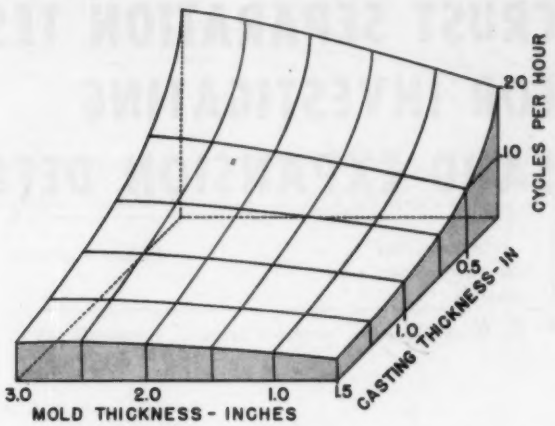


Fig. 7 — Operating rates which limit maximum mold temperature of 800 F.

SUMMARY

When aluminum castings are produced cyclically in permanent molds, mold temperatures, casting solidification times and permissible operating rates are influenced considerably more by the thickness of the casting than by mold thickness. Significant changes in these operating characteristics, however, can be made by changing the mold thickness within practical limits. For a selected production rate, thicker molds operate at lower and more uniform temperatures and provide more rapid casting solidification. For any specified rate of solidification, higher production is possible with thick molds.

Moreover, with thick molds short delays in production cause less disturbance of normal operating conditions. These advantages are in some cases partially offset by the greater difficulty of handling and operating heavy molds, and compromises are sometimes required.

Although the solidification rate of a casting is affected by mold thickness, variations in thickness within a single mold are not particularly effective in controlling the direction of solidification and feeding. This problem must be solved by appropriate gating and risering augmented by control of coating thickness and sometimes by localized heating or cooling.

CRUST SEPARATION TEST FOR INVESTIGATING SAND EXPANSION DEFECTS

By P. W. Goad

ABSTRACT

The development of a new technique for studying expansion defects is described, together with some applications of the technique. It has been shown that expansion failures can occur at relatively low temperatures at which sand expansion is in the range 0.25 — 0.7 per cent, and in the presence of wood flour in excess of 3 per cent. These results indicate that expansion failures can occur even when deformation substantially exceeds expansion.

Moisture content of the sand has been found to be a vital factor in determining the nature of the expansion failure and susceptibility.

INTRODUCTION

A conspicuous gap in past investigations of expansion defects has been the failure to determine sand temperatures at their initiation. It has generally been believed that temperatures in the vicinity of the α-β quartz inversion at 573 C (1063 F) were most favorable because expansion reached its maximum at this point.

This paper describes a technique for investigating the behavior of molds under thermal shock conditions similar to those during casting, and its application to the measurement of sand temperatures at the initiation of expansion defects. A study of the mechanisms of formation of scab and rattail defects, which accompanied the development of the technique, has led to a more complete understanding of the fundamentals involved.

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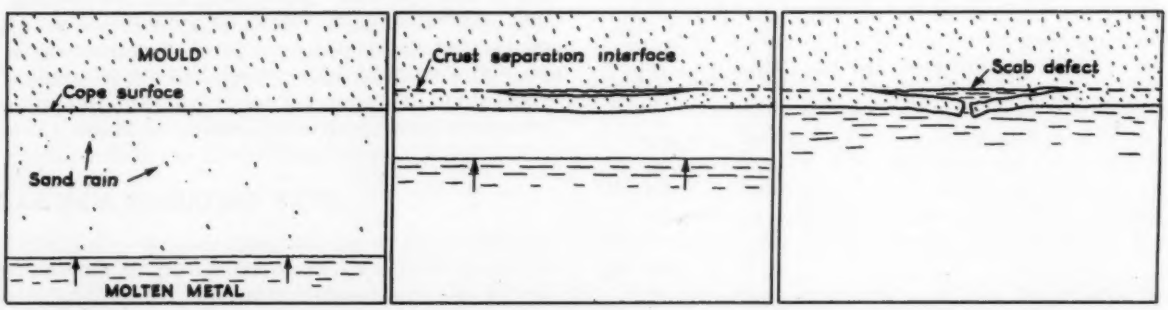
P. W. GOAD is associated with Foundry Sands Section, Chemical Research Laboratories, Commonwealth Scientific and Industrial Research Organization.

Scab and Rattail Formation

Scabs and rattails result from the same basic cause, the separation of a crust of sand under the influence of rapid thermal expansion from the main body of the mold. The defects differ in appearance owing to the manner in which the mold surface is ruptured, which in turn depends on how the crust separates. The formation of a scab at a typical site on a flat cope surface occurs in three stages:

Stage 1. The mold surface is heated rapidly by radiant heat from the molten metal as it rises in the mold, and rapid expansion of the surface sand follows. The first effect of expansion is the expulsion of some individual grains from the surface as grain repacking commences, and a "sand rain" falls[1] (Fig. 1a).

Stage 2. Heat penetrates sufficiently to cause expansion at an appreciable depth (several grain diameters at least). Grain repacking locks the sand into a cohesive crust, and the sand rain ceases. Lateral expansion of the crust proceeds beyond the stage where it can be accommodated by grain repacking, and the crust then separates from the cooler adjoining sand,

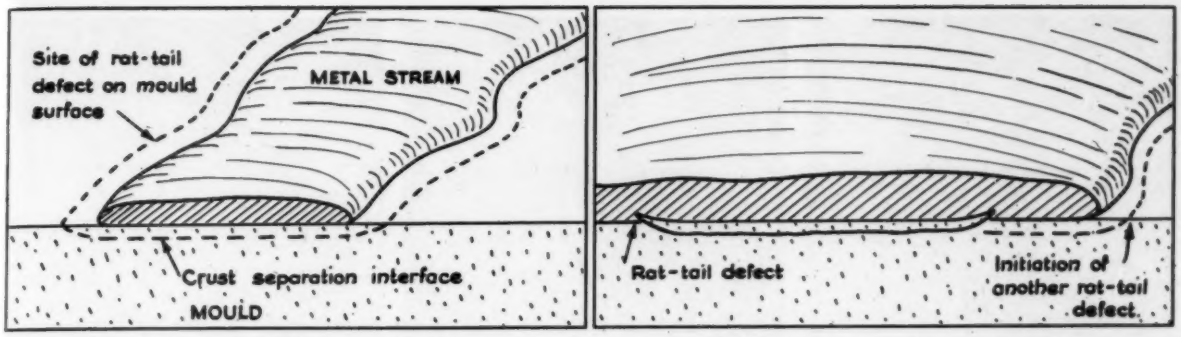


a): Expansion of surface sand causes expulsion of some grains and a "sand rain" falls.

b): Cohesive crust of dried-out sand separates from moist and adjacent sand.

c): Metal flows through fractured crust to form scab defect.

Fig. 1 — Stages in the formation of a scab at a cope surface.



a): Crust of dried out sand forms under metal stream, and as it expands initiates mold surface cracks at sides of metal stream.

b): Expansion of crust continues when molten metal covers mold face.

Fig. 2 — Formation of a rattail at drag surface.

and with the assistance of gravity bulges downward toward the rising metal (Fig. 1b).

Stage 3. The separated crust breaks when or just befor the metal touches it, and the metal flows through the break and into the space behind the crust (Fig. 1c). If the crust breaks and the broken parts are displaced but the metal does not penetrate behind, the resulting casting defect is usually called a buckle, but it is in fact a partly formed scab. If part or all of the separated crust falls onto the metal and is carried away, the defect becomes a "blind" or "dummy" scab.

While the typical site of a scab is a cope surface receiving radiant heat from the metal as it fills the mold, rattails (except in thin section castings) occur only at drag surfaces. Their formation depends on a mechanism which was originally described by Pettersson.[2] In the initial stages of casting, molten metal flows over drag surfaces either in separate streams or with an irregularly advancing front. Underneath the metal a surface sand layer is rapidly heated, and accompanying lateral expansion causes it to separate as a cohesive crust from the underlying sand.

Crack Formation

Directly underneath the metal stream the crust is of uniform thickness, but beyond the metal boundary the crust separation interface slopes up to terminate at the mold surface. At this location a crack is formed (Fig. 2a). The crust continues to expand after the molten metal has covered the mold face (Fig. 2b), and the cracking is intensified either until expansion ceases or the metal solidifies. In thin section castings the metal stream may touch or closely approach the cope surface, and a rattail may then be formed there in an identical manner. The position of a rattail defect is invariably related to the position of the metal boundary at some stage in the covering of the surface (Fig. 3).

Rattails may have the appearance of furrows, steps or even of scab-like fins, if the metal penetrates down the crust separation interface (Fig. 4). Defects which are best described as scabs, but which share some of the characteristics of rattails, may be formed where there is an exposed edge in the mold (corresponding to a re-entrant angle in the casting, e.g., in the AFS 8-J test casting[3]), which allows a separated crust to expand freely into the mold cavity.

Moisture Effect

When crust separation occurs in a green sand mold, the separation interface is the boundary between a hot dried-out surface layer and moist underlying sand.[4,5] Separation cannot be regarded simply as a mechanical failure in a uniform material loaded to its limit, because the expansion forces are concentrated at a well-defined location—the junction between the relatively strong dry crust and the far weaker moist sand in the main body of the mold.

Conditions at the interface may be further influenced by the formation of a region of high moisture content behind the dried out surface layer, caused by the condensation of moisture evaporated from the surface.[6] Temperature rise in the high moisture region is delayed by evaporation while the surface temperature increases rapidly. This may lead to the formation of a temperature "step" across the interface.

Crust Separation Test

Scab defects can be studied by direct observation, since it is a simple matter to produce scab-type failures in a sand face by suddenly exposing it to a source of intense radiant heat. However, in determining initiation temperatures this is not a satisfactory approach because there is unlikely to be a definite indication of the precise moment of crust separation, and it is possible for a separated crust to reach maximum expansion without the occurrence of a surface rupture.

On the other hand, when the rattail mechanism operates, crust separation and surface rupture occur almost simultaneously, as even a small expansion of the separated crust must break the mold surface, and there is, therefore, a visible indication of the initiation of the failure. This characteristic of the rattail mechanism has been utilized in the "Crust Separation Test."

This test simulates the effect of metal streams flowing over a mold face and producing rattail-type failures in the sand surface. The specimen, which consists of a narrow strip of a flat sand face, is suddenly exposed to a source of intense radiant heat. The center of the strip is protected from direct heat by a shield placed a small distance above the sand face. The directly-heated areas behave as if they were cov-

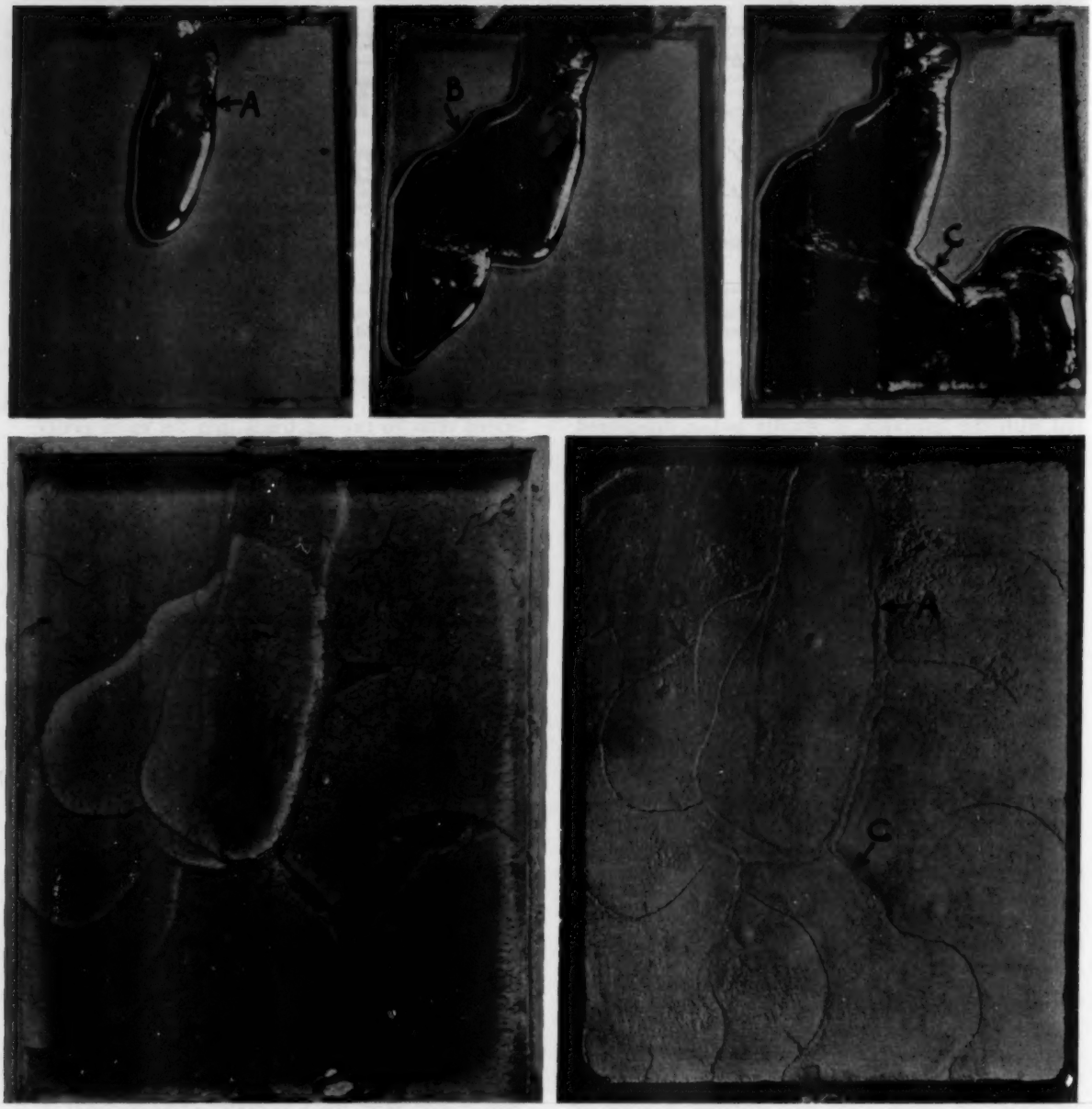


Fig. 3 — Relation between metal flow pattern and rattails. The upper row shows three stages in the pouring of a plate in an open-top mold. The position of the metal boundary at each stage (e.g., locations A, B, C), can be related to a rattail defect in the casting (lower right). The mold, after removal of the casting (lower left) shows typical crust separation. (The photograph of the casting has been reversed in printing to aid comparison with the other pictures.)

ered by broad metal streams, and the sand protected by the shield corresponds to the sand lying alongside the stable boundary of a metal stream.

Consideration of the factors which might be involved in the operation of the metal stream mechanism suggested that surface failure would probably be accelerated if:

1) Two metal streams were so situated that their crust separation interfaces met at the mold face.
2) These streams were wide so that lateral expansion of the crusts beneath them was considerable.
3) Stress concentrators were present at the failure line.

Obviously, these conditions seldom all apply in practice, but they can be easily produced in the test. By adjusting the width and height of the shield, the crust separation interfaces on either side of the shield can be made to meet at the sand surface along the transverse center line of the sand strip (Fig. 5). The length of the strip is limited mainly to the type of heat source used, but, in the present form of the test, metal streams 4½-in. wide are simulated. Stress concentrators at the failure line are provided by notches on each side of the strip. The general arrangement of the test apparatus is shown in Fig. 6.

The test specimen is rapidly slid or swung under



Fig. 4 — Bronze plate casting, showing scab-like defect in drag surface caused by penetration of metal down crust separation interface.

the heat source (this takes less than 1 sec), and the strip surface is observed until either a failure occurs or until the surface temperature in the directly-heated areas is above about 700 C (1292 F). The sand surface under the shield is illuminated by a powerful light beam projected at a low angle. Small disturbances of the sand surface in the early stages of crust separation, which would be invisible with normal illumination, are plainly silhouetted. Usually, the failure can be seen to flick across from notch to notch in a thin dark line (Fig. 7).

The appearance of typical failures in moist synthetic sands is shown in Fig. 8. In Fig. 9 rattail-type failures produced by metal streams are compared with that produced in a crust separation test specimen.

Test Uses

Any of three criteria might be used in evaluating the susceptibility of a sand to expansion defects.

1) Crust temperature at initiation of separation failure.
2) Amount of standardized additive (e.g., wood flour) required to prevent separation failure.
3) Time interval to separation failure.

As will be shown later the usefulness of (1) appears more limited than (2) and (3).

Sand Specimen

A flat sand face measuring 10 x 5 in. is prepared by ramming up a 10 x 5 x 2 in. box. The face is cut away to a depth of 3/8-in., leaving a central strip 2½-in. wide extending from end to end of the box. Notches ¼-in. deep are cut on both sides of the strip at its center. Over the notches the heat shield, which is made of 0.1-in. steel sheet, is placed with its top surface parallel to and 5/16-in. from the sand face. The width of the shield is 1 1/16-in., but this is not a highly critical dimension. The test specimen is illustrated in Fig. 10.

Heating Method

As a readily controllable source of heat is neces-

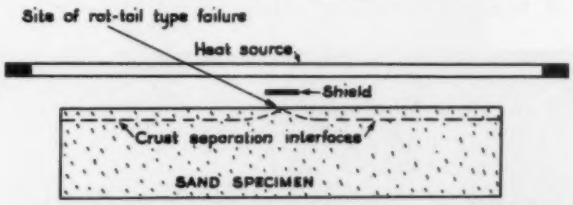


Fig. 5 — Diagram showing location of crust separation interfaces in the crust separation test.

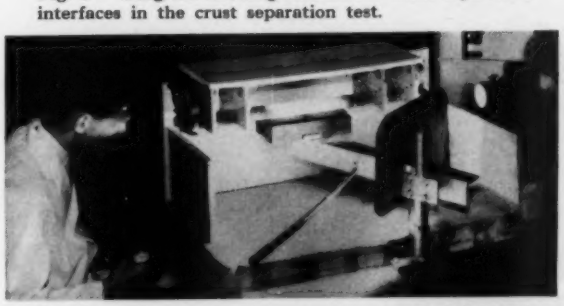


Fig. 6 — General arrangement of crust separation test apparatus.



Fig. 7 — Crust separation failure soon after initiation, as seen by operator.

sary if tests are to be reproducible, and as it is essential to observe the sand surface throughout the test, it is not practicable to use molten metal. Wire wound resistance elements operating at approximately 900 C (1652 F), were used in early work, but a more intense heat proved desirable. Two 1-in. diameter silicon carbide rods set at 1⅞-in. centers, operating at 1300-1400 C (2372-2552 F) have proved satisfactory.

Casting Tests Correlation

Experiments were made to compare the susceptibility to rattails, as indicated by the crust separation test, with the formation of these defects on a test casting. Sand mixes are based on a silica sand of AFS fineness 100 and contained 2, 4 and 6 per cent western bentonite, and 2.4 per cent moisture. Wood flour (soft wood, -85 mesh) was added in steps of 0.25 per cent until the sand mix showed no tendency to failure in the crust separation test. The mix with the "end-point" addition of wood flour was used as a facing for molds in which the test castings were poured. End-point wood flour additions for the three sands were:

Bentonite in sand mix, %	Wood flour required to prevent failure, %
2	3.5
4	3.5
6	3.0

Casting conditions were designed to favor the initiation and maximum development of a large number of failures. Aluminum, chosen because of its low density and intermittent, sluggish-type of flow, was poured slowly at a temperature well above its freezing point into an open top plate mold. Details of the casting procedure were as follows:

Mold: 11 x 9 x ½ in., major face horizontal, open top, single gate centrally placed on one of the short sides; mold hardness approx. 75.
Metal: Commercially pure aluminum.
Pouring conditions: Temperature, 800 C (1472 F), mold filled in 30 sec (pouring rate about one-sixth of normal).

The appearance of castings made in sand mixes with and without the end-point wood flour additions, are shown in Fig. 11. The wood flour additions were obviously insufficient to eliminate the defects completely. However, when casting conditions were altered to bring them more into line with normal foundry practice (viz., the mold was poured through two gates, and filled in 5 sec) the defects were eliminated. The castings were still imperfect, particularly near the gates, because the extreme friability of the sands rendered the molds susceptible to erosion with the faster pouring rate (Fig. 12).

In selecting a casting test as a standard of comparison it is difficult to decide how severe the conditions should be. However, it is clear that the end-point additions of wood flour necessary to inhibit rupture of the surface in the crust separation test are greater than the amounts usually considered necessary or desirable in foundry practice. This fact suggests that the test is more exacting than normal foundry conditions.



Fig. 8 — Typical rattail-type failures produced in crust separation test. Top left: Failure immediate after initiation. Top right: Failure after completion of sand expansion. Bottom: Appearance of test specimen after failure. Part of separated crust has been displaced to show its cohesive nature. Transverse cracking of crust at left occurred on cooling of the specimen.

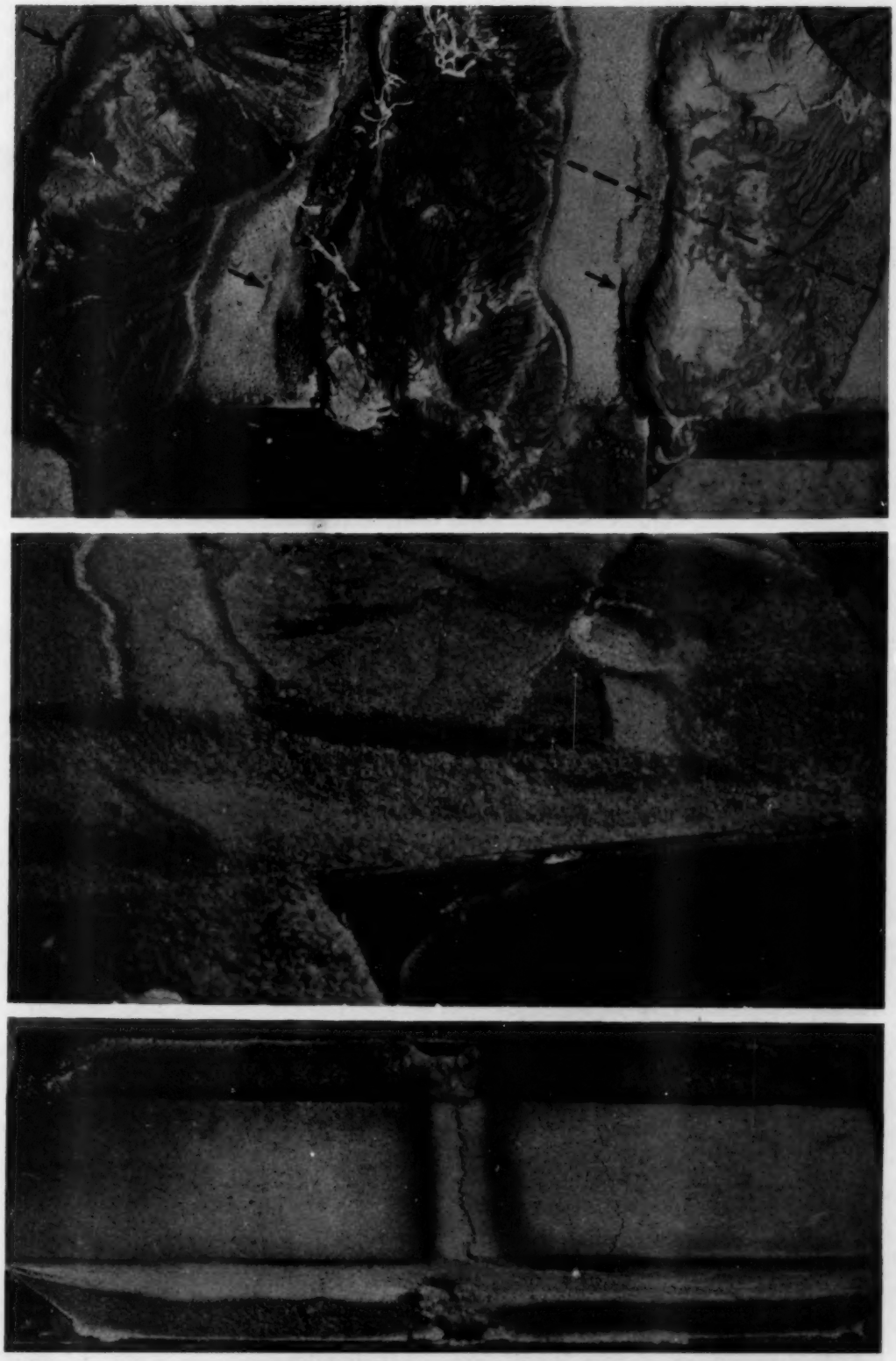


Fig. 9 — Comparison of crust separation failures produced by metal streams and in crust separation test. Top: Surface cracking (arrowed) produced alongside streams of molten bronze allowed to flow continuously over a sloping mold face. Photograph taken after streams solidified. Center: Mold face after removal of solidified metal and sectioning (at location shown by dotted lines in top photograph), showing crust separation. Bottom: Failure in crust separation test specimen, showing surface cracking in shielded area, and separated crust. This specimen was not notched.

It is of interest to note here the small effect of considerable variation in bentonite content on the end-point wood flour additions.

As yet, no work has been done to correlate results from the crust separation test with the occurrence of scab defects in test castings. However, it seems unlikely that a sand could show scabbing tendencies without being susceptible to rattails, as conditions pertaining when rattails are formed are far more favorable for crust separation and rupture of the mold surface. In this connection, the following aspects of rattail and scab-type failures are important:

1) When the rattail mechanism operates, there is little opportunity for expansion to be accommodated by grain repacking, as resistance to expansion is small. The crust merely slides up the sloping portion of the crust separation interface. At the typical site of a scab defect, the crust separation interface does not meet the surface and, in the early stages at least, more resistance to expansion is likely. Also, crust separation does not immediately cause surface rupture.
2) The degree of heat shock is greater when rattails are formed as the metal is in contact with the sand.
3) Molten metal lying on the drag surface of a mold face assists in the formation of a thin but cohesive crust, since surface grains are not pushed out of position as readily as from a cope surface. Observations by the author have indicated that a crust capable of causing a severe rattail may be only two or three sand grains in thickness.

CRUST SEPARATION TEST APPLICATION

Sand Temperature at Initiation of Expansion Defects

Since the actual site of a rattail-type failure in a green sand mold is part of the crust separation interface, and its temperature when failure is initiated is, therefore, probably close to 100 C (212 F) [the interface being the boundary between the hot dried out surface layer, and the moist underlying sand], measurement of its temperature serves no useful purpose. However, measurement of the maximum temperature of the crust at the moment of separation a) provides a measure of the upper temperature involved in crust separation, and, therefore, in expansion defects, and b) can be used in determining the expansion necessary to cause separation.

The temperature of the crust rises steeply from 100 C (212 F) at the interface to a maximum at the mold surface. Knowing the surface temperature, an

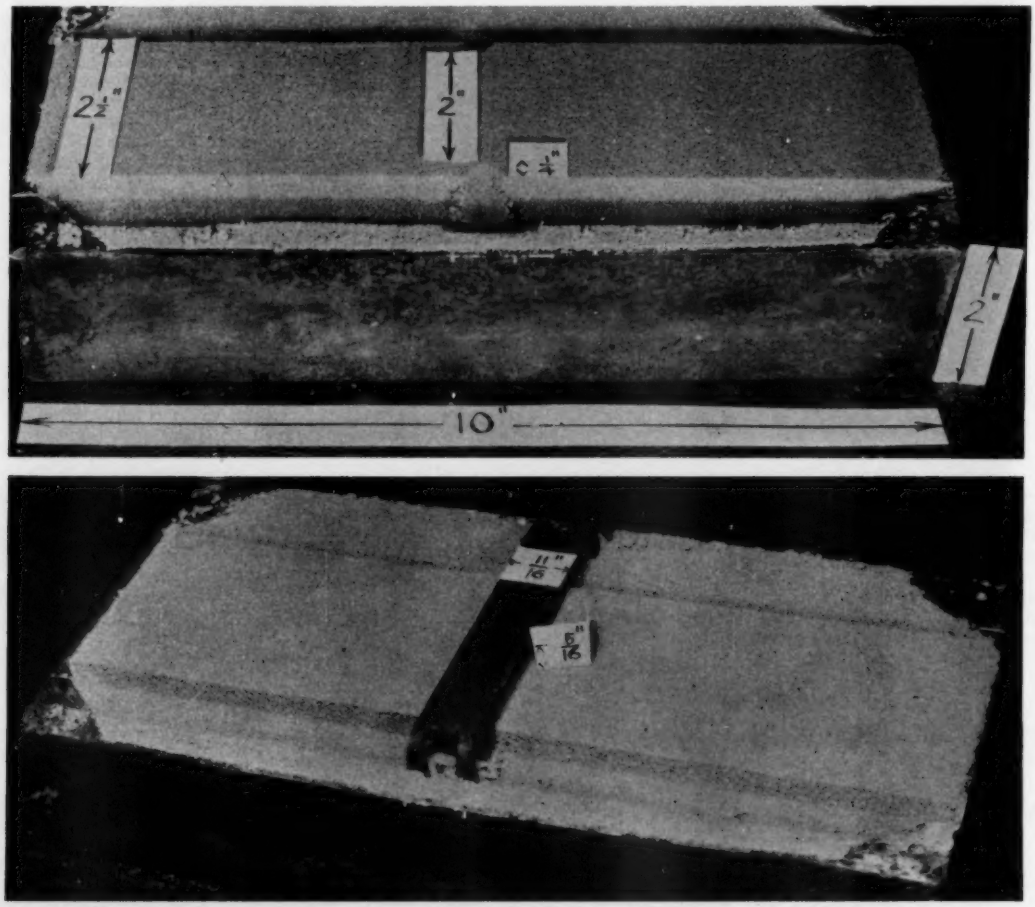


Fig. 10 — Crust separation test specimen. Top: After cutting a strip and notching. Lower: With shield in position. Projecting strips are placed at the ends of the specimen (Fig. 7) if the sand crusts tend to expand over the top of the box.

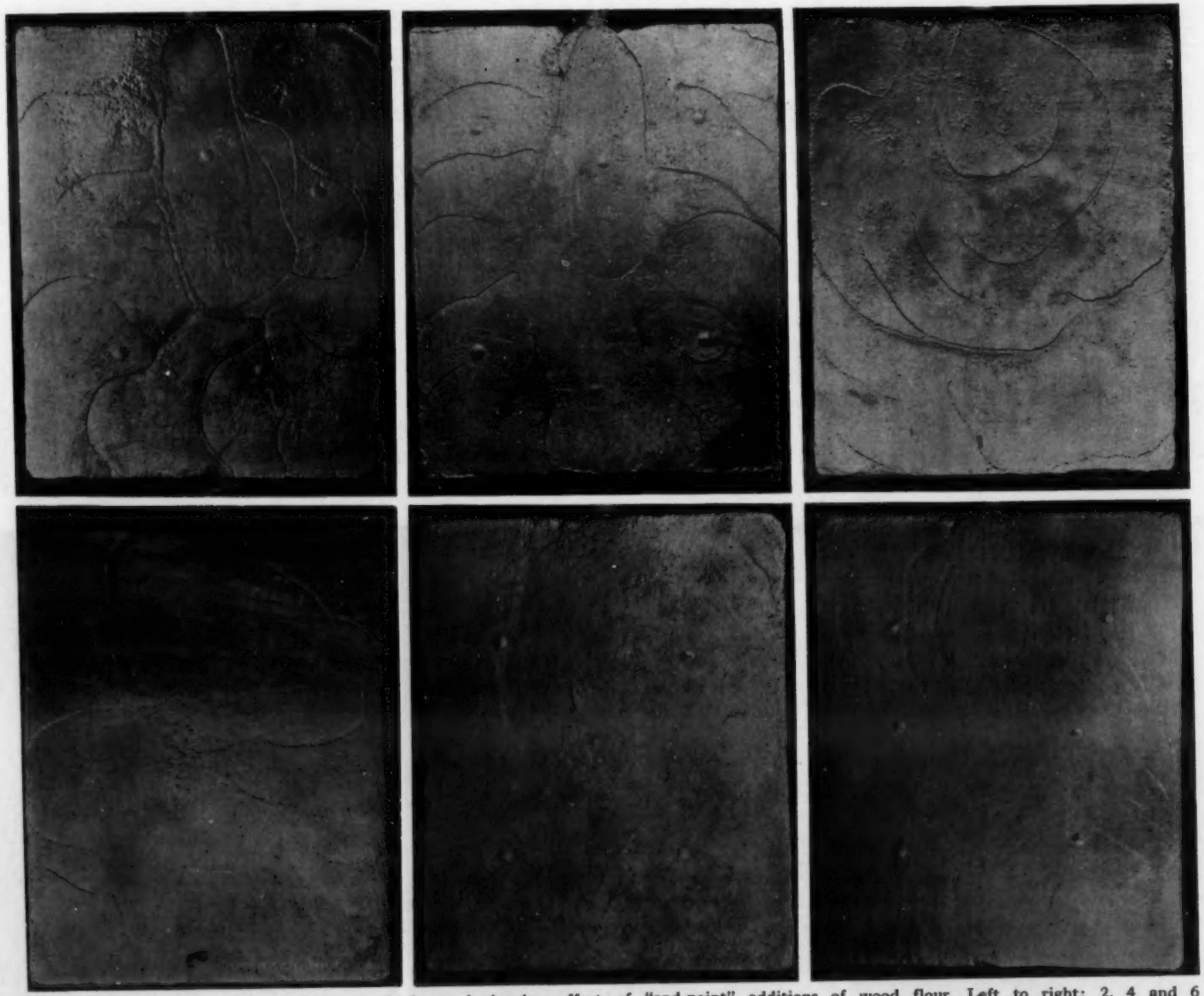


Fig. 11 — Drag surface of plate castings made in green synthetic sand showing effect of "end-point" additions of wood flour. Left to right: 2, 4 and 6 per cent western bentonite. Top: No wood flour. Severe rattailing. Bottom: "End-point" additions of wood flour (3.5 per cent, 3.5 per cent and 3.0 per cent, respectively). Slight rattailing.

adequate estimate can be made of temperature conditions within the crust. It is to be expected that measurements can only be made with a low degree of accuracy, as rate of temperature rise at the specimen surface when the failure is observed is of the order of 30-50 C (86-122 F) per sec, and there can be no certainty that the temperature measuring device is exactly at the same temperature as the surrounding sand grains.

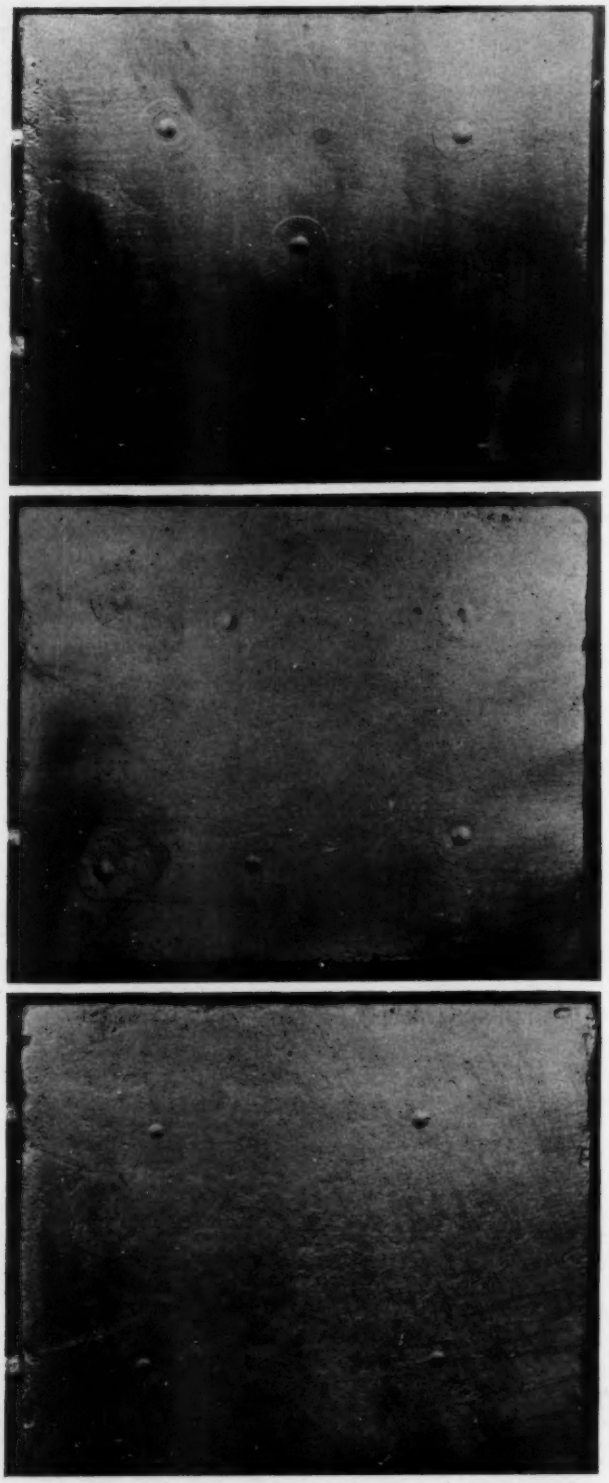


Fig. 12 — Drag surface of Casting made in green synthetic sands containing "end-point" addition of wood flour, and cast at fast pouring rate through two gates (Top of casting is at left in photo). Top to bottom: Sand contained 2, 4 and 6 per cent western bentonite.

However, even an accuracy of ±50 C (122 F), which should be achieved without great refinement of technique, yields a useful result. Because of the equipment, time and experimental care necessary for measurement of crust temperatures, this approach should find application in research rather than in routine testing.

Fine thermocouple wire (0.010 in. diameter or less), and a fast response electronic recorder are used. The coupled wires are embedded in the specimen in a directly heated zone. The welded junction, which is made as small as possible, is placed with its top surface level with the sand surface and slightly exposed, and is packed tightly around with sand grains, which should remain undisturbed throughout the test. The recorder trace is interrupted when the failure is first observed.

Results from the limited amount of work performed to date are shown in the Table.

Temperatures chosen for publication in the Table were those considered likely to err on the high rather than the low side. Much lower temperatures, which suggest that crust separation can occur in natural and synthetic sands when maximum crust temperatures are in the vicinity of 200 C (392 F), have been recorded, but further work will be necessary before this claim can be made with confidence. Duplicate determinations have not been in close agreement, but this is felt to be caused more by experimental error than by variation in the sands behavior.

Conclusions

Conclusions which may be drawn are:

1) Expansion at temperatures between 100 C (212 F) and the quartz inversion at 573 C (1063 F) can initiate crust separation failures.
2) At the initiation of the failure, the temperature of the bulk of the sand in the separating crust may

TABLE — MOLDING SAND TEMPERATURE AT INITIATION OF EXPANSION FAILURE IN CRUST SEPARATION TEST SPECIMEN

Molding Sand	Max. Sand Temp. When Crust Separation Failure Observed, C (F)	Approx. Expansion of Molding Sand at Temp. of Crust Separation Failure,* %
Dandenong natural sand		
AFS fineness—120	400 (752)	0.5
Clay substance—15%	480 (896)	0.7
Moisture—6%		
Synthetic Sand A		
AFS fineness—100	270 (518)	0.25
Western bentonite—2%	380 (716)	0.45
Moisture—2.4%		
Synthetic Sand B		
AFS fineness—100	345 (653)	0.35
Western bentonite—4%	410 (770)	0.5
Moisture—2.4%		

*From data of Parkes and Godding.[7]

be closer to 100 C (212 F) than 573 C (1063 F).

3) Regarding susceptibility to crust separation failure, there appears to be little significance in an expansion deformation relationship. The Table also contains an estimate[7] of the sand expansions at the maximum temperatures of crust separation. These are relatively small (0.25-0.7 per cent), and corresponding deformation values of the sands would be unlikely to fall below these expansions at any temperature. In comparing the values, it should be remembered that expansion of the bulk of the sand in the crust is considerably lower than that occurring at the surface.

Additive Amount Required to Prevent Failure

The amount of a standardized additive to prevent failure (end-point addition) is readily determined, and there is little possibility of experimental error. However, as the amount of the addition is likely to be so large as to be impracticable for foundry use (see final discussion), the test cannot be used in a straight forward manner to estimate the amount of an additive needed to ensure freedom from expansion defects under production conditions. Rather, it is a means of comparing one sand with another, and a guide to the effectiveness of different additives and the influence of sand variables.

An example of the use of the end-point addition criterion is to be found in the description, given earlier, of work to correlate results from crust separation tests and casting tests. It was found there that differences in the end-point additions for synthetic sands containing 2, 4 and 6 per cent western bentonite were only slight, and this indicates that in practice these sands would show a similar susceptibility to rattailing. The results of the casting tests support that conclusion.

The end-point addition criterion has also been used to show the influence of moisture content on susceptibility to crust separation failure (Fig. 18).

Time Interval to Failure

Wood Flour and Other Additions Effect. Time interval to failure is directly related to temperature of the sand crust, and measurements are far more easily made and less subject to experimental error than temperature measurement. Accurate standardization of heating rate is not necessary if results are expressed in terms of the time interval to failure of a standard specimen of high susceptibility. The only additional equipment required is a stop watch.

An example of the effect on time interval to failure of progressive additions of wood flour to a synthetic sand containing 6 per cent western bentonite and 2.4 per cent moisture, is shown in Fig. 13. The curve shows the time interval to be a function of wood flour content. It appears probable that if such a curve were determined for a system sand, it could be used in the routine checking of wood flour contents. If a number of additives are used in the sand, their combined effect will be reflected in the time interval to failure. If this value is maintained at a constant level, this should ensure consistency in those sand properties affecting susceptibility to expansion defects.

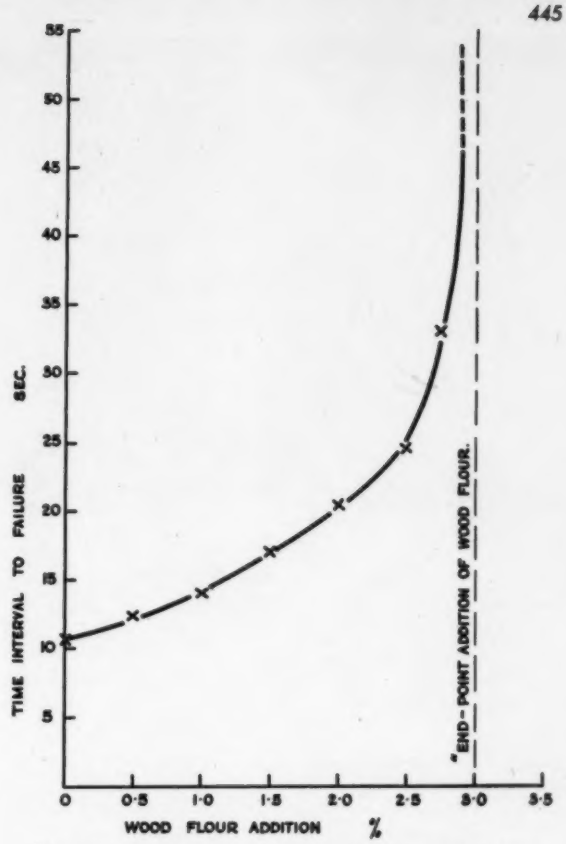


Fig. 13 — Effect of wood flour additions on time interval to failure in crust separation test. Tests performed on synthetic sand (fineness 100) containing 6 per cent bentonite and 2.4 per cent moisture.

Moisture Content Effect on Mold Behavior Under Thermal Shock. Moisture content of the sand is frequently disregarded in investigations of expansion defects, but great importance has been attached to this by Continental workers.[8,9] Pettersson, for example, considered that below about 0.4 per cent moisture, molds were probably not subject to scabbing. Some experiments on the effect of moisture content on susceptibility to crust separation failures were carried out by the author using both the crust separation test and the casting test described above (single gate, slow pouring rate). The synthetic sands containing 2, 4 and 6 per cent bentonite, which were used in the correlation tests, were again employed in this work. The sands were rammed to a mold hardness of approximately 75.

Behavior of Completely Dry Sands. Crust separation test specimens and plate molds were dried by heating for 4 hr at 150 C (302 F), and cooled and stored until required in sealed bins containing silica gel.

In the crust separation test, the dry sand showed behavior quite distinct from that of green sand. The cohesive crusts of uniform thickness in the directly heated areas, and the single crack between the notches in the shielded area characteristic of green sand, were not produced. Instead, failure of the specimen was by a random flaking over the entire surface (Fig. 14).

The appearance of the drag surface of plates cast

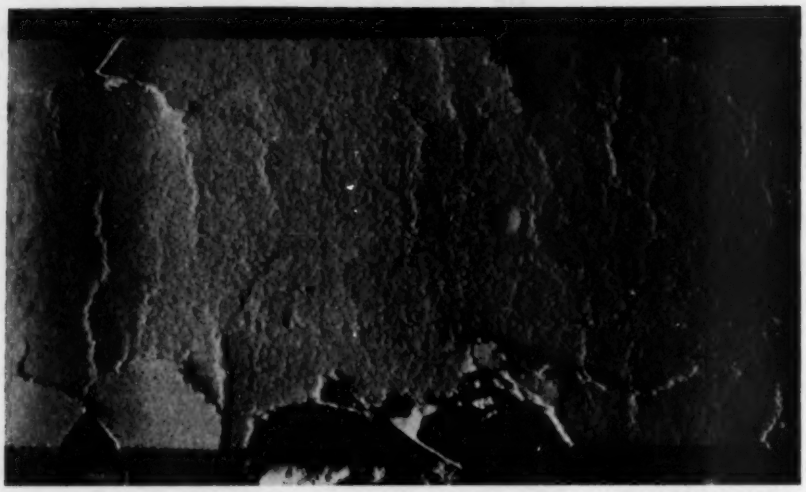


Fig. 14 — Appearance of completely dry sand specimen after testing, showing random flaking.

in the dry molds is shown in Fig. 15. The defects exhibited were probably best described as crazed buckles. Unlike rattails, these defects were not oriented with respect to the metal flow pattern, but were the result of random cracking of the mold surface under the molten metal (Fig. 16). Examination of the mold after removal of the solidified castings showed that failure of the surface was by flaking in a manner similar to that shown by the crust separation test specimens.

Determinations of time interval to failure in the crust separation test indicated that dry sand containing no wood flour or other additive had only a slightly greater resistance to expansion failures than green sand. However, results of work which is now proceeding suggest that the amount of additive required to inhibit expansion failures in a completely dry sand may be much less than that required when the sand contains sufficient moisture to impart normal moldability.

Absorption of Moisture from Atmosphere After Drying. Crust separation test specimens and open-top plate molds were dried by heating at 150 C (302 F), then allowed to cool and stand overnight exposed to the normal laboratory atmosphere.

Behavior of the sands on testing was in all respects similar to that of the sand in the green condition. An example of the extent of rattailing in the casting test is given in Fig. 17. This work showed that it was possible for sufficient moisture to be absorbed from the atmosphere during the cooling and storing of a dried mold to cause the behavior characteristic of a green sand. A similar conclusion was reached by Pettersson.

Moisture Content Effect on Wood Flour Required to Inhibit Rattail-Type Failure in Green Sand. It had been noted during the work on correlation of results from the crust separation test and casting tests, that if moisture content of a sand containing the end-point amount of wood flour were increased, then the sand again became subject to rattailing.

Crust separation tests were made to determine end-point wood flour additions with increasing moisture content, and the results are shown in Fig. 18.

DISCUSSION AND CONCLUSIONS

The results of the crust separation test, which show that the expansion required to cause mold failures is small, do not support the widely held belief that expansion must exceed deformation for the sand to be susceptible to expansion defects.[10,11,12] This idea is also not supported by the results showing that rattailing can occur in synthetic sands containing more than 3 per cent wood flour, the deformation values of which should be high.

Doubt as to the significance of deformation (that is, amount of which the sand can deform under a compressive load, without loss of cohesion) arises when the mechanism of crust separation is considered. Unless resistance is encountered, molding sands expand freely, and if there is free expansion the deformation of the sand is of little account.

When a crust forms under a metal stream, resistance to expansion cannot be great as the crust is able to slide up the sloping portion of the separation interface. Similarly, there can be little effective resistance to expansion when a crust, which may be only ⅛-in. thick, forms over a large area of the cope surface, and the combined effect of gravity and its own expansion is acting to bow it downwards.

The significant factors appear to be the lateral expansion force exerted by the crust and the sand strength at the site of separation interface. Crust separation occurs when the former exceeds the latter. Sand behavior in the crust separation test indicates that failure occurs suddenly after a period of heating, which means that there is some build-up of resistance to expansion. Measurement of the expansion force at the temperature of separation (as could be done with the AFS "Restraining Load Test" apparatus) gives the limiting value of the sand strength at the interface. This value is of considerable theoretical interest.

One of the most important results of this work was to show the effect of moisture. While in moist sands expansion failures occurred by separation of a cohesive crust (the cause of scab and rattail casting defects), in the complete absence of moisture failure occurred by random flaking (the cause of a crazed buckle-type of casting defect). With sands having a

high susceptibility to crust separation in the moist state, e.g., synthetic sands containing no additive such as wood flour, rattailing occurred regardless of variation in moisture content, providing a low minimum moisture content was exceeded.

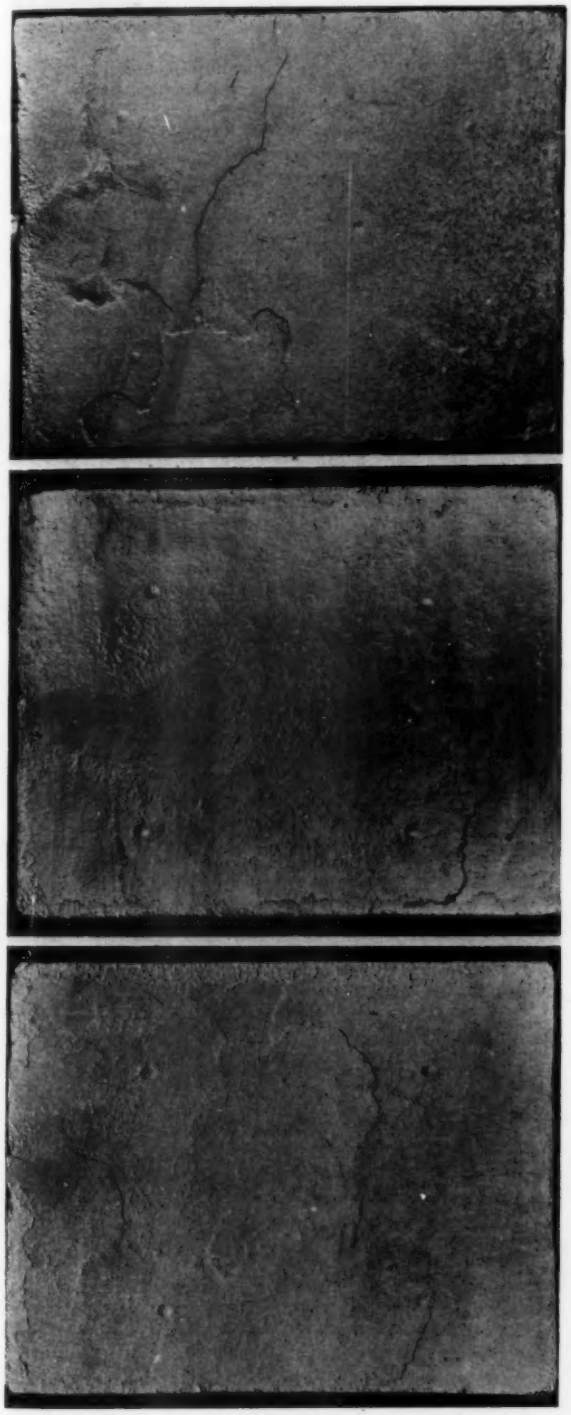


Fig. 15 — Drag surface of casting made in complete dry synthetic sands (Top of castings is at left of photo). Top to Bottom: Sands contained 6, 4 and 2 per cent western bentonite. The castings all show crazed buckle defects, caused by mold failures, similar to those shown in Fig. 14.

Sufficient moisture could be absorbed from the atmosphere on cooling and standing, after drying to render a mold as readily susceptible to rattails as a green sand mold. However, with sands containing an



Fig. 16 — Appearance of completely dry mold after removal of solidified casting, showing random flaking type of failure. The metal flow pattern is indicated by the light colored streaks. The casting made in this mold is shown in Fig. 15 (top) (The photograph of the mold was reversed in printing to aid comparison of mold and casting surfaces).



Fig. 17 — Drag surface of plate casting made in dry synthetic sand (containing 6 per cent western bentonite), showing severe rattailing. The mold was dried and allowed to stand overnight before casting.

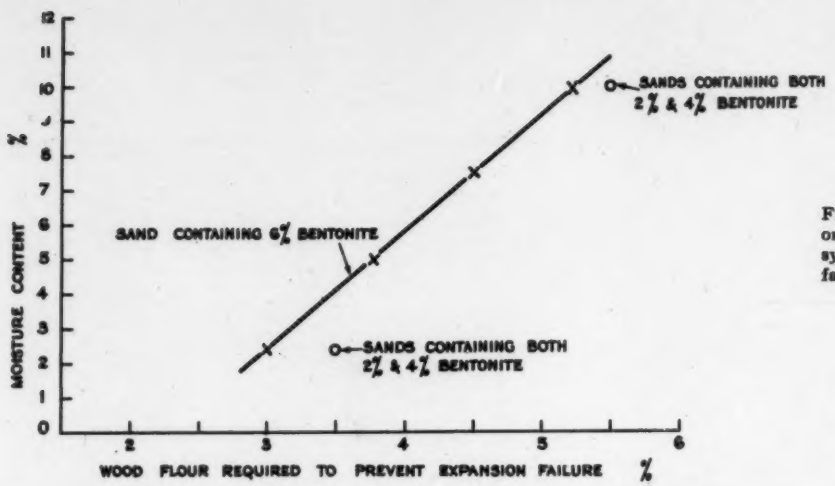


Fig. 18 — Moisture content effect on amount of wood flour required in synthetic sands to prevent expansion failure in the crust separation test.

appreciable amount of additive, variation in moisture content had a significant influence on susceptibility. The conclusion to be drawn is that in practice, mold drying may have no useful effect on reducing scab or rattail defects, unless either moisture absorption from the atmosphere (and from the backing sand of skin dried molds) is completely prevented (which would mean pouring almost immediately after drying, or storing molds in a perfectly dry atmosphere), or else the sands contain an additive such as wood flour.

An interesting result was the high percentages of wood flour required in the crust separation and casting tests to prevent rattailing (e.g., with a synthetic sand containing 4 per cent bentonite and 2.4 per cent water, 3.5 per cent wood flour was shown to be insufficient, in a severe casting test). The amount of additive required in synthetic sands was obviously far too high for foundry use. It would not be economical and the sand would have unsatisfactory molding properties. The aim rather should be to use a mimimum of additive, and pay greater attention to practical aspects of molding and casting likely to affect the incidence of the defects.

The crust separation test is proposed as a simple means of reproducing in the laboratory casting conditions which are highly favorable to the formation of crust-separation-type defects. The test is still under development, and the research program for which it was devised is incomplete, but there appears good reason for believing that, as well as serving its original purpose, it will find application in routine testing in the foundry.

REFERENCES

1. Pettersson, H., "Spalling of Green Sand Moulds and its Relation to Casting Defects," *Proc. Inst. Brit. Found., vol. 48,* p. A203, 1955.
2. Ref. 1. p. A209.
3. AFS Sand Division Committee 8-J, "The Scabbing Defect," AFS TRANSACTIONS, *vol. 63,* p. 124, 1955.
4. Nicolas, P., "Behaviour of Moulding Sand at High Temperature," *Giesserei, vol. 41,* p. 337, 1954.
5. Levelink, H. G., "Behaviour of Green Moulding Sands During Shock Heating as Regards the Occurrence of Surface Defects on Castings," *Giesserei, vol. 45,* p. 4, 1958.
6. Gittus, J. H., "The Influence of Mould Factors on Casting Defects," *B.C.I.R.A. J. Res. and Devel., vol. 5,* no. 5, p. 268, April 1954.
7. Parkes, W. B., and Godding, R. G., "Behaviour of Moulding Sands at High Temperatures," *Proc. Inst. Brit. Found., vol. 48,* p. A192, 1955.
8. Ref. 1, pp. 204, 206, 210, 214.
9. Ref. 5, p. 7.
10. Ref. 1, p. A204.
11. Ref. 3, p. 135.
12. Ref. 7, p. A188.

SOLIDIFICATION TIMES OF SIMPLE SHAPED CASTINGS IN SAND MOLDS

By J. Berry, V. Kondic and G. Martin

ABSTRACT

A critical review of experimental and theoretical work concerned with the time of solidification of simple sand-cast shapes reveals the existance of many controversial points. These arise through the several assumptions made in the basic mathematical treatments of the rate of solidification when applied to sand cast shapes.

The assumptions generally made in these equations have been re-examined in this work, and certain experiments carried out to check them.

The solidification times of three simple shapes made in three casting alloys (Al with 12 per cent Si, Cu with 8.5 per cent Al and eutectic carbon equivalent cast iron) have been measured experimentally for a large range of dimensions and for several mold conditions. A prediction of the solidification times for the particular range of sizes and shapes examined, has been shown to be generally possible, providing that the values of physical constants for the mold and casting media, mean values of mold-metal interface temperature, and the maximum temperature attained in the mold after pouring are available.

The chilling power of several mold materials under various conditions of the mold have been determined by a calorimetric method. In addition, the prediction of chilling power has been attempted by means of a theoretical method. A satisfactory agreement between the two methods has been found, and reasons for the deviation occurring have been put forward.

PART I
LITERATURE REVIEW AND PRELIMINARY WORK

INTRODUCTION

Soundness is the aim one strives for in most sand castings. The methods practiced to achieve this objective vary, but are based mainly on the choice of appropriate alloy and casting technique and, when possible, on consideration of the casting design. For a given casting design the soundness is controlled chiefly by risering and running methods of castings, by metallurgical controls of the alloy and the mold, and, in addition but less frequently, by the application of chills and paddings. In the greatest number of cases the approach to solving the problem of casting soundness in industry is essentially empirical, and only recently, and with a moderate degree of success, has a more scientific method of dealing with this problem been attempted.

The reasons for this belated application of science are readily apparent, since the solution to the problem of casting soundness depends mainly on the application of basic scientific knowledge from three different fields; flow of metal into the molds (hydraulics), heat flow from the casting into the mold (heat transfer) and laws of solidification of alloys (physical metallurgy). However, a good deal of such basic knowledge is still lacking in these fields, although a steady progress has been maintained during recent years. This paper is an attempt to review the present position in one of these fields, namely, that of heat transfer.

In the field of heat transfer two major difficulties have been encountered. First, the solution of heat flow equations requires the availability of thermal data on both the metal and the mold, data which are not yet generally available. Second, shaped castings in sand molds present such variations of designs that a purely mathematical approach to solving the heat transfer problems in such cases is almost impossible. Consequently, the important question that immediately arises is whether a theoretical approach to the heat flow problem in sand molds is worth while. The question of collection of thermal data is not unsurmountable, but the question of applying such data still remains to be discussed.

Empirical Relationship

The most significant early work with the question of heat flow in sand molds is due to Chvorinov.[1] Fundamentally, he showed that the theoretical problem of heat transfer in molds can be reduced to an empirical relation; that is, solidification time in a mold is dependent on the ratio of volume to surface area of the casting. More recently, some American workers have shown that Chvorinov's relationship is

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subject to certain limitations, particularly those concerning the size and shape of castings. This has also been confirmed by other workers.

The main outcome of such theoretical studies could be summarized as follows: Heat flow equations for certain types of simple-shaped castings can be reduced to a few elementary relationships which can be then used to predict solidification time of geometrically similar shapes. There is little doubt that further theoretical progress in this field is most likely to continue in the same general direction—toward obtaining basic heat flow relations which are derived from the heat flow theory combined with the experimental data. These relationships could be more readily applied than could the more complex mathematical equations of heat flow.

It is a distant but, nevertheless, a foreseeable hope that such relationships will be found to embrace the basic sizes and shapes which are encountered in the sand foundry industry.

Assuming that the knowledge concerning the solidification times and rates in a sand mold exist, the question of practical application of such data requires some examination. Perhaps the most direct use of solidification-time data is for dimensioning risers in shaped castings. If a riser is to efficiently feed the solidification shrinkage of a casting, it is essential to ensure that the casting solidifies towards the riser. At the same time, from an economic and production standpoint, it is required that the riser should be of a minimum size.

A quantitative solution of such problems require data concerning the times of solidification in sand molds. The problem of successful risering of castings, of feeding distances, i.e., of alloy characteristics, is thus a heat flow as well as a metallurgical problem.

Other possible uses of heat flow in sand founding are—the control of casting stresses, the calculation of knockout times and the interpretation of cast structure with their resultant mechanical properties.

The object of the present paper is that of reviewing some recent results in heat flow studies in sand molds. A later paper will examine the application of such data to the production of sound sand castings. Some recent work of the authors is used for the basic construction of the present paper.

LITERATURE REVIEW

The general problem of heat transfer in relation to shaped castings in sand molds can be considered under three headings:

A) Basic laws of heat transfer.
B) Variables of shape and size.
C) Variables of the mold and casting material.

Basic Heat Transfer Laws

Castings (ingots as well as shaped castings) fall into the group of unsteady state heat transfer phenomena. The development of the general heat flow equation for sand castings, and the conditions of its application, have been reviewed by Ruddle[2] and King.[3] For that reason only a brief statement of this aspect will be made here.

The mathematical statement of the problem can be made with respect to the case of a plane body of semi-infinite length and width initially at temperature θ_1, brought suddenly into contact with a second, similarly shaped body, initially at a lower temperature θ_2. This second body (e.g., the mold) absorbs per unit area a certain amount of heat q from the first body (e.g. molten metal) in time t. In order to set up a mathematical equation which will give the amount of heat exchanged between the two bodies with respect to the time, specific conditions must be stated. Still further conditions are necessary for this equation to be solved.

Conditions Necessary

Both of these aspects of the problems have been adequately reviewed by Ruddle and King. The important conclusion which is obtained from such an analysis is that an equation for the above case, relating to the exchanged heat q per unit area with respect to the metal and mold temperatures θ_1 and θ_2, respectively, can be obtained when the following conditions are fulfilled:

1) The interface is at constant temperature θ_1' during the period for which the heat flow is studied.
2) The mold has an infinite heat capacity, i.e., will not loose any heat to the surroundings during the same period.
3) The thermal properties (specific heat, c, and thermal conductivity, k) for both metal and mold are constant and not variable for the temperature range under examination.

Under these conditions the heat absorbed per unit area by the mold is given by:

$$q = \frac{2}{\sqrt{\pi}} \cdot (\theta_1 - \theta_2) \cdot b_2 \sqrt{t} \qquad (1)$$

where $\theta_1' > \theta_2$ and $\theta_1 > \theta_1'$ and constant b_2 is the property known as the heat diffusivity of the mold: $b_2 = \sqrt{k_2 \rho_2 c_2}$, where all the quantities have been defined above and ρ_2 is the density of the mold.

From the point of view of applying this heat flow equation to the shaped castings in sand molds the following problems arise:

1) Errors due to the finite size and variable shape of castings as distinct from the plane and infinite bodies.
2) Errors due to the lack of constancy as well as the experimental shortage of exact data on the value of the interface temperature.
3) Errors due to the changes in the thermal properties of the mold and metal as function of temperature.

Geometrical Limitations Errors

There is no way of predicting or assessing quantitatively the error due to the geometrical factors of shaped casting apart from the actual experimental measurement. Indeed, the whole field of application of heat transfer laws to shaped sand castings depends at present on obtaining sufficient experimental evidence to understand full implications of any specific heat transfer equation. Once the stage is reached

when the limitations of simplified equations are fully understood on the basis of experimental evidence, then the comparatively simple task which will remain is that of feeding the thermal data into the appropriate equations to obtain full benefit of such data in metal founding.

Interface Temperature Errors

The question of the interface temperature raises certain separate issues:

A) Nature of the contact between the metal and molding sand.
B) Measurement and calculation of the interface temperature.
C) Temperature distribution at the interface.
D) Heat capacity of the mold.

Nature of Contact. Recent research[4] on metal penetration into sand compacts has indicated its dependence on grain size, superheat of the metal cast, the time of contact and metallostatic pressure. This study was carried out using an inert atmosphere or hydrogen, under which conditions metals do not wet silica sand. Nevertheless, the penetration measured was so considerable that under normal sand casting conditions the effect of the type of contact on the variables in the heat transfer equation cannot be neglected.

Gittus,[5] in studying the quality of gray and white-cast iron surfaces for various sand conditions, has suggested that under normal foundry conditions wetting will occur with carbon-free sands due to the preferential formation of silica-wetting iron oxide. Nonwetting occurs where pitch or coal dust is present in the sand due to preferential formation of the gaseous carbon oxides with these materials. Penetration of brass and aluminum has been studied statistically[6] under normal foundry conditions. Penetration of brass cast at 1295 C (2363 F) was found to be considerable and to depend upon liquid pressure, sand grain size, density and moisture.

However, no penetration was observed with an aluminum-5 per cent silicon alloy cast at 705 C (1301 F). This absence of penetration was explained as resulting from the low running quality of the material at these temperatures. Reference was made to similar work[7] on commercially pure aluminum. It seems doubtful that this applies to all light alloys when poured with their usual superheats.

In addition to the sand grains immediately in contact with the cast metal, attention must be directed to the adjacent sand grain layers. The expansion of sand compacts has been studied by Atterton,[8] and appears to hold values between 1.6-1.8 per cent (linear per cent) for temperature between 600 (1112) and 1000 C (1832 F). This expansion is attributed to the ballooning of the interstitial air spaces. Measuring sand movements in actual molds by spherical ended silica probes, Gittus[9] has shown how such an expansion of interstitial air spaces proceeds in the presence of solidifying cast iron.

The direction of the expansion varies with time after pouring, and also with the sand conditions, that is, whether the sand is green or dry. Dry sand first expands into the skin of solidified metal, since the rigidity of its cooler layers is greater than that of the thin solidified metal skin. As the casting solidifies further, and the precipitation of the eutectic begins, the metal expands out into the sand. In green molds the cooler portion of the sand at first has a lower rigidity, consequently an outward movement of the hot sand takes place.

This continues until the rigidity of the drying sand is sufficient to change the direction. Hereafter the sand moves in on to the casting until the expansion from the eutectic precipitation takes place.

To summarize, the available evidence confirms the assumption of good thermal contact between metal and mold with the probable exception of two cases:

1) When casting alloys of pronounced oxide film character (e.g., aluminum alloys especially when cast at low temperatures).
2) When casting in green sand molds during the period of drying out of the layers of sand adjacent to the interface.

With castings made from high-melting-point temperature alloys the second factor is small, and consequently there remains only the former factor to be considered. Paschkis,[10] using the electrical analog method of calculation of heat transfer, has computed the effect of a probable air gap in the solidification time of aluminum castings. He shows that with an air gap formation the solidification time is altered by about 44 per cent.

Unfortunately, no details of time of formation or thickness of air gap are given, so that this result must be regarded only as an indication of the order of magnitude of the effect. This effect is probably far more pronounced at low temperatures when heat transfer by radiation from a metal surface is not of the same order as the heat transfer through sand by conduction.

Calculation of the Interface Temperature. One available relationship for calculating the interface temperature is from Riemann.[11] This equation for the temperature θ_i', attained at the interface of plane semi-infinite bodies initially at temperatures θ_1, and θ_2, is

$$\theta_i' = \frac{b_1\theta_1 + b_2\theta_2}{b_1 + b_2} \qquad (2)$$

where b_1 and b_2 are the respective mean heat diffusivities. Halbart[12] approximates this to

$$\theta_i' = \frac{b_1\theta_1}{b_1 + b_2}$$

In addition, Halbart gives the following relationships for copper and cast iron:

$$\theta_i' = 0.942\,\theta_1 \text{ for copper.}$$

$$\theta_i' = 0.868\,\theta_1 \text{ for cast iron.}$$

Neither Riemann's nor Halbart's relationships take account of the fact that during casting the mold is part of the time in contact with liquid and part with solid metal, and will thus give only an approximate result for the interface temperature. The existence of

pre-solidification and solidification periods of heat transfer could be taken into consideration in the following way.

An average value $\bar{\theta}_1$ might be used instead of θ_1 by considering solidification in two parts:

A) Mold in contact with a liquid mass at temperature
$$\theta_1' \text{ for } t_1' \text{ sec.}$$

B) Mold in contact with a solid mass of average temperature
$$\theta_1'' \text{ for } t_1'' \text{ sec.}$$

The times of contact would hold the ratio:

$$\frac{t_1'}{t_1' + t_1''} = \frac{q_1^2}{q_t^2}$$

where q_1 is sensible heat of casting.
q_t is total heat of casting.
i.e.,

$$\frac{t_1'}{t_1' + t_1''} = \frac{(c\Delta\theta)^2}{L^2 + 2Lc\Delta\theta + c^2\Delta\theta^2} \qquad (3)$$

Thus, from the ratio t_1'/t_1'' a weighted mean value $\bar{\theta}_1$ to replace the quantity θ_1, in equation (2), is obtained.

In addition, we write b_1 as

$$b_1' = \sqrt{k_1 \rho_1 \left(c_1 + \frac{L_1}{\theta_m}\right)}$$

$\bar{\theta}_1$ is then calculated as

$$\bar{\theta}_1 = \frac{\bar{\theta}_1 b_1' + \theta_2 b_2}{b_1' + b_2} \qquad (2b)$$

The values of $\bar{\theta}_1$ have been obtained experimentally by Ruddle and certain other workers,[15] and will be discussed more fully in subsequent paragraphs.

Temperature Distribution at the Interface. In the two preceding sections the nature of the interface and the possibility of calculating the interface temperature have been considered.

These features are intimately connected with the temperature distribution at the mold-metal interface. The type of temperature distribution assumed greatly affects the accuracy of calculation of solidification time. This particular problem has been examined by Girshovich and Nekhendzi,[16] who consider the following cases:

1) Exact gaussian distribution across the interface.
2) Straight line distribution in the solidifying casting, taking account of the thermal capacity of the solid metal.
3) As (2) above, but ignoring thermal capacity.
4) Ignoring temperature gradient in solidifying casting.

The errors involved in the mold constant used in the calculation of rate of solidification vary according to the distribution assumed. For steel in green sand molds the most serious variation 7 per cent is in (4). Smaller errors of 2 or 6 per cent are achieved with (3) when calculated from the distributions obtained in the metal or the mold, respectively. Identical results come from (2) and (1) (the correct distribution).

The errors incurred with steel cast in metal molds are more serious. Case (4) (from mold only) 200 per cent; case (3) 27 and 39 per cent; case (2) 11 and 9 per cent, as calculated from the distributions in the metal and mold, respectively. It is worth noting that changes of 10, 30 and 200 per cent in the mold constant would lead to changes of 18, 41 and 89 per cent in solidification time, all other things being equal. For present purposes, i.e., sand molds, the above errors are of a smaller order than many of those arising from the use of thermal constants of doubtful accuracy, as well as from other assumptions.

Heat Capacity of Mold and Air-Mold Interface. In addition to the mold-metal interface, the external or air-mold interface must be considered.

Chvorinov,[1] Ruddle[14] and many others have shown that during the solidification of sand castings of thickness not greater than that of the surrounding molding sand, the external mold face temperature does not rise significantly over that of the surroundings.

The antithesis of this condition is provided by shell molds, the outer surface temperature of which was found to increase appreciably during the solidification period, according to Pellini and co-workers.[17]

Metal and Mold Thermal Properties Errors

Bearing in mind the scarcity of data in this field, one must carefully consider the important assumption of a constant value for the following:

A) Thermal conductivity to the mold k_2.
B) Density of the mold ρ_2.
C) Specific heat of the mold c_2.
D) Specific and latent heats of the liquid metal c_1 and L_1.

It is a well-understood principle that conductivity and specific heat of continuous solids vary considerably with temperature. Consequently, in heat-flow problems a mean value is taken appropriate to the type of dependence. For most purposes a simple integrated mean is quite satisfactory. Russell[18] and Eucken[19] have pointed out the complexity of the nature of heat flow through porous media, so that the foregoing treatment must be performed graphically on the best set of data available, unless a measured mean value of k_2 has been obtained for the temperature concerned.

Similar remarks regarding complexity apply to the dependence of both c_2 and ρ_2 on temperature, although in the latter case it is generally considered permissible to insert the room-temperature value of ρ_2 in calculating b_2.

Shape and Size Variables

In general, shaped castings produced in sand molds and similar materials are of extremely varying design. External and internal surfaces may be of such a complicated form that it is impossible to reduce

such a casting into a number of simple geometric shapes. Similarly, the weight of the casting may vary from a few ounces to several tons. In general, it would be desirable to predict the solidification times of such complicated castings, because such castings are most difficult to produce and lead to high losses if defective. With a clear picture of the solidification rate and mechanism, one would expect that a more rational approach to the casting problem would reduce the number of defective castings.

An approach to this problem must necessarily proceed along the line of studying the solidification times and rates of simple geometric shapes. In reviewing the available data on the heat transfer in simple shaped castings in sand molds, the general equations of heat transfer in plates, cylinders and spheres will be summarized.

Plates. For a wide plate casting, the area of whose main faces is A, and the area of whose end or minor faces may be neglected, the end point of solidification t_s is obtained from

$$q = A \cdot \frac{2}{\sqrt{\pi}} (\theta_i - \theta_0) b_2 \sqrt{t_s} \qquad (1a)$$

equating this with the heat lost during solidification we have

$$\rho_1 V [L_1 + c_1(\theta_1 - \theta_m)] = A \cdot \frac{2}{\sqrt{\pi}} (\theta_i - \theta_0) b_2 \sqrt{t_s}$$

$$\frac{V}{A} = \left(\frac{2(\theta_i - \theta_0) b_2}{\sqrt{\pi} \rho_1 [L_1 + c_1(\theta_1 - \theta_m)]} \right) \sqrt{t_s}$$

$$\frac{V}{A} = k \sqrt{t_s} \qquad (4)$$

where V is the volume and θ_1 is the casting temperature (assuming that there is no drop in temperature during pouring)

θ_0 is the initial temperature of the mold.
θ_m is the solidification temperature.
L_1 is the latent heat of solidification.

The validity of this equation further depends upon the assumption that the sensible heat $\rho_1 V c_1 (\theta_1 - \theta_m)$ is liberated with θ_1 constant, so that this former quantity may thus be added to the latent heat. The quotient V/A is generally written R and therefore we obtain

$$t_s = q \cdot R^2 \qquad (5)$$

Alternatively one can apply equation (1) to determining the depth d_t to which the casting has solidified at a time t.

$$q_t = A \frac{2}{\sqrt{\pi}} (\theta_i - \theta_0) b_2 \sqrt{t}$$

$$(d_t A) \rho_1 [L_1 + c_1(\theta_1 - \theta_m)] = \frac{2A}{\sqrt{\pi}} (\theta_1 - \theta_0) b_2 \sqrt{t}$$

$$d_t = \left(\frac{2(\theta_i - \theta_0) b_2}{\rho_1 [L_1 + c_1(\theta_1 - \theta_m)]} \right) \sqrt{t} \qquad (4b)$$

$$d_t = k' \sqrt{t} \qquad (4c)$$

where κ' is termed the mold constant.

It was Chvorinov[1] who first suggested that equation (5) holds good for many castings, if external curvature and end effects are neglected. In order to substantiate this, he showed a double logarithmic plot of

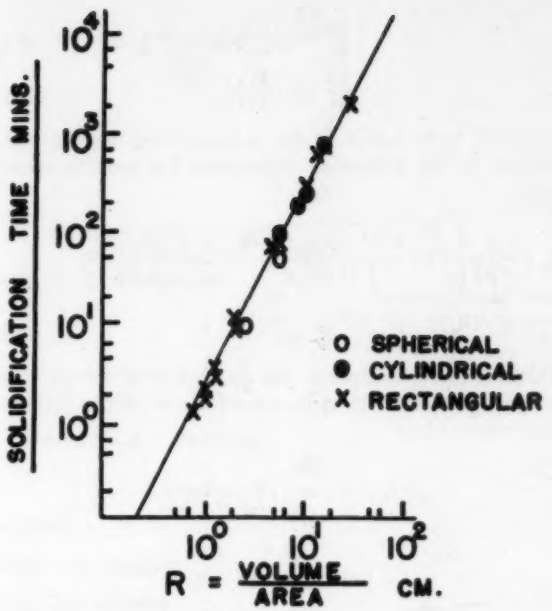


Fig. 1-1 — Relationship of volume/surface ratio to solidification time.[1]

R and t_s (Fig. 1-1) which revealed a good agreement for steel castings of various sizes and shapes. The results included certain data of Briggs and Gezelius.[20]

The value of q was derived in two ways, both apparently in good agreement. The first method employed was that previously indicated in equation (1a), that is, from the thermal constants involved. The second involves the calculation of q as defined by Schwarz's solution:[21]

$$L \rho_1 \sqrt{\frac{\pi}{4}} q = \frac{b_2 e^{-q^2/4a_s} \theta_m}{\frac{b_s}{b_2} + G(q/\sqrt{4a_s})} - \frac{b_2 e^{-q^2/4a_1} \Delta\theta}{1 - G(q/\sqrt{4a_1})} \qquad (6)$$

where θ_m is the temperature of solidification.
$\Delta\theta$ is superheat, i.e. $\Delta\theta = \theta_1 - \theta_m$
and subscript
1 Indicates liquid metal.
s Solid metal.
2 Mold.

Unfortunately, the assumptions of neglecting curvature and corner effects have not always been recognized by subsequent investigators. In examining the general application of equation (5), Ruddle and Skinner[22] have shown that a correction can be made for corner effects. The correction consists of adding to the surface area unaffected by the corners the equivalent plane surface area of the corners. This equivalent area is obtained as the product of a semi-empirically derived factor and the actual surface area of the corners.

Spheres and Cylinders. Regarding shapes other than flat plates, few workers have attempted to derive the fundamental equations describing either progress of solidification or total solidification time.

Halbart[12] has examined the case of the sphere, in which the basic differential equation for unsteady heat flow was translated into the form:

$$\frac{\partial \theta}{\partial t} = a_2\left(\frac{\partial^2 \theta}{\partial x^2}\right) \qquad (7)$$

$$\frac{\partial x}{\partial t} = a_2\left(\frac{\partial^2 x}{\partial r^2}\right) \qquad (7a)$$

where $X = \theta r$ and r is the radius of the sphere. He arrives at the following expression for solidification time:

$$t_s = \frac{9}{\pi a_2^2}\left[\sqrt{1 + \frac{\pi \rho_1 [L_1 + c_1(\theta_1 - \theta_m)] - 1}{3\rho_2 c_2 \theta_1}}\right]\left(\frac{V}{A}\right)^2 \qquad (8)$$

$$t_s = q' \cdot (R_s)^2 \qquad (8a)$$

Halbart then compares the quantity of q^1 in the above equation with q in equation (5) in the following way

$$\sqrt{t_s} = A' \cdot \frac{V}{A}, \text{ for sphere.}$$

$$\sqrt{t_s} = B' \cdot \frac{V}{A}, \text{ for plate.}$$

where $A' = \sqrt{q^1}$ and $B' = \sqrt{q}$
and that shows B′ and A′ are connected in the following way:

$$B' = A' + \frac{\sqrt{\pi}}{\sigma} a_2 A'^2$$

which, for sand molds quoted with $a_2 = 0.07$, becomes: $B' = A' + 0.02\,A'^2$

B′ and *A′* are given for the cases of copper and cast iron in sand molds:

Metal	B′	A′
Copper	6.8	6.1
Cast iron	5.9	5.3

Thus it can be readily shown that, since $(A')^2$ cannot be negative, a plate will solidify in a longer time than a sphere of equal volume-to-surface ratio. Trencklé[23] has used forms of the equations to calculate the permissible pouring times of various shapes. The other shapes he considered were examined using theoretical work on temperature distribution.[25-27]

Mold Temperature Distribution

Abcouver[24] takes a rather different approach to calculation of solidification time of any given shape. He presents a single equation describing the temperature distribution in the mold, and the casting which takes into account the temperature discontinuity at the mold-metal interface. Exactly how the equation has been derived is not specified. However, it is used to determine total solidification time t_s through knowledge of volume-to-surface ratio and the usual thermal constants:

$$\frac{b_2}{\pi(b_1 + b_2)}\sqrt{4\pi a_2 t_s \left(\frac{A}{V}\right)^2} = \frac{L_1}{c_1\theta_1} + \frac{H}{H+1} \qquad (9)$$

where $H = \dfrac{\theta_1 - \theta_m}{\theta_m}$ and $\dfrac{L}{C_1\theta_1} + \dfrac{H}{H+1} = 0.35$ for cast iron.

Apparently, this equation may be applied to any simple-shaped casting. Chvorinov,[28] in his most recent paper, accounts for the effect of casting variables L_1 and θ_1 on the value of κ' in the equation:

$$d_t = k'\sqrt{t} \qquad (4c)$$

If the equation is divided by *R*, (the ratio V/A) he suggests that for conditions of superheat (U)

$$U = \theta_1 - \theta_m$$

$$\frac{d_t}{R} = \frac{k'_o}{R}\sqrt{t} - \frac{Uc_1}{RL_1} \qquad (10)$$

may be written for the plate, where κ'_o is the value of κ' at $\theta_1 = \theta_m$. This gives a straight-line relationship of $\dfrac{d_t}{R}$ to $\dfrac{\kappa'_o}{R}\sqrt{t}$ for this particular body. Chvorinov extends this to cylindrical and spherical castings by assuming that heat transferred over a given area is identical in all three basic shapes.

Experimental Evidence

It is simple to show by geometry then that for spheres,

$$d_s = d_p\left(1 - \frac{d_p}{r}\right) \qquad (11)$$

for cylinders,

$$d_c = d_p\left(1 - \frac{d_p}{2r}\right) \qquad (12)$$

where d_p is d_t as derived for plates by equation (10). Thus, he shows that three families of curves can be drawn—one linear, and two parabolic—each curve a family of different superheats. Experimental evidence is given to confirm these relationships for steel castings, and shows fairly good agreement. The relationship of $\dfrac{d}{R}$ to $\dfrac{t}{R^2}\kappa'^2_o$ is also provided, and appears to be a better means of expressing the result. Figure 1-2 shows a set of curves relating $\dfrac{d}{R}$ to $\dfrac{t}{R^2}$, a particular case of this latter relationship.

The present authors feel that it would be worthwhile to examine the basic expressions for spheres and cylinders in the hope that these expressions might possibly be approximated to the simple form of equation (5).

Jaeger[29] gives the following relations for $\dot{q}$, the rate of heat extraction for unit area for long cylinder and spheres with radius *a* as:

For long cylinders:

$$\dot{q} = k_2(\theta_i - \theta_0)\frac{4}{a\pi^2} I\left(0, 1, \frac{a_2 t}{a^2}\right) \qquad (13)$$

For spheres:

$$\dot{q} = k_2(\theta_i - \theta_0)\left[\frac{1}{\sqrt{\pi a_2 t}} + \frac{1}{a}\right] \qquad (14)$$

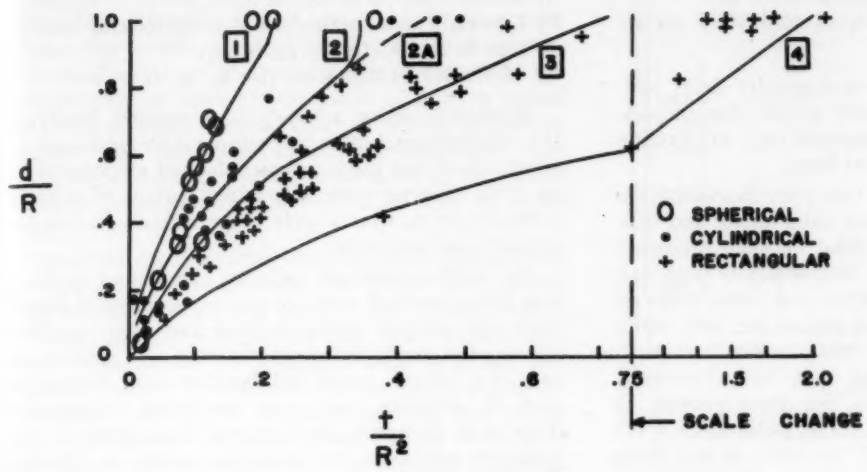


Fig. 1-2 — Theoretical and measured values for the relationship d/R and d/R².

d – DEPTH SOLIDIFICATION
R – RATIO V/A
t – TIME

1 SPHERES
2 SHORT CYLINDERS
2A LONG CYLINDERS
3 RECTANGULAR SHAPES
4 WIDE PLATES

However, for small values of $\frac{\alpha_2 t}{a^2}$ equation (13) becomes

$$\dot{q} = k_2(\theta_i - \theta_0)\left[\frac{1}{\sqrt{\pi a_2 t}} + \frac{1}{2a}\right] \qquad (15)$$

Since values $\frac{\alpha_2 t}{a^2}$ are small for most sand castings poured with normal amounts of superheat, the use of equation (15) may be permitted. Equations (15) and (14) are written in a general form:

$$\dot{q} = k_2(\theta_i - \theta_0)\left[\frac{1}{\sqrt{\pi a_2 t}} + \frac{1}{RC_1C_2}\right] \qquad (16)$$

Since

R = a/3 for sphere.
R = a/2 for long cylinder.

Where

$C_1 = 3$ for sphere.
$C_1 = 2$ for cylinder.
$C_2 = 1$ for sphere.
$C_2 = 2$ for cylinder.

If we now consider q the heat flowing across an area A in time t, we have an integrating equation (16)

$$q = \sigma b_2 \sqrt{t} A(\theta_i - \theta_0) + \frac{k_2 t(\theta_i - \theta_0)}{C_1 C_2 R}$$

(where $\sigma = 1.13$)

again as before:

$$q = V \cdot \rho_1 [L_1 + c_1(\theta_i - \theta_m)]$$

or writing $q = V \cdot \rho_1 \cdot \bar{q}_1$

we have:

for t_s $\quad \sigma b_2 A\sqrt{t_s}(\theta_i - \theta_0) + \frac{k_2 t_s A(\theta_i - \theta_0)}{RC_1C_2} = V \cdot \rho_1 \cdot \bar{q}_1$

or $\quad \sigma b A\sqrt{t_s} v + \frac{k_2 t_s A}{RC_1C_2} \cdot v = V \cdot \rho_1 \cdot \bar{q}_1$

Dividing throughout by V and writing $\frac{V}{A} = R$ we have:

$$\sigma b_2 \sqrt{t_s}\frac{v}{R} + \frac{k_2 t_s v}{R^2 C_1 C_2} = \rho_1 \bar{p}_1$$

It is eventually found that for cylinders and spheres:

$$t_s = \frac{R^2}{\left[\frac{\sigma b_2 v}{2\rho_1 \cdot \bar{q}_1} + \sqrt{\frac{v}{\rho_1 \bar{q}_1}\left(\frac{\sigma^2 b_2^2 v}{4\rho_1 \bar{q}_1} + \frac{k_2}{C_1 C_2}\right)}\right]^2} \qquad (18)$$

Equation (5) may be written in a similar manner for plates:

$$t_s = \left(\frac{\rho_1 \bar{q}_1}{\sigma b_2 v}\right)^2 R^2 \qquad (19a)$$

Thermal Properties of Casting and Molding Media

An attempt to predict solidification time of a casting, whether by calculation or by electrical analog methods, depends for its accuracy on the knowledge available on the values of the thermal properties of the casting and molding media. The basic properties required are:

A) Thermal conductivities of mold, liquid and solid metal, k_1, k_s and k_2.
B) Specific heats of mold, liquid and solid metal, c_1, c_s, c_2.

C) Latent heat of fusion of liquid metal L_1.
D) Densities of the mold, liquid and solid metals ρ_1, ρ_s, ρ_2.

The appendix of Ruddle's monograph[2] deals adequately with such data available at the time of publishing (1950),* hence, it is proposed only to examine similar data published since that time.

In the field of metals few new contributions have since been made. This is indeed unfortunate and may account for discrepancies obtained between predicted and experimentally obtained solidification time values. The determination of specific and latent heats of metals and alloys necessitates expensive and often complicated apparatus, particularly when high-melting-point materials are to be examined. Successful methods have been developed for measurement of specific heats by the cooling curve method for solid metals. However, this method has only of late been applied to liquid metals, and much work remains to be done.

Latent heat value measurements have been even more neglected. In some cases, thermodynamic methods can be utilized to obtain latent heat values. Ruddle and Mincher[30] have given values for several materials which have been calculated from the density of the solid at the melting point.

Thermal Constants

Abcouver[24] and Halbart[12] have given some values of thermal constants of metals which appear to have been determined experimentally, but no data have been given concerning the accuracy of the figures. Until this dearth of information on thermal data of metals has been dealt with, the progress of many heat flow calculations in many metallurgical fields will be delayed.

Some strides have been made recently in the field of measurement of thermal properties of molding materials. It is an advantage to classify the methods of determination as follows:

A) Steady state methods (for k_2, α_2).
B) Unsteady state methods, involving making of castings in molds (for k_2, α_2, b_2, c_2).
C) Methods of calculation (for k_2, α_2, b_2).

Atterton,[31] using a steady state method involving the measurement of radial temperature gradients in small cylindrical sand compacts heated electrically by an axial heating wire, determined values of thermal conductivity k_s over a wide range of temperature for several mold materials.

The sands examined included coarse and medium fine sands, bonded with various percentages of bentonite, and samples made at three ramming densities. Among the materials examined were also glacial, zircon and olivine sands (all bonded with bentonite) and, in addition, two riser insulating compounds. This work shows the fundamental importance of temperature, grain size and ramming density in affecting the thermal conductivity. However, in applying Atterton's results to solidification problems, care must be taken that a correct mean value is used.

Using unsteady state methods of determination involving melting and casting, many recent experiments have been reported by a number of workers. Ruddle,[14] Ohira,[32] Halbart,[12] Abcouver[24] and Briggs and Locke,[33] as well as the present authors.[34] Ruddle[14] used the method of Chvorinov for determining α_2 the mean temperature diffusivity for the following materials:

1) Fine silica sand, (Erith), in dry state.
2) Naturally-bonded sand (Mansfield), in dry state.
3) Bonded silicon carbide grit.
4) Bonded magnesite.

Interface Temperatures

The determinations were carried out for various interface temperatures, and the results may be readily applied as mean values for relevant cases. Both this method and a mold calorimetric method have been used by the present authors on behalf of a subcommittee of the Institute of British Foundrymen, to determine b_s for the casting temperature of several alloys. This second method consists of determining b_s, the mean heat diffusivity, by equating heat liberated on solidification of the casting with that absorbed by the mold in the same time.

Determinations of b_s were carried out on a coarse and a medium-fine sand, so that a comparison can be made with the appropriate figures calculated from Atterton's results for two fairly similar sands from the same geological source (Fig. 1-3). The order of the values of b_s is in reasonable agreement; however, the difference in slopes is difficult to explain. Halbart,[12] Trencklé[23] and Abcouver[24] also appear to have derived values of thermal conductivity for sands from their experimental work, all of similar order to the other investigators work.[31-34] However, they[12,23,24] give no details of the method used or the accuracy of the results.

Ohira also derived similar values for this quantity for a naturally bonded sand in the green and dry state. He has, in addition, determined the mean specific heat of these sands by calorimetric means. Ohira's

*A 1957 edition of this work is now available.

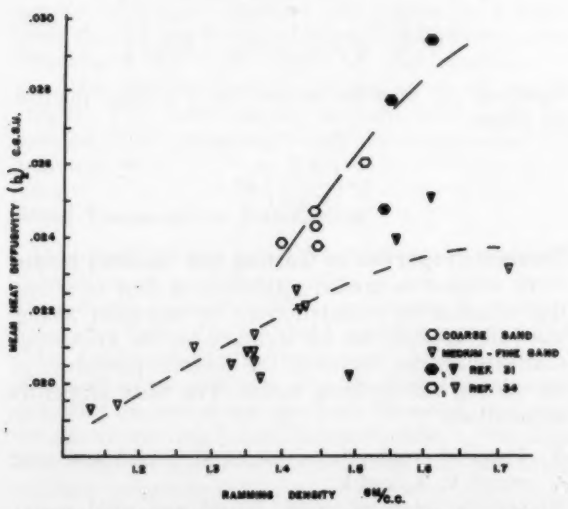


Fig. 1-3 — Variation of mean heat diffusivity b_2 with ramming density.

data show good agreement with previous results (Fig. 1-4). Recently Briggs and Locke[33] have published the results of a series of tests to determine the chilling power of a wide range of materials. Unfortunately, no attempt was made to put the results in order of mean thermal conductivity or diffusivity. Results were merely presented as solidification time of a 6-in. diameter sphere poured in a low-carbon steel at a temperature of 1576 C (2867 F) of superheat.

With the use of existing data for mean specific heat of the mold materials and for latent and specific heat of steel, the present authors have derived values of mean themal conductivity and mean thermal and heat diffusivity of the materials used by Briggs and Locke (Table 1-1). The accuracy of these results depends not only on the experimental error but on the data used in the calculation, and so may only be of the order of 10-15 per cent in certain instances. Ohira[32] and work by Kondic and Martin, for the subcommittee of the Institute of British Foundrymen previously mentioned, are the only sources of thermal constants for green sands. Herein lies the advantage of unsteady state methods, as values of the thermal constants for green sands can be given if only approximately.

Green Sand Chilling Power

Although Ruddle and Mincher[14] give the temperature distribution for green sand, no value of b_s or α_2 could be obtained using Chvorinov's method. However, the application of a corollary of Schmidt's graphical method[35] might resolve this difficulty. Paschkis,[36] using the electrical analog with Tanasawa's data[37] to study the effect of moisture content in heat extraction, points out that the presence of moisture has little effect after a critical time for a particular interface temperature. Thus, the chilling power b_s, of a green sand would appear dependent on the size of the casting made because of the variable nature of the magnitude of drying out of layers of sand adjacent to the casting.

However, the fact that no value of mean interface temperature has been given to Tanasawa's curves raises the question of the accuracy of Paschkis' results. Although temperature distributions in cores have been

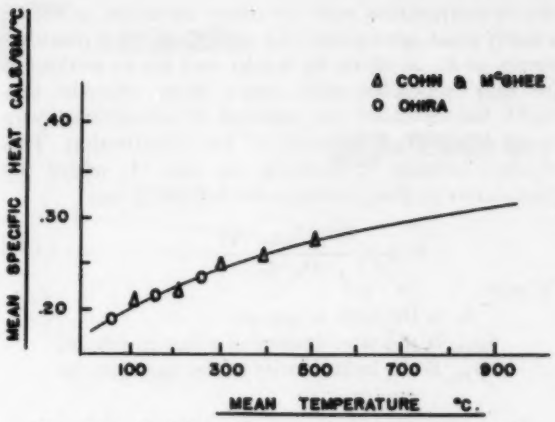


Fig. 1-4 — Variation of mean specific heat of a molding sand with mean temperature of the mold.

measured with good accuracy by Chamberlain and Peters,[38] no one has attempted to determine their thermal properties. Core sand temperature distributions during heating, like those of green sand molds, contain discontinuities which make theoretical analysis difficult.

The method of calculation of thermal properties is due mainly to Russell[18] and Eucken,[19] who have shown that for porous refractory materials such as sand, the observed thermal conductivity of the aggregate k_a depends only upon the apparent conductivity of the pores k_p and the porosity fraction P;

$$k_a = k_p\left(\frac{3}{P} - 2\right) \qquad (20)$$

Russell[18] also gives

$$k_p = k_t + (5.44 \times 10^{-12})\, T^3 X \qquad (21)$$

where

k_p is in cgs units.
k_t is the thermal conductivity of air in cgs units.
T is the absolute temperature in °K.
X is the pore space diameter in cm.

Bell[39] gives an expression for pore space diameter for sands of a two-sieve type distribution and has used

TABLE 1-1 — THERMAL PROPERTIES OF SPECIAL MOLDING MATERIALS*

Materials	Mold Density, gm/cc	Mean Specific Heat, cal/gm/°C	Mean Thermal Conductivity, cgs	Mean Temperature Diffusivity, cgs	Mean Heat Diffusivity, cgs units	Temperature Range C	Temperature Range F
Graphite	2.0	0.35	0.00294	0.00420	0.0454	0-1470	32-2678
Cu-shot	4.4	0.1	0.00338	0.00768	0.0386	0-1470	32-2678
Steel-shot	3.74	0.16	0.00176	0.00294	0.0325	0-1470	32-2678
Silicon carbide grit	3.7	0.24	0.00119	0.00135	0.0325	0-1470	32-2678
Magnesite grit	1.67	0.25	0.00206	0.00494	0.0293	0-1470	32-2678
Alumina grit	1.9	0.30	0.00145	0.00254	0.0288	0-1470	32-2678
Chrome ore	2.14	0.27	0.00130	0.00225	0.0274	0-1470	32-2678
Chamotte	1.48	0.28	0.00144	0.00348	0.0244	0-1470	32-2678
Alumina (bubbled)	0.92	0.30	0.00146	0.00529	0.0201	0-1470	32-2678
Diatomaceous earth 40% coarse sand 60% (mulled)	1.11	0.28	0.00116	0.00373	0.0190	0-1470	32-2678
Diatomaceous earth 40% coarse sand 60% (unmulled)	0.78	0.28	0.00086	0.00394	0.0137	0-1470	32-2678

*Mean thermal conductivity, temperature diffusivity and heat diffusivity derived from results of Briggs and Locke.[33]

this in conjunction with the above equation to obtain a fairly good agreement with the steady state measurements of k_s, as made by Lucks and his co-workers[40] for four unbonded silica sands. More recently, Gittus[41] has extended the method of calculating pore space diameter X to sands of any distribution. This method consists of utilizing the area A_c under the cumulative grading curve in the following way:

$$X = \frac{200\,[\rho_{SiO_2} - \rho_B]}{3A_c \cdot \rho_B} \qquad (22)$$

Where

A_c is the area in gm cm.
ρ_{SiO_2} is the true density of silica in gm/cc.
ρ_B is the bulk density of the aggregate in gm/cc.

Using this method Gittus has obtained relationships of thermal conductivity to the cube of the absolute temperature for the unbonded sands examined by Lucks and his co-workers.[40]

The agreement between these results and those of the experiments is good for all the sands, except the coarse one, where the disagreement varies between 0 and 10 per cent. The same worker has found that the results of Atterton[31] for a bentonite bonded coarse sand (52 A) may be predicted with good accuracy. However, in the case of the fine sand investigated (26 A), considerable disagreement was found. This was thought to be due to the fact that clay bond increases the contact area between the small-sized grains and so invalidates the previous equation.

A PRELIMINARY STUDY OF INTERFACE TEMPERATURE

An examination of the simple theory of heat transfer, as applied to sand castings and described in the previous paragraphs, indicates the necessity for an accurate value of interface temperature for any calculation of solidification time.

It was decided, in the absence of a thoroughly reliable theoretical method of predicting the value or range of values, that this quantity would be experimentally obtained in castings of simple shape and of various sizes cast in three different alloys. This was done in order that convenient mean values might be derived for use in any subsequent calculations.

Although Ohira[32] has considered interface temperature as a function of time in determining the thermal conductivity of the mold, it was anticipated that any extension of this method would be arithmetically laborious.

The experimental measurement of interface temperature itself presents certain difficulties. The location and nature of the mold-metal interface itself is somewhat obscure. Several workers such as Pellini,[42,43] Gittus and Hughes[44] and Middleton and Protheroe[45] have defined it as the temperature indicated by means of a butt-welded, silica-sheathed thermocouple lying at the interface half within the mold and half within the cavity. The axis of such a couple was in the plane of the mold face. Ruddle[13,30] and his co-workers have preferred the use of twin-bore, refractory-sheathed thermocouples, the tips of which were lightly covered with alundum cement.

However, Atterton and Houseman[46] have shown the temperature-time relationship thus obtained differs considerably from that measured with a bare-wire unjoined thermocouple. With this latter type of couple the insulation is discontinued just before emerging through the mold face so that the two wires separated by about 2 or 3 mm eventually have as their thermo-junction the penetrating liquid or the solidified metal skin. It would appear that such a system is not subject to the thermal lag of the former methods. The use of this latter method was preferred for the present experimental work.

Experimental Work

The alloys used were:

1) 12 per cent silicon, aluminum-base alloy (unmodified).
2) 8.5 per cent aluminum, copper-base alloy.
3) Gray cast iron of carbon equivalent 4.2 per cent.

Details of the metal analyses, thermal constants, etc., appear in the Table 1-2 together with details of experiments carried out. The alloys were chosen to cover a range of interface temperature ~500 C to ~110 C (~932 to 2012 F), but at the same time are of considerable interest because of their frequent use in founding. In the cases of silumin and aluminum bronze, the temperature was recorded using 28-gage chromel-alumel thermocouples connected through a selector switch to a square-wave amplifier-recorder unit of full scale deflections 20 and 2mv with a total response time of 0.2 sec.

The major part of the thermal emf was backed off. Thermocouples at the center of the castings were insulated by twin-bore porcelain refractory along their lengths, and by a thin wash of an alumina-water suspension at their ends. To prevent metal penetration, the twin-bore tubes were sealed off with alumina.

The hot junctions of the interface thermocouples were left unjoined,[46] and were intended to measure the metal "skin" temperature only. Their lengths were also porcelain insulated. For the castings made in cast iron the interface temperatures were measured using an accurate portable millivoltmeter. This had a response time of approximately 1 sec, and was shown to be accurate to ±1 C (±2 F). Thermocouples were again insulated but were also covered by thin steel sheaths which were alumina washed along their lengths.

Without the added mechanical protection couples invariably failed. The center thermocouples were all insulated at their tips by a somewhat thicker wash of alumina than that used with the other materials. The results of these experiments will be discussed separately for each alloy.

Aluminum-Silicon alloy. A series of spheres with diameters ranging from 1 to 4⅜-in. was cast into sand molds. The method of ramming was such that the ramming density gradients were minimized in the mold. No thermal gradient could be detected in the case of the smallest diameter sphere before, during or even after solidification. In the sphere of 1¼-in. diameter a gradient of 1 C (2 F) along its radius was detected immediately prior to completion of solidification. With spheres of 2 in. diameter upward,

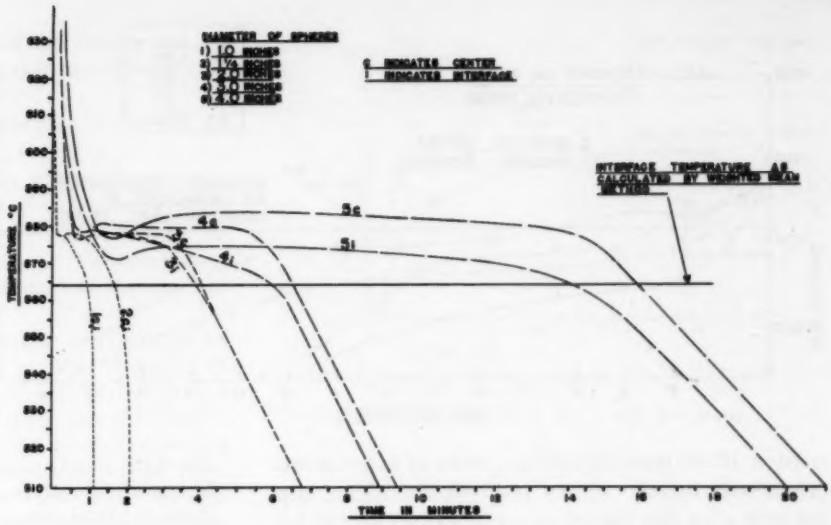


Fig. 1-5 — Temperature-time relationships in aluminum, 13 per cent silicon alloy spheres of various diameters.

radial temperature gradients were clearly discernible, and appeared to hold maximum values immediately prior to the completion of solidification.

The actual values of maximum radial temperature difference rose from 3 C (5.5 F) in the case of the 2 in. diameter sphere to about 15 C (27 F) in the case of the sphere of largest diameter (4 in.). The cooling curves obtained for the center and interface thermocouples are shown in Fig. 1-5. The approximate mean values for interface temperature cover a range of roughly 20 C (36 F). The minimum values are recorded for the largest sphere (4 in.), and the maximum for the fourth largest (2 in.). The fact that the small spheres had lower mean values of interface temperature, which was assumed to be coincident with that of the center, is due to a fundamental alteration in the shape of the cooling curve, which is no doubt due to the non-equilibrium solidification.

Cast Iron. Using a cupola-melted cast iron with a carbon equivalent approximately that of the eutectic composition, spheres of two sizes, 4⅜-in. and 3-in. diameter, were cast in green sand molds. The sand used in molding was a medium-grain naturally-bonded type. The maximum difference between interface and center temperature was not more than about 24 C (43 F) in either case. The cooling curves in Fig. 1-6 represent the averaged values of temperature meas-

TABLE 1-2 — PHYSICAL PROPERTIES OF ALLOYS INVESTIGATED

Alloy and Chemical Analysis	M.P. and Freezing Range, C (F)	Latent Heat, cal/gm	Density, gm/cc	Thermal Conductivity, cgs units	Specific Heat, cal/gm/°C	Casting Temp. C (F)	Casting Size Shape and Mold
12% Si (unmod.) remainder superpure Al., scrap remelted	584 (1085) 7 (13)	95•	2.42†	0.35••	0.259†	700 (1292)	Spheres 1 to 4⅜-in. diameter medium grade sand
8.5% Al, remainder electrolytic Cu, scrap remelted	1038 (1985) 0 (0)	55†	6.9†	0.1••	0.131†	1250 (2282)	Plate 1 x 6 x 6 in. thick fine and coarse sands
Cupola melted cast iron Average analysis: C 3.35% Si 1.98% P .53% S .11% M .70% (carbon equiv. 4.2%)	1150 (2102) (L) 1120 (2048) (S)	47²⁴	6.8††	.09‡	0.212‡‡	1300 (2372)	Spheres 2 to 4⅜-in. diameter, medium-grade sand

(L) indicates liquidus
(S) indicates solidus

•Determined calorimetrically.
†Calculated by the Neumann-Kopp rule (See A.S.M. *Metals Handbook*, 1948).
••R. Ruddle and A. Mincher, *Jnl., Inst. of Metals 78*, p. 229, 1950-51.
††C. Schwartz, AFA TRANSACTIONS, *53*, p. 1, 1945.
‡J. Donaldson, Proc. *I.B.F. 32*, p. 125 (Extrapolation) 1938-9.
‡‡S. Unimo Toh. Imp. Univ. Sci. Rep. *15*, p. 331, 1926.
NOTE: Super figure refers to reference already listed.

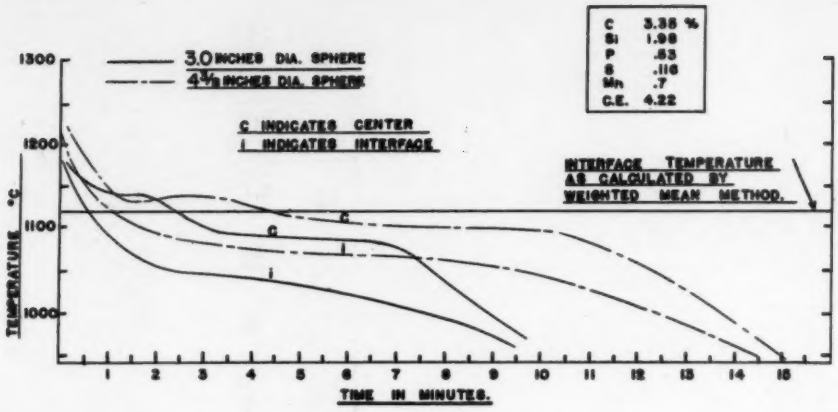


Fig. 1-6 — Temperature-time relationships of spheres in cast iron.

ured at 10 sec intervals for two series of experiments.

Aluminum Bronze. An 8.5 per cent aluminum, copper-base alloy was melted in a gas-heated crucible furnace, and cast in a variety of plate sizes between ¼-in. and 3 in. thick. Two sand types were investigated:

1) Fine sand, soft-rammed.
2) Coarse sand, hard-rammed.

Two spheres, 2-in. and 4⅜-in. diameter, were also cast. For the two thinner plates, no difference in center and interface temperature could be detected at any time during solidification. The experiments with the 1-in. thick plates indicated a maximum difference of 20 C (36 F) for both mold conditions (Fig. 1-7). Similarly, both mold conditions in the instances of the 3-in. block castings showed identical maximum temperature differences. It would seem that a maximum value of temperature is reached for only a small increase in casting size with this alloy.

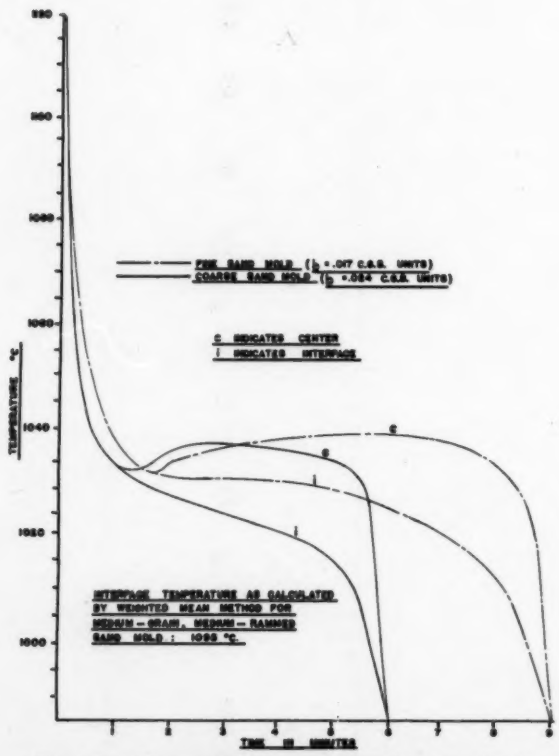


Fig. 1-7 — Temperature-time relationships for aluminum bronze (8.5 per cent Al) plates 1 in. x 6 in. x 6 in., cast in molds of differing chilling power.

DISCUSSION OF RESULTS

The assumption of a constant, instantly attained value of interface temperature is only true for a limited range of casting sizes. The magnitude of this value appears to be always below the solidification temperature and not above it as predicted by the Riemann equation (2). Other proposed equations do, however, give a little better agreement. The increasing difference between interface and center temperature with increasing casting size is due to the fact that the solidified surface, which is cooling continuously below the solidification range, is influenced by the thermal blocking effect of the slowly solidifying center in varying degrees according to casting size.

Thus, in the instance of a large casting, thermal blocking of the center has less effect on the surface cooling. The obvious inference from this is the dependence of temperature difference between the center and interface on solidification temperature and temperature diffusivity of the metal cast. Considering the results for all the materials it might be thought that variation of maximum temperature difference depends upon 1) the casting size, 2) thermal diffusivities and 3) the difference between the initial mold temperature and some temperature lying between casting and solidification temperature of the alloy.

Thus, maximum temperature difference, for a particular mold condition results from a small increase in section with a casting medium of low diffusivity solidifying at a high temperature, or from a large increase in section with a casting medium of high diffusivity solidifying at a low temperature. Expressing this in the form of an equation for a particular type of sand mold, we obtain:

$$\Delta\theta_{MAX} = f\left[x'(\bar{\theta}_1 - \theta_0)\frac{a_2}{a_1}\right] \qquad (23)$$

where

$\Delta\theta_{max}$ is the maximum temperature difference.

X is the shortest distance between the center and the major surface of the casting.

$\bar{\theta}_1$ is the weighted mean temperature between pouring and solidification.

θ_o is the temperature of the mold.
α_1 is the mean temperature diffusivity of the metal.
α_1 is the mean temperature diffusivity of the mold.

The temperature $\bar{\theta}_1$ is taken as the weighted mean described previously. A general correlation between $\Delta\theta_{max}$ and the factor $[X\,(\bar{\theta}_1 - \theta_o)\colon \alpha_2/\alpha_1]$ might be expected. Figure 1-8 gives evidence of the correlation, the scatter occuring is probably due to the not entirely reliable values of α_1 used.

In addition to the present results, certain data by Ruddle and Gittus and Hughes are also included in Figure 1-8. A possible second source of deviation of a lesser magnitude is to be found in the fact that the actual center temperature often drops gradually during solidification. Apart from the magnitude of the maximum temperature difference between the center and interface, the change of the shape of the interface temperature cooling curve with increase in section size is also to be considered.

Ruddle,[30] in investigating several eutectic alloys, suggested that three shapes of such a curve occurred, two of which were probably typical of eutectic alloys whose mechanism of solidification was not that of the skin formation generally associated with pure metals and eutectic alloys. However, since the shape of the interface temperature cooling curve approximates to each of these shapes in turn as the casting size is increased, it would appear that a physical interpretation in terms of temperature, diffusivities and section thickness is more appropriate than a metallurgical one.

GENERAL CONCLUSIONS

For the purpose of applying simple heat flow theory to the progress of solidification and total solidification time in simple sand-cast shapes, certain values of the thermal constants involved must be available. In addition, it is necessary to make certain assumptions so that the basic equations are applicable.

In the preceding section the validity of the assumptions has been discussed and, in one important case, examined experimentally, while the sources of available thermal constants have been collected. The present position in the application of the heat flow would appear to be governed by:

1) Assumption of constant, instantly attained interface temperatures. Here it has been shown that only an averaged value can be applied.
2) Inaccuracy introduced through the use of approximate thermal constants. Here there is little data available and a necessity for further work.
3) Inaccuracy introduced through several sources of heat loss, such as gates, flashes, mold corners, etc.

The first point, that of interface temperature, is indeed important. In view of the few attempts to apply theory in this field, mean values derived from experiment must be relied upon. Regarding the second point, only continued research involving costly apparatus can yield more data in the field of metals and alloys; data not only necessary in the field of solidification but also in other branches of metallurgy. The

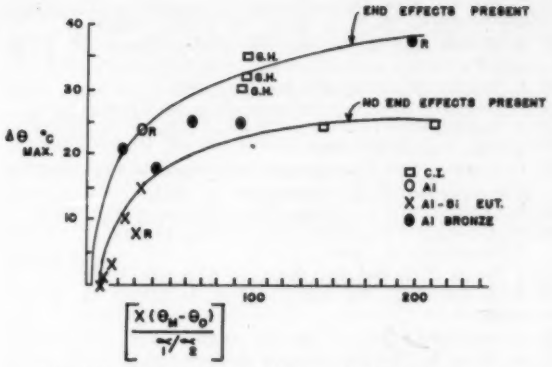


Fig. 1-8 — Correlation of $\Delta\theta_{MAX}$ and the group

$$\left[\frac{X\,(\bar{\theta}_1 - \theta_o)}{\alpha_1 / \alpha_2}\right]$$

third and final point borders on the shaped sand casting. The mathematical analysis of all but the most simple shapes in terms of heat flow in sand molds often results in the most complex and sometimes insoluble expressions.

Consequently, the division of the casting and the risers into an assemblage of simple shapes A, B, C, D, etc., such that

$$t_{s_R} > t_{s_A} > t_{s_B} > t_{s_C} > t_{s_D}, \text{ etc.}$$

and the subsequent application of the foregoing equations, must be accompanied by careful thermal analysis such as has been carried out by Ruddle[22] and Pellini and Bishop[47-50] and their respective coworkers.

The present authors feel that much can be gained from such a conjoint investigation and the testing of the simple heat flow theory, without resorting entirely to the empirical.

REFERENCES

1. N. Chvorinov, *Proc. I.B.F., 32* (1938-9); also *Giesserei, 27,* (1940) p. 177, 201, 222.
2. R. Ruddle, *Solidification of Castings,* Inst. of Metals, London, (1950).
3. T. King, *Iron and Steel, 26,* 10, 11, 12, 1953, and *27,* 1 (1954).
4. D. Atterton and T. Hoar, *J.I.S.I., 174* Sept. (1950).
5. J. Gittus, *B.C.I.R.A. Res. and Dev. Jnl., 5,* August (1954).
6. H. Gonya and D. Ekey, AFS TRANSACTIONS, *59,* p. 253 (1951).
7. A. Krynitsky and C. Saeger, *Jnl. Res. Nat. B. Stand. 13,* p. 579 (1934).
8. D. Atterton, *Proc. I.B.F., 45* p. A.28 (1952).
9. J. Gittus, *B.C.I.R.A. Res. and Dev. Jnl., 5,* April (1954).
10. V. Paschkis, AFS TRANSACTIONS *56,* p. 366 (1948).
11. Riemann and Weber, "Die Partieller Differtial Gleichungen Vieweg," *Bruns.,* VII, p. 100, 1912.
12. G. Halbart, "Element d'une theorie mathematique de la fonderie," (M. Vaillant), Liege (1945).
13. R. Ruddle, *Jnl., Inst. of Metals, 77* p. 1 (1950).
14. R. Ruddle and A. Mincher, *Jnl., Inst. of Metals, 76* p. 43, (1949-50).
15. W. Pellini and H. Bishop, AFS TRANSACTIONS, *62,* p. 646, (1954).
16. N. Girshovich and J. Nekhendzi, *Liteinoe Proizvodstro, 3,* p. 14, (1956).
17. W. Pellini, R. Morey and H. Bishop, AFS TRANSACTIONS, *62,* p. 448, (1954).
18. N. Russell, *Jnl. Am. Ceram. Soc., 18,* p. 1, (1935).
19. A. Eucken, *V.D.I. Forschungsheft* (B), *3,* p. 352, (1932).
20. C. Briggs & R. Gezelius, AFA TRANSACTIONS, *43,* p. 274, (1935).

21. C. Schwarz, *Arch. Eisenhüttenwesen, 5*, p. 139, p. 177, (1931), and Z. Angew, *Math. u. Mech., 13*, p. 202, (1933).
22. R. Ruddle and R. Skinner, *Jnl. Inst. of Metals, 79*, p. 35, (1951).
23. C. Trenkle, *Fonderie, No. 126*, July 1956.
24. J. Abcouver, AFA TRANSACTIONS, *43*, p. 90, (1952).
25. A. Schack, *Der Industrielle Warmeübergang*, V. Stahleisen m.b.H., Düsseldorf (1953).
26. H. Grober, *Die Grundgesetze der Warmeleitung und des Warmeüberganges*, J. Springer, Berlin, 1921.
27. E. Williamson & L. Adams, *Phys. Rev. 14*, p. 99, (1919).
28. N. Chvorinov, *Hutnicke Listy, 6*, Nos. 11 and 12, (1951), *8*, Nos. 1, 2 & 5, (1953).
29. J. Jaeger, *Proc. Roy. Soc.* (Edin) *61* A, p. 229, (1942).
30. R. Ruddle & A. Mincher, *Jnl. Inst. of Metals, 78*, p. 229, (1950).
31. D. Atterton, *J.I.S.I., 174*, p. 201, (1953).
32. G. Ohira & I. Igrashi, *Nippon Kinzoku Gakkai Si, 2*, No. 1, p. 40, (1940).
33. C. Briggs, C. Locke, and R. Ashbrook, AFS TRANSACTIONS, *62*, p. 589, (1954).
34. a) Report of subcommittee, TS21/33. *I.B.F. - Proc. I.B.F., 48*, p. A 128, (1955).
    b) TS.46 — as above.
35. E. Schmidt, *Föppls Festschrift*, Julius Springer, Berlin, p. 179, (1924).
36. V. Paschkis, AFS TRANSACTIONS, *59*, p. 381, (1951), *60*, p. 163, (1952).
37. Y. Tanasawa, *Trans. Soc. Mech. Engrs. Japan, 1*, p. 217, (1935).
38. M. Chamberlin & A. Peters, AFS TRANSACTIONS, *51*, p. 155, (1943).
39. A. Bell, AFS TRANSACTIONS, *53*, p. 366 (1945).
40. C. Lucks, O. Linebrinks and K. Johnson, AFS TRANSACTIONS, *55*, p. 62 (1947).
41. J. Gittus, *B.C.I.R.A. Jnl. Res. and Dev.*, June (1954).
42. W. Pellini, AFS TRANSACTIONS, *59*, p. 429 (1951).
43. W. Pellini, AFS TRANSACTIONS, *59*, p. 435 (1951).
44. J. Gittus and I. Hughes, *B.C.I.R.A. Jnl. Res. and Dev.*, Jan. (1955).
45. J. Middleton and H. Protheroe, *J.I.S.I. 168*, p. 384 (1951).
46. D. Atterton and D. Houseman, *Nature, 171*, p. 980, May (1953).
47. W. Pellini & Bishop, AFS TRANSACTIONS, *58*, p. 188 (1950).
48. W. Pellini, Bishop, & Myskowski, AFS TRANSACTIONS, *59*, p. 171 (1951).
49. W. Pellini, Bishop & Myskowski, AFS TRANSACTIONS, *60*, p 389 (1952).
50. W. Pellini, Bishop & Myskowski, AFS TRANSACTIONS, *61*, p 657 (1953).

APPENDIX

To facilitate use of various data cited in the current paper the following conversion factors are tabulated.

Parameter	Units A	Units B	Relation of A to B
Temperature	C	F	A = (B-32)5/9
Time, t	sec	min	A = B(60)
Time, t	sec	hr	A = B(3600)
Length	cms	in.	A = B(254)
Length	cms	ft	A = B(30.48)
Weight	gms	lb	A = B(453.59)
Area, A	cm^2	$in.^2$	A = B(6.4156)
Area, A	cm^2	ft^2	A = B(929.0)
Volume, V	cm^3 (cc)	ft^3 (cu ft)	A = B(28,317)
Density, ρ	gm/cc	lb/cu ft	A = B(1.602 × 10^{-2})
Density, ρ	gm/cc	lb/cu in.	A = B(27.68)
Quantity of heat, q	cal	Btu	A = B(252)
Specific heat, c	cal/gm/C	Btu/lb/F	A = B
Latent heat, L	cal/gm	Btu/lb	A = B(55.57 × 10^{-2})
Thermal Conductivity, k	cal/sec/cm^2/cm/C*	Btu/hr/sq ft/ft/F†	A = B(41.34 × 10^{-4})
Heat diffusivity, b $\left(=\sqrt{(k\rho c)}\right)$	cgs units	ft/lb/hr units	A = B(8.13 × 10^{-3})
Temperature diffusivity $\left(\alpha=\frac{k}{\rho c}\right)$	cgs units	ft/lb/hr units	A = B(25.52 × 10^{-2})

*Abbreviated frequently to cgs units.
†Abbreviated frequently to ft lb hr units.

PART II
SIZE AND SHAPE EFFECT ON SOLIDIFICATION TIME

INTRODUCTION

In Part I it was pointed out that progress in the application of heat flow theory to the problem of solidification of castings in sand molds depends, among other things, on the experimental verification of a number of currently used heat flow equations. Part II will describe and analyze some of the results that have been obtained with some simple shapes, namely, plates, cylinders and spheres. This work was carried out with two eutectic alloys, silumin (Al-Si eutectic) and cast iron, using a number of molding materials.

The accuracy of the equations, already derived by the authors in the first part

$$t_s = \frac{R^2}{\left[\frac{\sigma b_2 v}{2\rho_1 \bar{q}_1} + \sqrt{\frac{v}{\rho_1 \bar{q}_1}\left(\frac{\sigma^2 b_2^2 v}{4\rho_1 \bar{q}_1} + \frac{k_2}{C_1 C_2}\right)}\right]^2} \qquad (18)$$

for cylinders and spheres

$$t_s = \left(\frac{\rho_1 \cdot \bar{q}_1}{\sigma b_2 \cdot v}\right)^2 \cdot R^2 \qquad (19a)$$

for plates, together with the thermal constants fed into them, has been experimentally examined.

In order to make the text that follows clearer, results and experiments with each alloy are discussed separately.

EXPERIMENTAL WORK

Silumin Alloy

In order to limit experiments to a reasonable number, the three major variables, casting size, shape and mold condition were considered. Solidification times of castings were measured for the following sets of conditions:

a) Plates cast in dry sand molds of a medium-grained, medium-rammed synthetic sand.
b) Cylinders cast in dry sand molds of medium-grained, medium-rammed synthetic sand.
c) Spheres cast in green sand molds of medium-grained, medium-rammed synthetic sand.

Casting Sizes. The sizes of castings were as follows:

Plates ¼ x 6 x 6 in.
½ x 6 x 6 in.
1 x 6 x 6 in.
2 x 12 x 12 in.
Cylinders 4 in. high, diameters—0.6, 0.75, 0.85, 1, 1.5, 2, 3, 4, 5, 6 in.
3½-in. high, diameters—0.6, 0.85, 1, 1.5 in.
Spheres Diameters 1, 1.25, 2, 3, 4, 4.4 in.

Procedures

Melting Procedure. The alloy was based on super-purity aluminum, to which was added the requisite quantity of a high-purity temper alloy. The obtained scrap was subsequently remelted for castings within the series (plates, cylinders and spheres), any losses being made good with virgin alloy when necessary. Melting was carried out in a gas-fired crucible furnace. Casting temperature was 700 C (1292 F), and the metal was not allowed to exceed the temperature of 800 C (1472 F) in the furnace.

Molding Methods and Sand. Two distinct methods of molding were used, the first consisted of the use of a molding machine, which squeezed layers of sand that had been previously weighed into a suitable box. This method and the sand density distribution that it produces within the mold have been described previously.[1] The second method, which was used in making molds for the spheres only, consisted of the careful hand-ramming of similar layers around metal or wooden patterns. The molds for the cylinders were made by a method similar to the first, except that the cavities were cut out from the rammed sand by means of cylindrical steel shells.

The sand used was a 3½ per cent bentonite-bonded silica sand of medium grain size (distribution 60-180 B.S.S. No., 60-170 U.S.S. No.) and was rammed to a density between 1.4 and 1.5 gm/cc. After milling in a plough-type batch mill for 5 to 10 min, molding was immediately started. A moisture content of 3 per cent was maintained throughout. Drying for 8 to 10 hr at 200 C (392 F) then followed in the case of the dry-sand molds. Castings were made in the green sand molds not later than 20 min after assembly.

Gating and Pouring. The plate castings up to 1 in. wide were gated in the following manner—From a small pouring basin 2 in. in diameter and 2 in. deep, a ½-in. diameter runner led directly onto the top surface of the plate-shaped mold cavity at one of its corners. For the sake of symmetry and, in addition, to indicate approximately when the mold was filled, a cylindrical vent of exactly similar shape as the proper runner was made in the diagonally opposite corner of the mold. For the larger plate castings, the diameters of the runner and the vent were increased to ¾-in. The cylinders and all but two of the smallest spheres were bottom-gated, with channels which were of diameter not greater than one sixth the diameter of the casting. The two smallest spheres were top-gated.

Temperature Recording. Thermocouples used in the experiments were, a) 28 gage chromel-alumel type (sheathed within the mold cavity with twin-bore refractory tube along all but the last 3 mm of their lengths), for all castings except the large diameter spheres and the 2 in. thick plate, b) 24 gage chromel-alumel type, sheathed in thin tubular steel and insulated in a refractory powder. With both types of thermocouples the hot junctions and the last 3 mm of wire were coated with an alumina wash, thinly over the bead itself and slightly thicker elsewhere, so that penetration of the sheathing was prevented. The temperature-recording apparatus was of the pen-potentiometric type with a temperature accuracy ±2.0 C (3.8 F) and a full-scale deflection response of 1 sec.

Cast Iron

The cast iron experiments differed in certain respects from the experiments carried out with silumin. Such differences will be briefly described, all other experimental details being identical to those of silumin.

Casting Sizes. Sizes of castings were as follows:

Plates ¼ x 6 x 6 in.
½ x 6 x 6 in.
1 x 6 x 6 in.
2½ x 12½ x 14½-in.
Spheres Diameters 2, 3, 4, 4⅜, and 10½-in.

Materials Cast. Castings were poured in three irons, cupola-melted in a gray iron foundry. The compositions were as follows:

1. High-phosphorus gray iron (common engineering grade) C-3.5; Mn-0.4; Si-2.5; P-1.7; and S-0.12 per cent.
2. Medium-phosphorus gray iron (hydraulic castings grade) C-3.35; Mn-0.70; Si-1.98; P-0.53; and S-0.12 per cent.
3. Medium phosphorus inoculated iron (high-duty castings grade) C-3.25; Mn-0.17; Si-1.49; P-0.5; S-0.1; and Cr-0.05 per cent.

Molding Methods and Sand. The molding sand was a medium-grained, naturally-bonded red Bunter deposit sand. For all the castings made, two-part molds were used. In the case of plates, the drags contained the mold cavities. Initial experiments showed no significant difference between solidification time of castings made in molds compacted by a sandslinger, and those made by an experienced molder which were rammed to as near constant hardness as possible. Subsequent experiments were carried out using the hand-rammed type of mold. The green sand density was of the order of 1.6 gm/cc and the moisture content maintained at 5 per cent.

Gating and Pouring. The plates were gated in the same way as for silumin. The spheres were all top-gated except for the largest, which was side-gated.

Temperature Recording. The two types of thermocouple discussed in the silumin experiments were again used. The pouring temperature in both silumin and cast iron was measured in the ladle by means of a Pt/13 per cent Rh Pt thermocouple, also potentiometrically or with a quick-reading millivoltmeter.

CALCULATIONS DATA USED

Density of the liquid metals was taken as 2.42 gm/cc at 700 C (1242 F) for a eutectic aluminum-silicon alloy,[2] and 6.8 gm/cc at 1300 C (2372 F) for gray cast iron (hypoeutectic iron).[3]

Specific heat of the liquid metal and the latent heat at solidification for aluminum-silicon alloys are not

TABLE 2-1 — TEMPERATURES OF SOLIDIFICATION, POURING, FILLING AND AT THE MOLD-METAL INTERFACE

Temp. of	Temp. C (F) for Type of Iron: High - P (C.E. 4.9%)	Med. - P (C.E. 4.2%)	Med. - P Inoculated (C.E. 3.9%)	Reference
Liquidus	1182 (2156)	1221 (2228)	1232 (2246)	8
Solidus	1138 (2084)	1138 (2084)	1138 (2084)	8
Liquidus	1380 (2516)	1140 (2084)	1165 (2129)	obtained from
Solidus	1130 (2066)	1130 (2066)	1130 (2066)	equ. dgm.[8]
Liquidus	1127 (2057)	1175 (2147)	—	9
Solidus	1110 (2030)	1130 (2066)	—	9
Liquidus	1120 (2048)	1148-1160 (2102-2240)	1186 (2165)	this study
Solidus	1110 (2030)	1120-1130 (2048-2066)	1125 (2057)	this study
Pouring	1300 (2372)	1300†† & 1360** (2354 & 2480)	1340 (2444)	this study
Ave. on filling of castings:				
plates	1290 (2354)	1290 (2354)	—	this study
spheres	—	1200‡, 1240‡‡* (2192, 2264)	1240 (2264)	this study
		1200‡ (2264)†		
Ave. at interface during solidification	1040 (1904)	1060 (1940)† (overall mean)	1055 (1931)	this study

*dry sand mold.
†green sand mold.
**spheres only.
††plates and spheres.
‡spheres poured at 1300 C (2354 F).
‡‡spheres poured at 1360 C (2480 F).

readily available in the literature. However, the latent heat may be calculated:

a) From the density of the solid at the melting point,[4] for 11 per cent Si it is equal to 109 cal/gm; and for 13 per cent Si, equal to 123 cal/gm.
b) From an entropy calculation for 12 per cent Si it is equal to 65 cal/gm.
c) An experimental (calorimetric) determination by the authors for 12 per cent Si gave the value of 95 cal/gm.

This last value was used in the light of disagreement between (a) and (b).

The latent heat of solidification of gray cast irons is similarly difficult to obtain accurately. However, a value of 47 cal/gm is given in the literature[5] for a typical gray iron.

The specific heat of liquid aluminum-silicon eutectic can be calculated by means of the Neumann-Kopp rule[2] on the assumption that this is valid for eutectics. The value obtained is 0.259 cal/gm/°C.

The specific heat of the liquid cast iron was taken as that given by Unimo[6] for a typical gray iron: 0.212 cal/gm/°C.

Temperatures Recorded

Temperatures of pouring, filling, solidification and at the mold-metal interface are briefly given below.

a) For aluminum-silicon, the pouring temperature was 700 C (1242 F), with an allowance of 10 C (50 F) made as a pouring loss. The temperature of solidification was taken as that of the liquidus.[2] The temperature at the interface was taken as that indicated by a modified form of the Riemann equation,[7] this value having been shown to be close to the measured values (Part I).
b) For cast irons, supplementary experiments reported in Part I with center and interface thermocouples gave the results in Table 2-1, which are compared with the existing values from the literature.

The disagreement between observed and calculated interface temperature is serious and presumably due to the unsuitability of the α_1 term for cast iron in the equation used.

Thermal Properties of the Molds

Aluminum-Silicon Eutectic. For dry sand molds the mean heat diffusivity b_2 was calculated from data of Atterton,[10] who gives the thermal conductivity k_2 at various temperatures in a similar sand, and data of Ohira[11] for mean specific heat.

$$b_2 = 0.0150 \text{ cgs units}$$
$$k_2 = 0.0006 \text{ cgs units}$$

For green sand molds, Ruddle and Mincher[12] have given the value for the mold constant $\sigma v b_2$ under conditions similar to those of the present experiments. The mean heat diffusivity b_2 was calculated from this constant. In turn, the mean thermal conductivity k_2 was calculated from b_2 using the expression of Ohira[11] for the specific heat of a green sand.

$$b_2 = 0.022 \text{ cgs units}$$
$$k_2 = 0.0011 \text{ cgs units}$$

Cast Irons. The mean diffusivity of the sand used (in its dry state) was calculated from data obtained by the authors:[1]

$$b_2 = 0.019 \text{ cgs units}$$

The mean thermal conductivity was again calculated from the relation

$$b_2 = \sqrt{k_2 \rho_2 c_2}$$

where c_2 was taken from Ohira's data

$$\rho_2 \ 1.6 \text{ gm/cc and}$$
$$k_2 = 0.008 \text{ cgs units.}$$

The volume to surface area R, for the cases of spheres, plates and cylinders whose major dimensions were at least six times their minor dimensions, was calculated ignoring the end effects. However, for all other castings (i.e. those deemed to exhibit the corner effect) the ratio of volume to effective surface area R' was calculated from the method and data of Ruddle and Skinner.[13]

DISCUSSION OF RESULTS

It has been shown in the preliminary discussion of the heat flow equations (Part I) that castings in the three basic shapes used, with volume-to-surface ratio and all other factors being equal, solidify at different times. However, the difference between the solidification time of a wide plate and that of a sphere is greater than the difference between those of a sphere and a long cylinder, all other factors being equal. This point is best illustrated diagrammatically by means of a log plot of volume-to-surface ratio R against solidification time t (Fig. 2-1).

The effect of other factors such as mold medium, casting alloy, temperature of mold or temperature of solidification of castings are also best illustrated in the same manner (Figs. 2-2 to 5). The experimental results, together with the calculated relationships for the three sets of experiments, are shown in Figs. 2-6 to 2-14.

Aluminum-Silicon Eutectic

Probably two of the important assumptions made in the application of simple heat flow theory to the present problem were a) the assumption of a mean interface temperature for a range of castings in dry sand and green sand molds and b) the assumption, that the drying out phenomenon,[1] which reduces the apparent thermal conductivity k_s of the mold with time, is not so effective at a lower interface temperature as it is with that of, for example, aluminum-bronze,[1] i.e., that mean value for the lower temperature range can be inserted in equations (18) and (19).

The first assumption stands or falls on the experi-

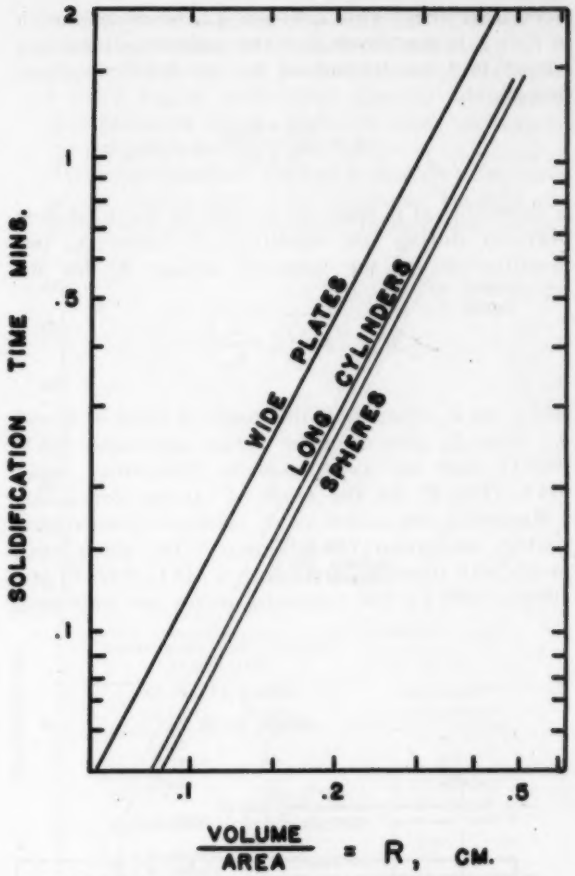


Fig. 2-1 — Casting shape effect on solidification time.

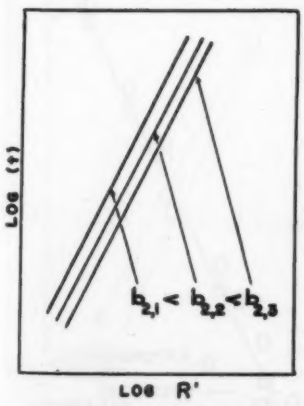


Fig. 2-2 — Mold material effect on solidification time ($\bar{q}_1$, θ_1 and θ_0 assumed unchanged).

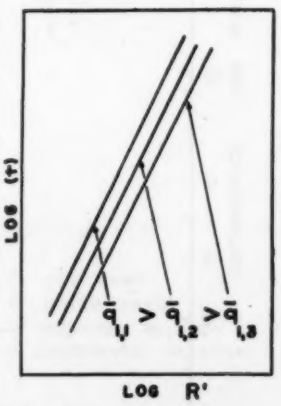


Fig. 2-3 — Total heat content effect per gram of molten metal on solidification time (b_2, θ_1 and θ_0 assumed unchanged).

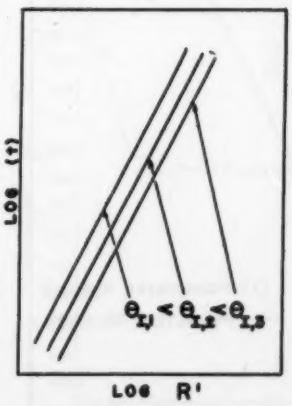


Fig. 2-4 — Interface temperature effect on solidification time (b_2, $\bar{q}_1$ and θ_0 assumed unchanged).

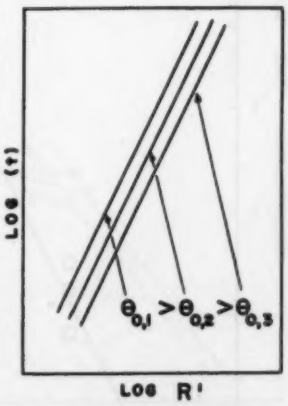


Fig. 2-5 — Mold temperature effect on solidification time (b_2, θ_1 and $\bar{q}_1$ assumed unchanged).

mental evidence. This question has been dealt with in Part I. It was shown that the mold-metal interface temperature, as determined by the Riemann equation (2),

$$\theta_i = \frac{\theta_1 b_1 + \theta_2 b_2}{b_1 + b_2}$$

is unreliable as it takes no account of the total heat liberated during the solidification. However, two modifications of the equation—writing b_1^1 for b_1, where

$$b_1' = \sqrt{K_1 \rho_1 \left(c_1 + \frac{L_1}{\theta_m}\right)}$$

and θ_1^1 for θ_1, where θ_1^1 is the weighted mean of θ_c and θ_m (Part I), give a rather better agreement 563 C (995 F) with the experimentally determined mean 574 C (1067 F) for the range of castings concerned.

Regarding the second major assumption, when comparing temperature distribution[12] for green sand molds with interface temperatures 548 C (1022 F) and 1083 C (1985 F), the following points are indicated:

Time After Casting, min	Depth of Mold (mm) above 100 C (212 F) 548 C* (1022 F)	1083 C* (1985 F)
3	13	25
10	29	37
20	44	48

*Nominal interface temperature

These data confirm that initially the speed of drying out is slower in a mold where a lower interface temperature is maintained. This permits the use of the approximate mean heat diffusivity value for the green sand in calculations of solidification times of castings with volume-to-surface-area ratios below about 25 mm. The agreement is seen clearly in Fig. 2-11.

Unfortunately, an assessment of accuracy for the spheres cast in green sand is impaired by the assumption that a 10 C (18 F) temperature drop occurred in both of the gating systems. It appears that with more quiescent methods of pouring this is approximately true (i.e., with plates and cylinders). However, the more turbulent gating methods, which were used for the spheres, no doubt result in higher temperature losses. Further experimental work will be needed to consider the possible variation of heat transfer in the liquid metal with varying degrees of turbulence reported by Chvorinov[14] in a recent literature review.

R to t Relationship

The calculated relationship of R to t for the castings in dry sand molds are slightly below the experimental points in each case. This is possibly due to inaccuracy of the specific heat value calculated by the Neumann-Kopp rule. The agreement of slopes indicates justification of assuming a mean interface temperature for the particular range of castings used. The scatter of experimental points, which seems to be more pronounced in this alloy than in others which were investigated, may also be partly due to compositional changes through scrap melting.

When the three series of experiments are examined, with due consideration to the assumptions made in

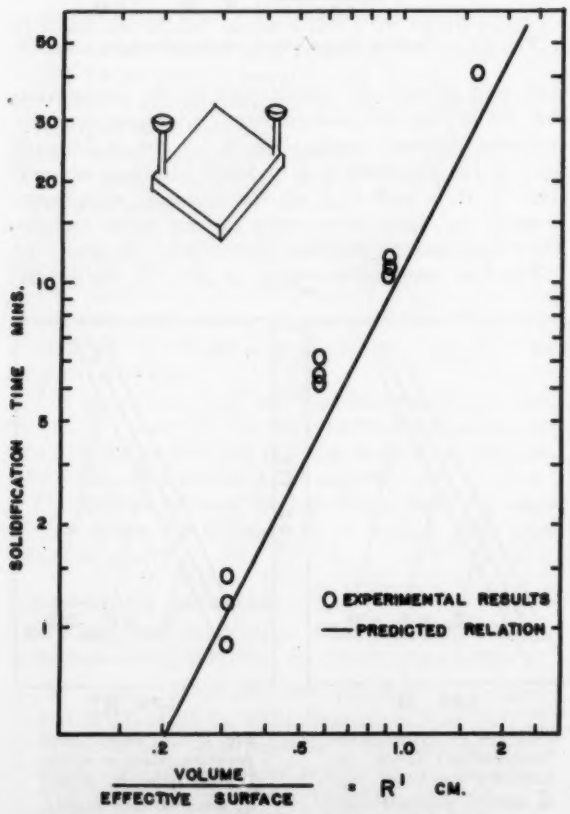


Fig. 2-6 — Silumin plates in dry sand molds.

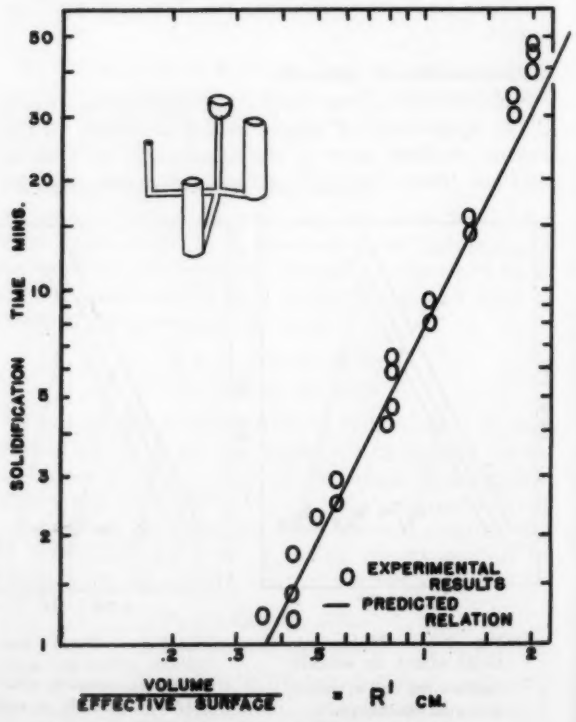


Fig. 2-7 — Silumin cylinders in dry sand molds.

applying simplified heat flow theory, the following comparison results:

Shape	Sand Mold	Discrepancy between calculated and experimental result, % Ave.	Max.	Min.
Plates	Dry	23	40	0
Cylinders	Dry	13	40	0
Spheres	Green	18	50	0

Cast Irons

1) Plates of a medium phosphorus ("hydraulic") iron (C.E. 4.2 per cent) cast in green and dry sand molds. The results of these experiments are found in Fig. 2-8. The solid line represents the relation predicted by equation (19) for dry sand mold castings. The broken line connects the experimental points for green sand mold castings. The agreement between theory and experiment is reasonably good in this case. It is also interesting to note the disappearance of the increase in chilling power brought about by the presence of moisture in the sand as the casting section increases.
2) Plates of a high phosphorus iron cast (C.E. 4.9 per cent) in green sand molds. In a recent paper Gittus and Hughes[9] have shown, by comparing the results of thermal analysis with those obtained by quenching, that in certain irons (particularly the higher phosporus type) eutectic solidification is still proceeding below the knee of the time-temperature curve. In fact, it may probably continue to do so until the commencement of the phosphide eutectic solidification. However, the work would seem to justify the assumption that the knee represents the end point of solidification for all but a small percentage of the mass of the casting.

Figure 2-9 shows the predicted relation of R and t for a higher phosphorus iron, in addition to that predicted for the hydraulic iron, and experimental points for both materials.

The points plotted, marked p represent the knees of the curves (Fig. 2-10). Disparity between the theoretical and experimental relationship has un-

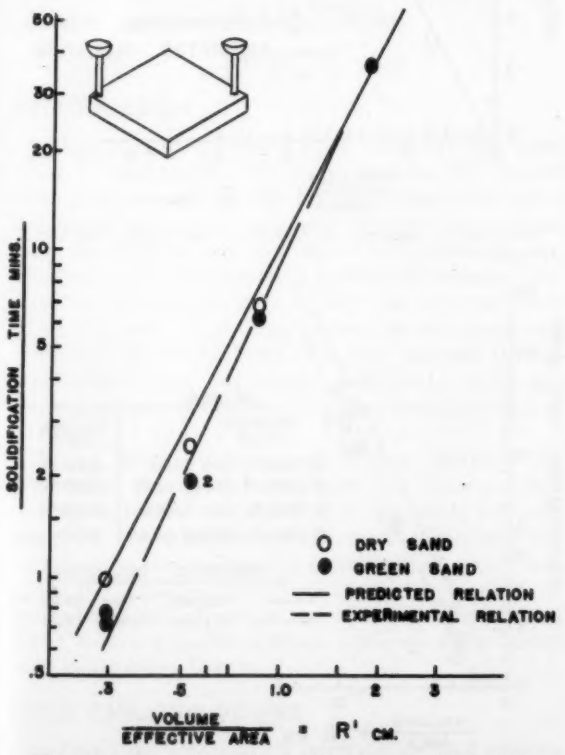


Fig. 2-8 — Gray iron plates, 4.2 per cent C.E. (hydraulic iron).

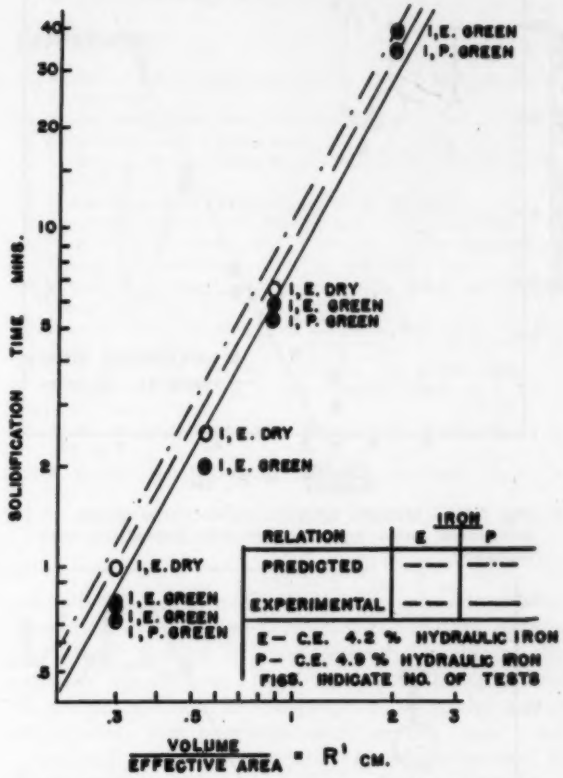


Fig. 2-9 — Gray iron plates in green and dry sand molds, irons E and P.

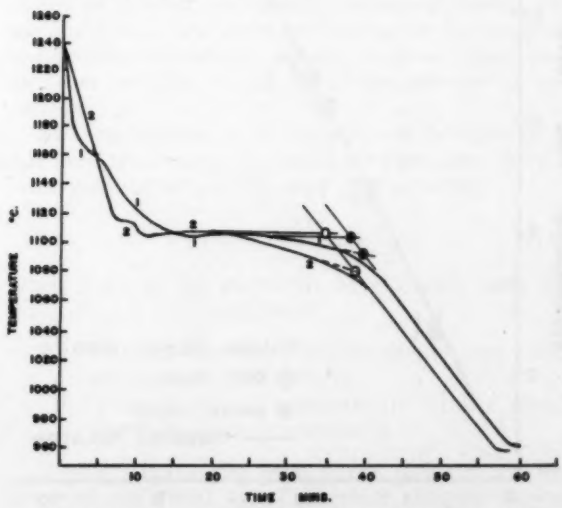


Fig. 2-10 — Time temperature relationships for 1) 4.2 per cent C.E. (hydraulic iron) and 2) 4.9 per cent C.E. (high phosphorus iron). Tangents drawn to give range of values of t_s.

doubtedly been caused by the difficulty in fixing an end-point of solidification for the iron of high phosphorus content.

3) Spheres of medium phosphorus (hydraulic) iron cast (C.E. 4.2 per cent) in green and dry sand molds. In Fig. 2-12 the theoretical relation for a given filling temperature for the hydraulic iron cast in dry sand molds is plotted as a solid line. Experimental points show a good agreement. It is interesting to note that, as before, only for the castings of small section (example, 2 in. diameter) has moisture had any prolonged effect on mold chilling power.

4) Spheres of medium phosphorus, inoculated (high-duty) iron (C.E. 3.9 per cent) cast in green sand molds. Figure 2-13 shows the theoretical relationship for the above iron, and agreement between theory and experiment is quite satisfactory.

5) Spheres of medium phosphorus hydraulic iron (C.E. 4.2 per cent) cast in dry and green sand molds at two pouring temperatures. The theoretical relationships drawn in Fig. 2-14 are 1) broken line—hydraulic iron, max. temp. in mold 1240 C (2264 F) and 2) solid line—hydraulic iron, max. temp. in mold 1200 C (2192 F). These correspond

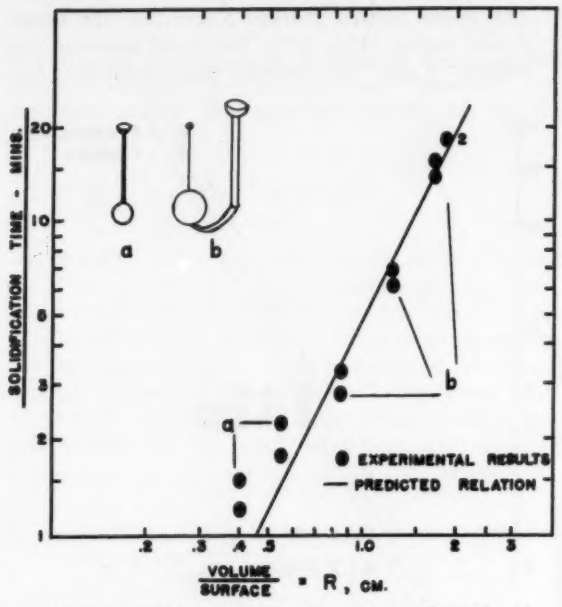


Fig. 2-11 — Silumin spheres in green sand molds, a) top-poured, plain ingate, b) bottom poured, reversed horngate.

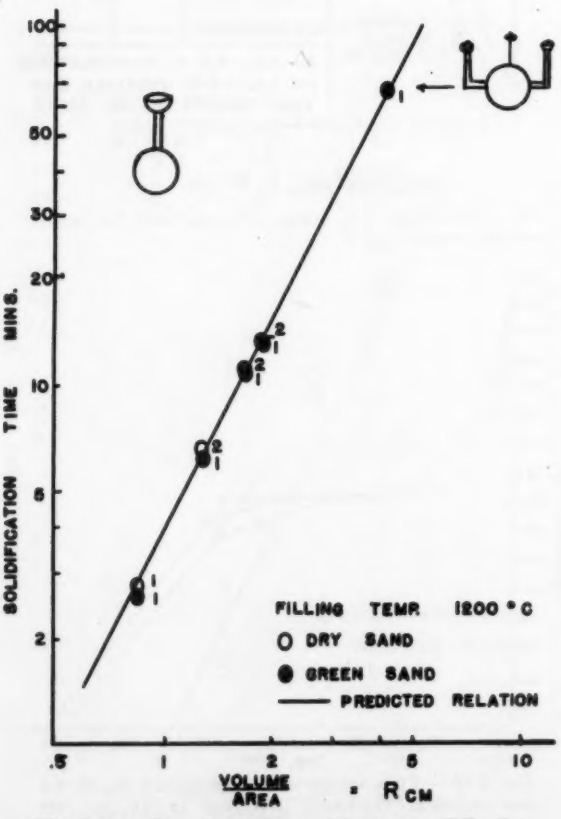


Fig. 2-12 — Gray iron spheres, 4.2 per cent C.E. (hydraulic iron).

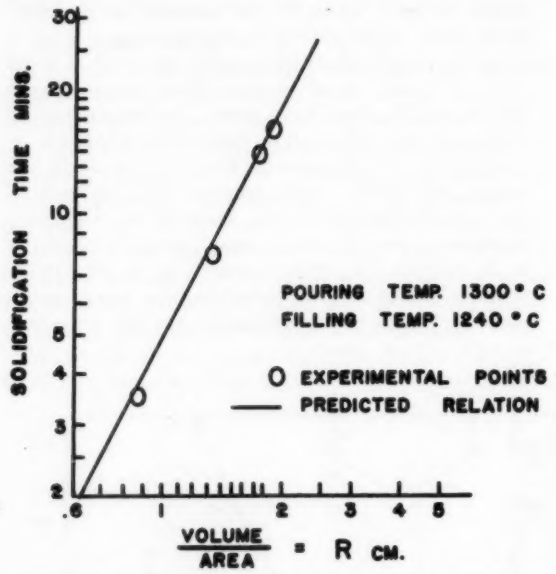


Fig. 2-13 — Spheres of high-duty iron (C.E. 4.9 per cent).

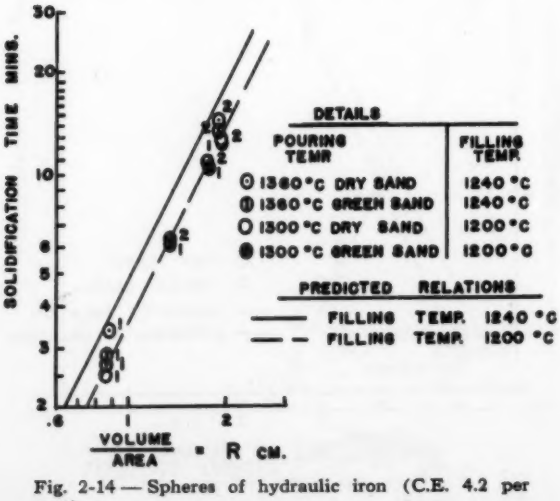


Fig. 2-14 — Spheres of hydraulic iron (C.E. 4.2 per cent).

to the two pouring temperatures 1360 C (2480 F) and 1300 C (2372 F), respectively, for dry sand molds. The experimental points shown include those mentioned subsection 3 (above) of this discussion. The results obtained for castings of two sphere sizes poured in green and dry sand molds at the higher temperature would seem to indicate the possible importance of moisture during the mold swilling action immediately prior to the onset of solidification.

On the basis of these latter results, further attention is justified to the phenomena of turbulence in deep castings to be carried out conjointly, if possible with study of the effect of gating systems on the temperature loss.

CONCLUSIONS

As a guide to calculation of progress of solidification times as applied to soundness problems, the following conclusions can be drawn for the range of mold shapes and metal temperatures examined.

1) Mold cavity shape, heat content of the casting, mold metal interface temperature and temperature and thermal properties of the mold have been shown to shift vertically the $\log \cdot t$-$\log \cdot R$ relationship.
2) The solidification time of three simple shapes and the three alloys used can be calculated with reasonable accuracy, provided certain assumptions are made regarding temperatures at the mold-metal interface and of the casting immediately after pouring. Castings such as plates (side or edge gated) exhibit minimum turbulence losses of 10 C (50 F) whereas naturally deep castings which are top-gated (e.g., the cast iron spheres) exhibit a maximum loss (example, 100 C (180 F) in cast iron spheres). Quiescently poured shapes (example, bottom-poured cylinders) would appear to have a minimum loss of a similar order to that in plate castings, 10 C (18 F).

REFERENCES

1. Report of Sub-committees *Proc. Inst. Brit. Found., 48,* TS 21, 33, 46, 1955.
2. A.S.M. *Metals Handbook,* Cleveland, 1948.
3. C. Schwarz, AFS TRANSACTIONS, *53,* p. 1, 1945.
4. R. Ruddle and A. Mincher, *Jnl. Inst. of Metals, 78,* p. 229, 1950-51.
5. J. Abcouver. AFS TRANSACTIONS, *60,* p. 90, 1952.
6. A. Unimo, Toh. Imp. Unvi. Sci. Rep., *15,* p. 331, 1926.
7. Riemann and Weber, *Die Partieller Differtialgleichungen Vieweg,* Braunschweig, *vol. 11,* p. 100, 1912.
8. E. Greiner, J. Marsh, and B. Stoughton, *Alloys of Iron and Silicon,* McGraw-Hill, N. Y., 1933.
9. J. Gittus and I. Hughes, *Jnl., Res. Dev. Brit. Cast Iron Res. Assn.,* Jan. 1955.
10. D. Atterton, *Jnl. Iron and Steel Inst., 174,* p. 201, 1953.
11. G. Ohira and I. Igrashi, *Nipon Kinzoku Gakkai Si, 3,* No. 1, p. 40, 1944.
12. R. Ruddle and A. Mincher, *Jnl. Inst. of Metals, 76,* p. 43, 1949-50.
13. R. Ruddle and R. Skinner, *Jnl. Inst. of Metals, 79,* p. 35, 1951.
14. N. Chvorinov, *Hutnicke Listy, 6,* Nos. 11 and 12, 1951, *8,* Nos. 1, 2, and 5, 1953.

PART III
MOLD MATERIAL EFFECTS

INTRODUCTION

It has been emphasized (Parts I and II) that there exists a considerable dearth of data on the various thermal constants involved for simple heat-flow calculations in castings. It has been shown[1] that a wide range of solidification times exists for a standard side-run 6 x 6 x 1 in. plate casting molded in various types of sands. The object here is, first, to interpret these results in terms of chilling power, or the mean heat diffusivity b_s of the mold. For this purpose further experiments, not previously published, were performed, involving temperature measurements in sand molds.

Second, from the values of b_s thus obtained, the prediction of solidification times for various plate castings in the three types of mold materials has been made. In addition, further information has been obtained on a) the movement of moisture in green sand molds and its subsequent effect on chilling power, and b) the feasibility of computing dry sand mold chilling power from a knowledge of the grain size distribution and certain other factors.

MOLD CHILLING POWER

In Part I the information referring to mold chilling power has been summarized. The majority of the methods practiced involved the measurement of sand temperatures.[2-4] It was decided that the data already published[1] should be re-examined to obtain the values of the chilling power, using solidification time of actual castings, in conjunction with the metal temperature measurements taken at the same time.

Chilling power may be conveniently expressed as the mean heat diffusivity of a mold b_s, where b_s is given by $\sqrt{k_2 \rho_2 c_2}$ where k_2, ρ_2 and c_2 are the mean values of thermal conductivity, ramming density and specific heat of the molding material in the range of temperature considered, namely, between room temperature and the liquid metal temperature of the casting.

By consideration of the simple heat flow theory it has been shown that the heat conducted away from a wide plate of area $2A$ in time t_s is given by:

$$q = (2A) b_2 (1.13) (\theta_i - \theta_0) \sqrt{t_s} \qquad (1b)$$

where some of the characters have already been defined (Part I) and, in addition,

θ_i indicates interface temperature (at the wide plate-mold face interface).
θ_0 indicates initial temperature of the mold (generally room temperature).

However, it has also been shown that the temperature of the mold metal interface changes during solidification. Consequently, this quantity must be either measured or calculated by some empirical or theoretical method. The unreliability of theoretical

methods (Part I) prompted the authors to carry out careful experimental measurements of this quantity. Two extreme molding sands were considered, one a coarse-grained and the second a fine-grained sand.

Details of the experimental measurements, and a graph showing the time-temperature relationship were discussed (Part I). From this work and the results of interface temperature measurements for several other sand conditions, a mean value for interface temperature was obtained:

$$\theta_i = 1020\ \text{C}\ (1868\ \text{F})$$

for the material currently being considered (aluminum bronze). If we equate the amount of heat absorbed by the mold as given by equation (1b) with amount of heat liberated on solidification we have:

$$(\text{mass})\,[(\theta_i - \theta_m)\,c_1 + L_1] = (2A)\,b_2\,(1.13)(\theta_i - \theta_0)\sqrt{t_s} \qquad (1c)$$

where

θ_1 = temperature of metal within the casting immediately after filling mold.
θ_m = temperature of solidification.
c_1 = specific heat of the liquid metal.
L_1 = latent heat of solidification.

Thus, when the necessary temperatures and physical constants are known, the mean heat diffusivity may be calculated from the solidification time t_s. Experimental measurement of the temperature of the metal within the mold cavity on pouring showed this figure to be within 8 to 10 C (46-50 F) of the pouring temperature for this particular casting.

Corner Effects

In equating the amount of heat absorbed by the mold with that liberated by the solidification of the casting, it has been assumed that no heat is transmitted from the casting through either the side walls of the mold cavity or through the gating system.

Ruddle and Skinner[5] have shown that the effect of side walls or, more specifically, corner effects can be dealt with in the following way.

By measurement of the mold temperature distribution around various types of corner for several casting media, these workers were able to derive numerical factors connecting the actual geometrical area of a corner portion and their effective plane area. Thus, by this method an effective plane area for the whole mold face A' has been calculated, and is used instead of $2A$ in equation (1c).

Finally the heat lost through the gating system was estimated by means of a thermal analysis method which is described in the Appendix to this section. The numerical value of the heat so lost was found to be 30.5 cal/sec of solidification time.

The chilling power b_s of the various dry molding sands can also be calculated from the properties of the sand. This method is due mainly to Russell[6] and Eucken,[7] who have shown that the apparent thermal conductivity of a porous refractory material k_a, such as sand, can be calculated from the conductivity of its pores k_p (i.e., air) and the porosity fraction P.

The relationship derived can be conveniently approximated to:

$$k_a = k_p\left(\frac{3}{p} - 2\right) \qquad (20)$$

However, a correction must be made at higher temperatures for radiation through these pores. Russell, correcting for this, gives the equation:

$$k_p = k_t + (5.44 \times 10^{-12})\,T^3\,X \qquad (21)$$

where

k_t is the thermal conductivity of air, in cgs units[6-8]
T is the absolute temperature in °K.
X is the pore space diameter in cm.

Bell[9] has given an approximation for pore space diameter for a two-sieve distribution sand.

More recently Gittus[10] has extended this method to sands of any distribution. This method utilizes the area A_c under the cumulative distribution curves[1] in the following way:

$$X = \frac{200\,[\rho_{SiO_2} - \rho_B]}{3A_c\,\rho_B}$$

where

A_c is in gm.cm.
ρ_{SiO_2} is the true (solid) density of silica in gm/cc.
ρ_B is the ramming or bulk density of the aggregate.

This method of obtaining the average pore space diameter X has been preferred in the present investigation.

Figure 3-1 shows the results obtained from equations (20-22), together with those obtained by the Institute of British Foundrymen, as b_s, the mean heat diffusivity, plotted against ρ_2, the ramming density. The data derived for b_s from the results of Atterton[9] are also included. It is to be noted that the results derived from the above,[9] are not the simple arithmetic means of the room temperature and casting temperature values but integrate means, as the variables concerned (ref. 6 and equation 21) are directly related to the cube of the absolute temperature.

It should also be noted that the figure for b_s of green sands shown on the graph apply strictly to molds with castings of 1-in. thickness, and that the mean heat diffusivity b_s plotted applies to castings solidifying in the temperature range 1000-1200 C (1832-2192 F).

MOLD CHILLING POWER AND CASTING SOLIDIFICATION TIME

It has been shown that the solidification time of a wide plate is given by:

$$t = \left(\frac{\rho_1\,\bar{q}_1}{\sigma\,b_2 v}\right)^2 R'^2 \qquad (19b)$$

where

R' is the ratio of volume to effective surface area.
ρ_1 is the density of the liquid metal.
σ is the constant 1.128.

$$v = \theta_1 - \theta_0.$$
$$\bar{q} = [(\theta_c - \theta_m)c_1 + L_1]$$

It might be expected for green sands and green or dry sands containing combustible material that b_2 changes with increasing casting thickness, due to the drying and burning-out phenomena described.[1] Similarly, the value of θ_1 (the average temperature of the mold-metal interface) also changes with casting thickness. This latter possibility has been examined in certain preliminary experiments (Part I), and thus a mean value of θ_1 for a large range of casting thickness has been evolved — $\theta_1 = 1020$ C (1868 F).

Since there is no readily available method of predicting chilling power of green sand mold or molds containing combustible materials, it was decided to compare the results for these materials with those for similar dry sand molds.

A convenient way of showing the relation of t (solidification time) to R' (or the ratio of volume to effective surface area) as expressed in equation (19) is to plot log t against log R'. The relationship thus obtained is a straight line of slope 2, intersecting the line R′ = 1.0 cms at

$$t = \left(\frac{\rho_1 \bar{q}_1}{\sigma b_2 v}\right)^2 \text{sec}$$

Thus, providing that the physical constants and other quantities in the above relation are known, the

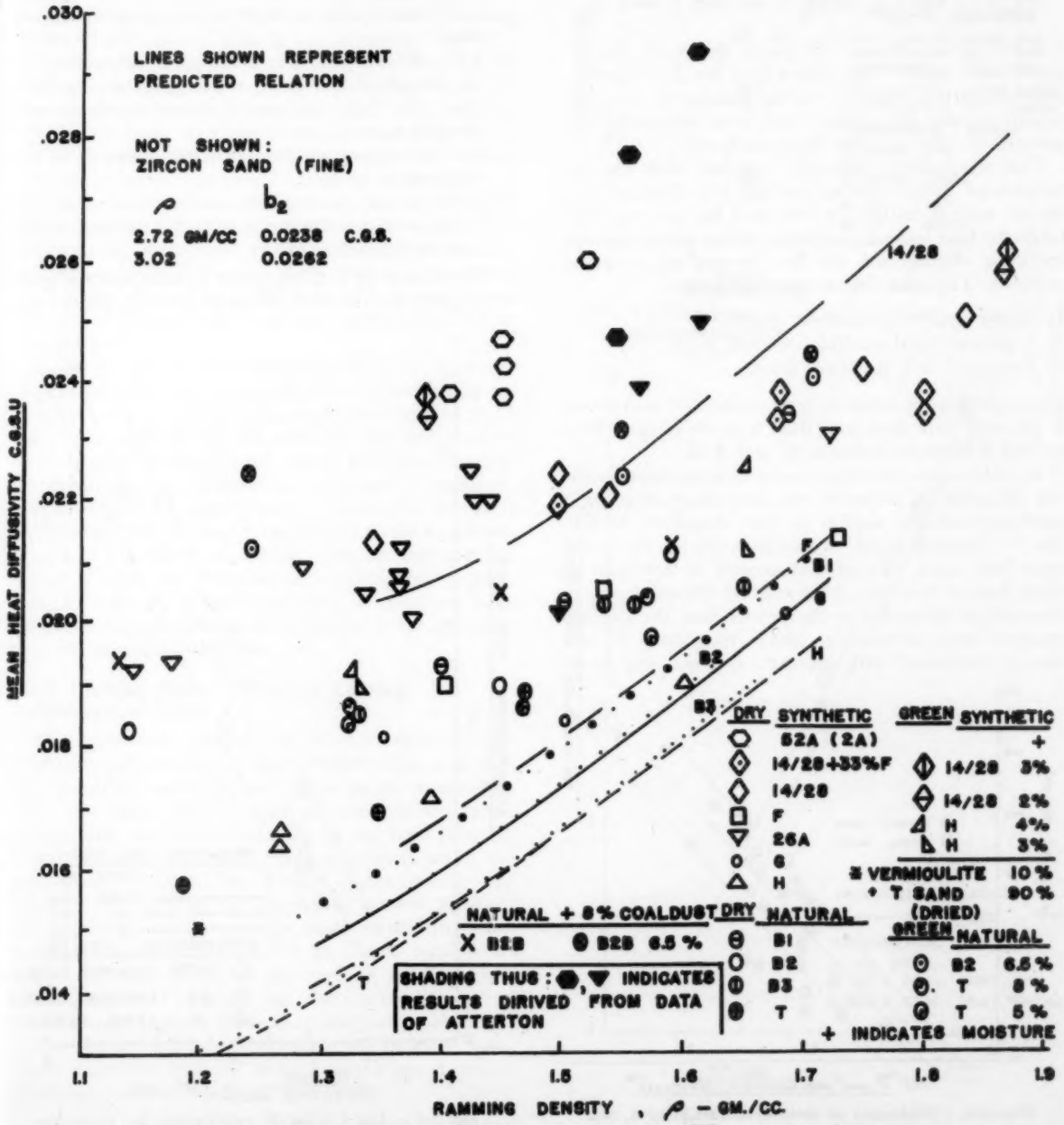


Fig. 3-1 — Relationship of heat diffusivity b_2 to ramming density ρ_2 for various sands.

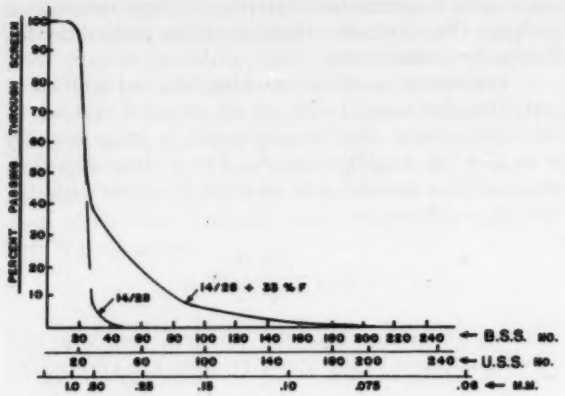


Fig. 3-2 — Effect of adding 33 per cent F sand to 14/28 sand.

line connecting R' and t is readily drawn on double logarithmic paper. The values used for L_1, c_1 and ρ_1 were calculated, (Part I). In the absence of accurate, experimentally determined data, these calculated values must be used as a first approximation.

The relationship obtained, together with the experimental data, is shown in Fig. 3-4. Only for the thicker section castings in fine sand has account been taken for heat losses through the gating system, as only for these castings did the loss become of great importance. The sand conditions studied were:

1) Coarse sand, hard rammed, dried.
2) A medium sand, medium rammed, dried.
3) Fine sand, soft rammed, dried.

Sands containing moisture (green) and one containing 8 per cent superfine coal dust were investigated for casting thicknesses between ¼ and 3 in.

In addition to the temperature measurements made, the diffusion of moisture was also observed using a method somewhat similar to that described by Gittus.[11] Thermocouples accommodated in the mold were here used as potential probes, in addition to their normal function. This method allows the measurement of resistance at the points near the thermocouples semi-continously. Since resistance is intimately connected with moisture content, the movement of a moisture wave with time can be shown graphically. The results of all the above experiments are shown in Figs. 3-4, 5, 6, 7, 8 and 9.

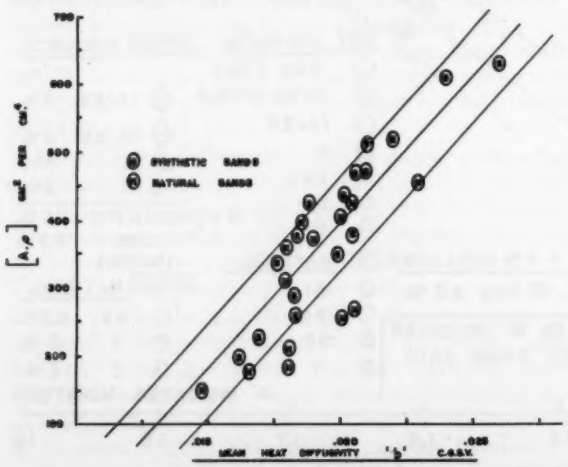


Fig. 3-3 — Correlation of mean heat diffusivity b_2 and the product of density ρ_2 and the Gittus factor A_c.

DISCUSSION

Chilling Power of the Mold

The relationship of chilling power b_s to ramming density ρ_s for all the sands examined is typified by a curve of positive slope. Both the slope and the vertical displacement of such a relation appear to be changed, respectively, with average grain size and grain size distribution of a sand.

The following generalizations can be made:

1) Chilling power b_s increases with increase in ramming density ρ_s.
2) b_s increases with increase in average grain size of sand.
3) Rate of change of b_s with ρ_s is dependent on the length of the "tail" of the grain size distribution (Fig. 3-2), i.e., the gradient decreases with the addition of fines.
4) For the casting thicknesses used, green sand has high values of b_s.
5) Addition of combustible material increases the value of b_s for the sands with the particular casting thicknesses used.

The change in chilling power b_s with average grain size, grain size distribution and ramming density has

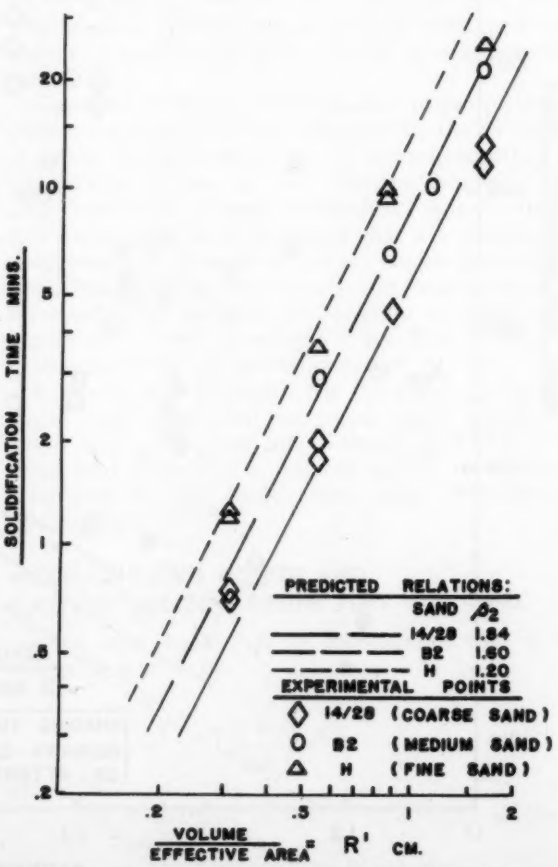


Fig. 3-4 — Log t — log R' relationship for aluminum-bronze plates.

been explained along the lines put forward for the corresponding change in solidification time t considered in ref. 1.

Figure 3-1 also includes the points derived from the results of Atterton.[4] It is to be noted that:

1) The results of Atterton expressed as b_s versus ρ_2 show a rather steep gradient, much steeper than any of the results derived from measurements of solidification times.
2) The values of 26A and 2A sands (which should be directly comparable to those of Atterton for 26A and 52A sands) are of a lower order than the values of Atterton.

It is possible that the single squeeze ramming methods which were used,[4] and the steep density gradients associated with this type of ramming have, in fact, displaced the results. This is not altogether unlikely, and is further inferred when one considers the results predicted, using equations (20-22) which, although they are only approximations,[11] do indicate a less steep slope than the figures derived from ref. 4.

The unsuitability of the wholly theoretical method for smaller grain size sands has been suggested[11] to be due to changes in condition of contact. Further, it was suggested that the formation of domains (i.e., the grouping of the small grains within the substantial coating of clay bond) takes place. This may probably explain the change in appearance of fine sand when first the bentonite bond and then water are added to it in the mill.

It was thought that considerable use could be made in practice of curves connecting mean heat diffusivity b_s with a factor combining ramming density and grain size. Such a factor can be obtained by the product of the Gittus number A_c of a sand and the ramming density ρ_2. Figure 3-3 shows the suggested correlation. The obvious limitation of such a correlation has been outlined in the previous paragraphs, i.e., the suggested change of conditions of contact with decreasing grain size of a sand. Nevertheless, it is interesting to note that 25 of the 29 points lie between or upon two lines drawn at an approximate deviation of 0.002 cgs units from the most favorable line.

Mold Chilling Power Effect on Casting Solidification Time

The agreement between the experimental and the predicted relationship of solidification time and volume-to-surface ratio appears to be quite reasonable, when one takes into account the assumptions made in applying the basic equations of the heat flow. It is interesting to note that when account is taken for heat losses through the gating system (using the method described in the Appendix to this section), the agreement for the castings made in the fine sand molds is improved (Fig. 3-10). It is also possible to correct for riser losses in the cases of medium and coarse-grain sand castings, but the corrected relation deviates only slightly from that originally predicted.

The results of the experiments made in green sand molds seem to indicate several important features which may be briefly summarized:

1) Moisture effect on the chilling power is clearly

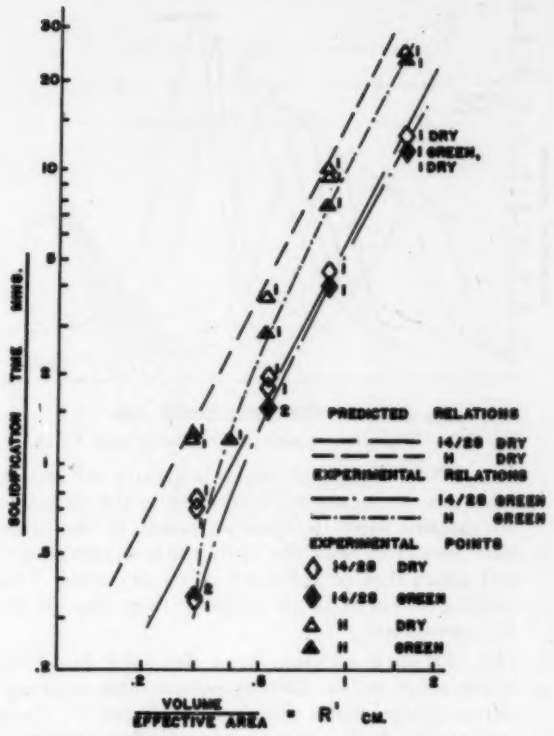


Fig. 3-5 — Moisture effect on solidification time for aluminum-bronze plates.

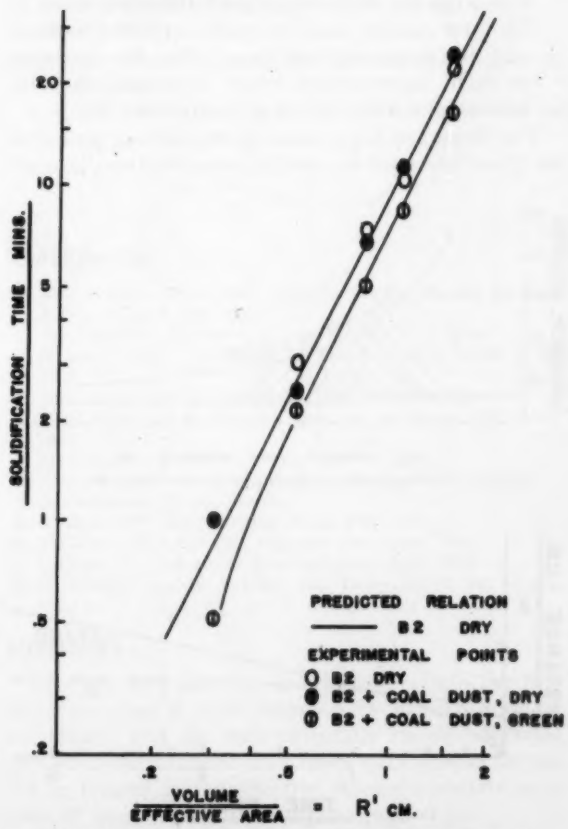


Fig. 3-6 — Coal dust effect on the chilling power of B2 sand for various plate sections in aluminum-bronze.

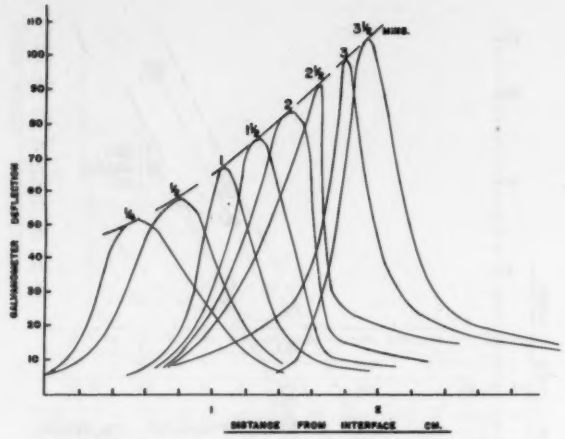


Fig. 3-7 — Migration of moisture in coarse sand 14/28.

shown, but the effect depends greatly on section thickness. Particularly interesting is the fact that for the fine sand the chilling power in the green state associated with the ¼-in. thick section, is several times that of the sand in its dry state. This chilling power is in fact greater than that of the dry coarse sand.

2) The coal dust addition to a dry sand has little or no effect on its chilling power. This is an apparent disagreement with previous work.[12] There appears to be a decrease in chilling power at larger casting thickness which is quite probably real, since the difference in solidification times of the three castings made in molds with and without coal dust is considerably larger than any deviation through experimental error (although the differences are reduced on a logarithmic plot).

The unusually high value of the chilling power of the green fine sand for small casting thickness as compared with the dry fine sand may be explained by means of the results of the moisture migration experiments (Fig. 3-7, 8 and 9). The movement of the peak moisture contents which may be considered as a prime mode of heat transfer in the time during and immediately after pouring, are shown in Fig. 3-9, being derived from the two previous figures. The figure would indicate a most sudden evaporation and outward movement of moisture in the fine sand during a period corresponding to the times of solidification of the ¼- and ⅜-in. thick castings, those casting thicknesses where the chilling power exhibited by the mold was greatest.

The fact that the curve then flattens out would also seem to correspond with the almost stationary value of chilling power for the greater casting thicknesses.

In the instance of the green coarse sand, the moisture movement is more gradual, corresponding to the gradual change of chilling power with casting thickness. Any explanation of why this change in chilling power differs for the two sands would undoubtedly involve the loss of moisture near the mold face on standing during the interval between molding and pouring (usually about 20-30 min). The drying-out of the fine sand during this period would probably be much less than that experienced in the coarse sand which has a much higher permeability.

The great increase in the height of the peaks

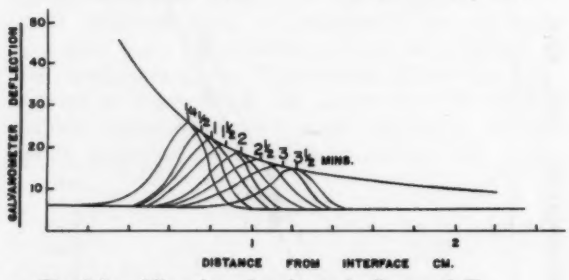


Fig. 3-8 — Migration of moisture in fine sand *H*.

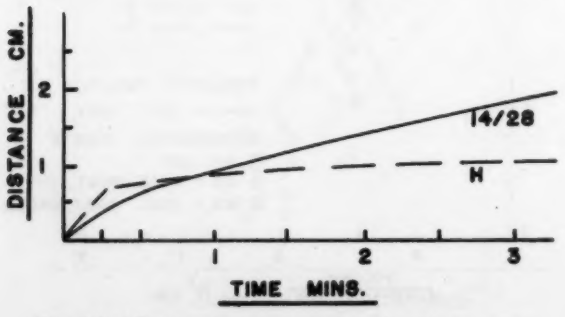


Fig. 3-9 — Rate of movement of peak moisture contents in the two sands studied.

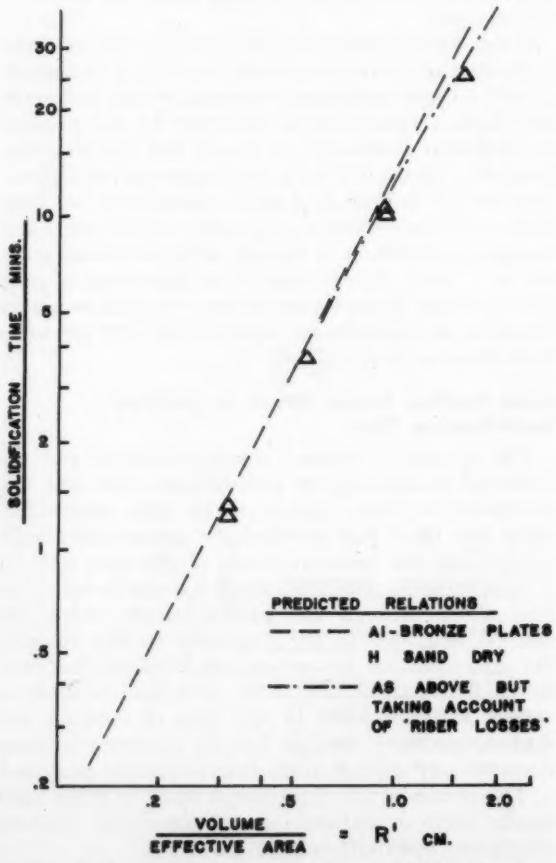


Fig. 3-10 — Effect of taking account of riser losses into heat dissipation.

(Fig. 3-7) in this latter case might possibly indicate the pattern of the dried-out portion.

SUMMARY AND CONCLUSIONS

The following summarizes the findings of the present work:

1) Using a mold calorimetric method, the mean heat diffusivity, b_g of several molding sands in green and dry states and in several conditions of ramming have been calculated and graphed.
2) Theoretical relationships between b_g and ramming density ρ_g have been obtained, derived from the methods of Gittus and Russell. The agreement between these relationships, and the results obtained by the mold calorimetric method, are as good as the present state of the theory permits.
3) Using the values of b_g obtained by measurement, the prediction of solidification times of plates of various sizes in three types of molding sands has been carried out with a good measure of success.
4) A potential probe technique has resulted in a satisfactory explanation being advanced to account for the variation of chilling power b_g of green sand molds with casting section thickness.

APPENDIX — HEAT LOSS THROUGH GATING SYSTEM

The method used to obtain a first or second order approximation of the amount of heat lost through gates and risers of the casting made was briefly as follows. The temperature distribution along a thin cylinder attached to a large mass (solidifying at a time greater than any of the castings to be considered) was measured semi-continuously with apparatus described previously (Fig. 3-11). From Fig. 3-11 it is possible to obtain the temperature gradient through the intersection of the two bodies at various times; by means of a graphical integration a mean value of this gradient over the time concerned (for example: $\rho - 10$ min 98.4 C/cm.

If then such a mean value is inserted into the equation:

$$\frac{\partial Q_g}{\partial t_{x=0}} = k_l \left(\frac{\partial \theta}{\partial x}\right)_{x=0} \quad \text{in cm/sec/cm}^2,$$

the rate of heat loss may be obtained, provided a value of k_l (the mean thermal conductivity) is available. The arithmetic mean of k_l at room-temperature* and at the melting point† should be a fair approximation (0.15 cgs units).

Since a mean value of $\left(\frac{\delta \theta}{\delta X}\right)_{x=0}$ has been assumed for the range of times considered, we are justified in writing:

$$Q_{g_t} = k_l \left(\frac{\Delta \theta}{\Delta X}\right)_{x=0} \cdot t$$

This quantity may then be subtracted from the total heat of solidification for the casting on pouring as in equation (lc).

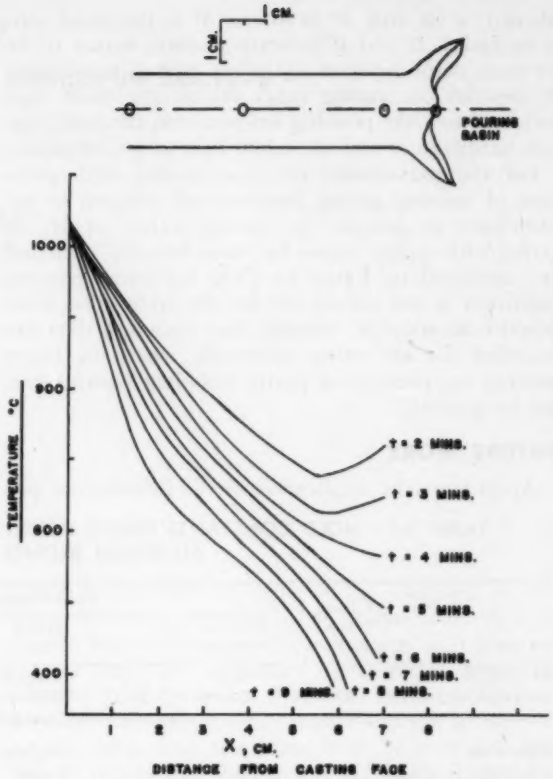


Fig. 3-11 — Temperature gradients developed in a narrow cylinder connecting a large casting with a pouring basin.

For the particular range 0 to 10 min with the ingate used:

$$\frac{\Delta Q_g}{\Delta t} = 30.5 \text{ cal/sec}$$

REFERENCES

1. Inst. of Brit. Foundrymen, Rept. of TS 21, 33, and 46, *Proc. I.B.F., 48*, p. A 128, 1955.
2. N. Chvorinov, *Giesserei, 27*, pp. 177, 201, 222, 1940.
3. R. Ruddle and A. Mincher, *Jnl. Inst. of Metals, 76*, 50, p. 43, 1949.
4. D. Atterton, *Jnl. Iron Steel Inst., 174*, p. 20, 1953.
5. R. Ruddle and R. Skinner, *Jnl. Inst. of Metals, 79*, p. 35, 1951.
6. M. Russell, *Jnl. Am. Cer. Soc., 18*, p. 1, 1935.
7. A. Eucken, *V.D.I. Forschungschaft (B), 3*, p. 352, 1933.
8. International Critical Tables.
9. A. Bell, AFS Transactions, *56*, p. 366, 1945.
10. J. Gittus, *B.C.I.R.A. Jnl. Res. and Dev.*, June 1954.
11. J. Gittus, *B.C.I.R.A. Jnl. Res. and Dev.*, April 1954.
12. R. Dunphy and W. Pellini, AFS Transactions, *59*, p. 425, 1951.

SUMMARY

The present paper has shown how, within the limits of accuracy that are imposed by simplification of the theory, and the lack of totally suitable thermal constants, the solidification times t_s of simple shapes can be related to the effective volume-to-surface area ratio R' by means of the simple equation

$$t_s = M_{s,c,F} (R')^2$$

*A.S.M. *Metals Handbook*.

†R. Ruddle and A. Mincher, *Jnl. Inst. of Metals 78*, p. 229, 1950-51.

where t_s is in min, R' is in cm, M is the mold constant and S, C and P indicate separate values of M for each shape (spheres, cylinders, plates) considered. M depends on, among other things, the mold type and hardness, the pouring temperature, the solidification temperature and the latent heat of solidification.

For the convenience of those dealing with problems of risering, gating, pouring and progress of solidification in general, the exteme values of M, together with typical values for "sand-foundry" practice, are tabulated in Table 3-1 (The turbulent pouring condition is not considered for the Al-Si alloy since practice should not warrant this, although data are included for the other materials, where in larger castings the presence of partly turbulent pouring cannot be ignored).

FUTURE WORK

Apart from the application of the information presented in the present paper to such problems as feeding, control of casting stresses, knock-out time calculations and the correlation of solidification time with mechanical properties, much ground remains to be covered in the basic sphere of heat transfer. In particular, the interrelation of heat loss and degree of turbulence for various shapes gated in differing modes is worthy of considerable study. Finally extension of work concerning mold chilling power and properties of liquid metals to cover the fullest possible range of materials is required.

ACKNOWLEDGMENT

The authors are grateful to many people at the University of Birmingham, England, particularly Prof. A. J. Murphy and Prof. E. Rollason, for making financial support available. Also to Mr. John Gittus, the Mond Nickel Co., Ltd.; the Directors and Management of Tangyes, Ltd., particularly Mr. John Grice, for providing iron casting facilities; and to the staff of Armour Research Foundation, Chicago.

TABLE 3-1 — MOLD CONSTANTS UNDER VARIOUS CONDITIONS FOR ALUMINUM-SILICON EUCTECTIC, ALUMINUM BRONZE, CAST IRON AND STEEL

Alloy	Filling Temp.[a] C (F)	Mold Conditions: Green or Dry	Grain Size	Soft or Hard Rammed	Mold Constants[b]: M_s	M_c	M_p
Al–13% Si		Dry	Medium	Medium	8.07	8.75	11.86
eutectic	690	Green	Medium	Medium	4.79	5.24	7.26
	(1274)	Dry bonded magnesite (1.95 g/cc)[e]			1.05	1.15	1.60
Aluminum bronze		Dry[d]	Medium	Medium	7.18	7.84	11.31
		Dry[d]	Coarse	Hard	4.26	4.64	6.54
eutectic	1240	Dry[e]	Fine	Soft	8.56	9.50	14.82
	(2644)	Plaster sleeving (0.745 g/cc)			12.97	14.62	28.39
		Dry bonded magnesite (1.95 g/cc)[e]			1.89	2.05	2.83
	1180	Dry[d]	Medium	Medium	6.04	6.58	9.24
	(2156)	Dry[d]	Coarse	Hard	3.59	3.91	5.33
		Dry[e]	Fine	Soft	6.57	7.73	12.23
		Plaster sleeving (0.745 g/cc)			11.10	12.63	23.21
		Dry bonded magnesite (1.95 g/cc)[e]			1.72	1.70	2.31
Cast iron		Dry[f]	Medium	Medium	6.66	7.27	10.36
(eutectic		Dry[f]	Coarse	Hard	3.97	4.29	5.98
carbon equiv.)	1290	Dry[g]	Fine	Soft	7.98	8.85	13.74
	(2354)	Dry bonded magnesite (1.95 g/cc)[e]			1.75	1.86	2.58
	1200	Dry[f]	Medium	Medium	4.13	4.45	6.09
	(2192)	Dry[f]	Coarse	Hard	2.52	2.70	3.52
		Dry[g]	Fine	Soft	4.91	5.64	8.08
		Dry bonded magnesite (1.95 g/cc)[e]			1.11	1.18	1.52
0.1 - 0.2% C	1585	Dry/green	Coarse	Hard[h]	1.28	1.38	1.99
Steel	(2885)	Diatomaceous earth and sand[i]			9.22	10.47	19.11
	1485	Dry/green	Coarse	Hard[h]	0.87	0.95	1.27
	(2705)	Diatomaceous earth and sand[i]			6.48	7.27	12.16

a) Filling temperatures represent:
690 (1274) — most quiscently poured small shapes in Al-Si, P.T. 700-720 C (1292-1328 F).
1240 (2644) — most quiescently poured shallow shapes in Al bronze, P.T. 1250-1270 C (2282-2318 F); also, top-poured deep shapes (S and C), P.T. 1310-1330 C (2390-2426 F).
1180 (2156) — top-poured deep shapes in Al bronze, P.T. 1250-1270 C (2282-2318 F); also, most quiescently poured shallow shapes, P.T. 1190-1210 C (2174-2210 F).
1290 (2354) — most quiescently poured shallow shapes in cast iron, P.T. 1300-1320 C (2372-2408 F); also, top-poured deep shapes, P.T. 1390-1410 C (2516-2570 F).
1200 (2192) — most quiescently poured shallow shapes in cast iron, P.T. 1210 C (2210 F); also, top-poured deep shapes, P.T. 1300 C (2372 F).
1585 (2885) — most quiescently poured shallow shapes in low-C steel, P.T. ~1600 C (~2912 F); also, top-poured deep shapes, P.T. ~1700 C (~3092 F).
1485 (2705) — most quiescently poured shallow shapes in low-C steel, P.T. ~1500 C (~2732 F); also, top-poured deep shapes, P.T. ~1600 C (~2912 F).

b) $t_s = M_{S,C,P}(R')^2$; M = t_s for R' = 1 cm. S = spheres, C = cylinders, P = plates.
c) See Ref. 3, Part 3.
d) Applies also to green molds over ~½ in. section.
e) Applies also to green molds over ~1 in. section.
f) Applies also to green molds over ~⅜ in. section.
g) Applies also to green molds over ~¾ in. section.
h) See Ref. 1, Part 1.
i) See Ref. 33, Part 1.

1959

Charles Edgar Hoyt Memorial Lecture

THE CONTROL OF QUALITY IN THE BRASS FOUNDRY

By Harry M. St. John

INTRODUCTION

It will be obvious to all that the subject of this lecture cannot be given exhaustive treatment in 30 or 40 minutes. One would have to include a detailed discussion of foundry practices, testing methods and an analysis of defects, any one of which would be more than sufficient to fill our allotted time. What I hope to do is to emphasize what might be called the philosophy of foundry control with such references to its technology as will cover useful procedures not commonly practised.

The control of quality in the brass foundry has two objectives, differing somewhat in purpose and in the methods employed.

First of all, the foundryman must guard against the delivery of unsatisfactory castings to his customers. If he fails in this, his reputation and, to some extent, the reputation of the whole foundry industry suffer.

Of no less importance, if the foundryman is to make a profit and continue in business, is the control of losses due to defective castings, particularly those which prove to be defective after machining or after assembly in a finished product. He must also realize that, if the percentage of rejected castings is large, many of those which pass all tests may be of substandard quality.

The captive foundry is fortunate in that it has a captive customer. Its castings are machined, assembled and tested in the plant which houses the foundry. Troubles can be quickly detected and corrective measures promptly taken. Partly for this reason, some jobbing foundries machine their own castings and give them a final test before delivery to customers.

SPECIFICATIONS

When castings are made to specification, such as are issued by A.S.T.M., S.A.E. and various departments of the Federal Government (Table), chemical composition and such physical properties as tensile strength and per cent elongation are emphasized. This information is of value to the customer as evidence that he is receiving what he has ordered; it is of little help to the foundryman except as it may indicate a trend in the quality of his metal. Even the customer must remember that there is no essential relationship between test-bar results and the quality of castings made from the same heat. A good test bar (Fig. 1) proves that the bar was well made from good metal. Castings from the same heat might be spoiled by any one of a number of faults, such as pouring at a wrong temperature. A poor test bar may be badly made from good metal.

Case History

A personal experience will serve to illustrate the point. The foundry was required to make a large number of manganese bronze valve stems of varying size. It was specified that test bars should have a minimum tensile of 60,000 psi, a minimum elongation of 18 per cent. But it was also specified that the maximum elongation should be 22 per cent, the explanation being that, in case of failure, the customer wanted the casting to break without too much stretch.

An unexpected difficulty was encountered. We

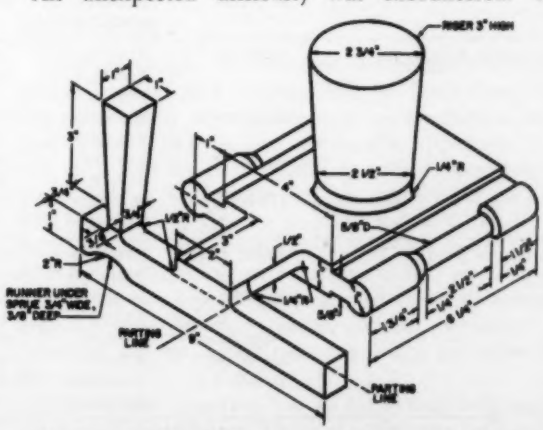


Fig. 1 — Double horizontal full-web type test bars (A.S.T.M. Designation B208-49T).

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TABLE — CONFORMING SPECIFICATIONS FOR COPPER-BASE ALLOYS

Class	Alloy	A.S.T.M.	Federal	Navy	Military	S.A.E.	A.M.S.
Tin Bronze	88-10-0-2	B143(1A) B30(1A)† B22-D				62	4845C
	88-8-0-4	B143(1B) B30(1B)†	691-5 701-5†	46M 6 46B25-III†	16576	620	
	85-15	B22-B	691-9 701-9†	46B22-III	16261		
	81-19	B22-A	691-10 701-10†				7322
	89-11					65	
Leaded Tin Bronze	88-6-2-4	B143(2A) B30(2A)† B61	691-1 701-1†	46B 8 46B25-I†	16541	622	
	87-8-1-4	B143(2B) B30(2B)†	691-6 701-6†	46B 5-I	16540	621	
	87-10-1-2						
	88-10-2-0					63	4846
	87-11-1-0-1					640	
High Lead Tin Bronze	80-10-10	B144(3A) B30(3A)† B22-C				64 792	4842 4827C
	83-7-7-3	B144(3B) B30(3B)†	691-12 701-12†	46B22-VI	16261	660	
	85-5-9-1	B144(3C) B30(3C)†				66	
	78-7-15	B144(3D) B30(3D)† B66	691-7 701-7†	46B22-IV	16261		4825A
	70-5-25	B144(3E) B30(3E)† B66-B67		46B22-V	16261		4840
	84-8-8		691-8 701-8†	46B22-II	16261		
	71-13-16		701-13†	46B22-VIII	16261		
	75-5-20	B66		46B22-I	16261	794	
	81-8-11	B66					
Leaded Red Brass	85-5-5-5	B145(4A) B30(4A)† B62	691-2 701-2†	46B23 46B25-II†	16444	40	4855A
	83-4-6-7	B145(4B) B30(4B)†		46B 5-III	16540		
Leaded Semi-Red Brass	81-3-7-9	B145(5A) B30(5A)†	691-4&11 701-4&11†	46B24 46B 5-III	16540		
	76-3-6-15	B145(5B) B30(5B)†					
	80-5-2-13		691-3 701-3†	46B21	16542		
Leaded Yellow Brass	71-1-3-25	B146(6A) B30(6A)†	621-C	46B11-II			
	67-1-3-29	B146(6B) B30(6B)†	621-B	46B11-I		41	
	63-1-1-35	B146(6C) B30(6C)†	621-A	46B10			
Manganese Bronze	60-20-15*	B147(7A) B30(7A)† B132-A	726-D 731-D†				
	65-25-20*	B147(8A) B30(8A)†	726-A 731-A†	49B 3 46B25-V†	16443	43	4860
	90-45-18*	B147(8B) B30(8B)†	726-B 731-B†	46B29-b 46B25-VII†	16522	430A	
	110-60-12*	B147(8C) B30(8C)†	726-C 731-C†	46B29-a 46B25-VI†	16522	430B	4862A
Aluminum Bronze	88-3-9	B148(9A) B30(9A)†	671-A 675-1†	46B18-1 46B19†‡	16033	68A	
	89-1-10	B148(9B) B30(9B)†	671-B 675-2†	46B18-2	16033	68B	
	(86-4-10 or (84-4-10-2	B148(9C) B30(9C)†	671-C 675-3†	46B18-3	16033		4870-71 4872-73
	79-5-11-5	B148(9D) B30(9D)†	671-D 675-4†	46B18-4	16033 6947		

(Continued on following page)

SPECIFICATIONS — Continued

Class	Alloy	A.S.T.M.	Federal	Navy	Military	S.A.E.	A.M.S.
Nickel Silver	12% Nickel	B149(10A) B30(10A)†					
	20% Nickel	B149(11A) B30(11A)†		46C12			
	25% Nickel	B149(11B) B30(11B)†					
Silicon Bronze	1-5% Si 5% Zn max.	B198(12A)	C-593	46B28-b			
Silicon Brass	2½-4% Si**	B198(13A)		46B28-a			
	3-5% Si**	B198(13B)		46B28-a			

*Mechanical properties of the manganese bronzes.
**12 to 16% zinc.
†Specifications are for ingot.
‡46B19 superseded by QQ-B-675-1.
ASTM—Year suffix not listed.
Federal—All specifications prefixed by QQ-B except QQ-C-593.
Navy—Small letter suffix not listed.
Military—All specifications prefixed by MIL-B except MIL-M-16576. These Specifications supersede Navy Specifications in column to the left.

couldn't get the elongation down to 22 per cent. Various things were tried, without success. For example, the addition of a little tin resulted in brittleness and an elongation far below 18 per cent. The customer agreed to accept a bronze of higher tensile and it was thought that in this way a lower elongation could easily be attained. To our amazement we made test bars having a tensile of 100,000 psi and 30 per cent elongation. The customer's inspectors were very helpful. They suggested that test bars be machined from castings, selecting for the purpose castings which would otherwise have been scrapped. This solved the problem.

The matter of chemical composition has two differing aspects. First of all, if a specification is to be met, the principal elements in the alloy must fall within the prescribed limits. Chemical analyses take time. Usually they tell what happened day before yesterday, too late to be of immediate use to the foundryman. The spectroscope offers a rapid means of analysis which, in iron and steel foundries, can be used to correct composition while the metal is still in the furnace.

The ingot manufacturer makes good use of the spectroscope to measure the progress of refining in his large reverberatory furnaces. But in nonferrous foundries, where heats are small and frequent, it is not practical to wait 10 or 15 minutes while a spectroscopic analysis is being made. About the best the brass foundryman can do is to study the analyses as he receives them and from these determine what variations are likely to occur. He can then compensate by adding more zinc or by any other change which seems to be needed.

Specification Importance

The importance of chemical specifications varies with their interpretation. If a specification calls for a maximum of 86 per cent copper and a minimum of 4 per cent lead, castings containing 87 per cent copper and 3 per cent lead are likely to be rejected in spite of the fact that they may have superior properties. The foundryman should bear in mind that the control of quality means giving the customer what he asks for, not something better than he wants or needs.

Deviations of composition such as the one just mentioned are most likely to result from the mixing of alloys in the handling of gates and machine turnings. If this occurs it means that expensive metal is being downgraded into a less expensive alloy with a resulting increase in cost.

The second important aspect of chemical composition is the presence of impurities which may be detrimental or will at least influence the behavior of the metal in the foundry. This applies particularly to those alloys which are most sensitive to pouring temperature, the red brasses and tin bronzes. Here uniformity is the prime requisite. For example, iron up to a permissible maximum of 0.25 per cent probably has no specific bad effect on the metal, but if this is followed by another lot of ingot containing 0.05 per cent iron there may be as much as 50 F difference in pouring temperature. The foundryman would be saved trouble and loss if both lots of ingot analyzed 0.15 per cent iron. In these alloys nickel may be termed a desirable impurity but variations of this element within specification limits can produce wide swings in the best pouring temperature.

INSPECTION

Castings should be inspected before machining for surface quality and evident defects. They should be inspected again after machining for machine spoiled and foundry defects not previously apparent. This can very well be part of the assembly operation. If inspection is well done the percentage of castings which leak when given a pressure test will usually be small, in the neighborhood of one per cent. If the foundry machines its own castings it may pay to pressure test a small percentage of each lot after machining. Then, if an exceptional number of leakers is found in any lot, a 100 per cent test is advisable for that lot.

Dimensional accuracy has an important bearing on the cost of machining. There must always be enough stock so that the casting will "clean up" to conform

with the blueprint. Excess stock is a needless cost to both foundry and machine shop. Modern sand conditioning and molding practice make possible close dimensional tolerances. Sometimes it is even possible to eliminate a machine operation previously thought necessary.

Marked improvements have recently been made in green sand molding. For many years the American Foundrymen's Society has made an intensive study of sand properties and sand conditioning. While brass foundries were a little slow to take advantage of the information thus made available, most foundries now employ sand testing equipment and do a reasonably good job with it. Much less attention was paid to the making of molds until the advent of shell molding, a process which makes possible the attainment of closer tolerances than previously found in brass foundry practice. While some brass castings are being made by the shell process, the principal impact of shell molding in the brass foundry has been to stimulate improvements in green sand molding.

It has been learned that the use of higher pressures and stronger sands result in finishes as fine and tolerances as close as can be obtained by any other method except the "lost wax" process. Where a mold hardness of 75 was formerly the usual thing and is still commonly found, most advanced practice calls for a hardness of 90 to 95. For larger castings the diaphragm method has been a factor in the improvement.

To obtain closer tolerances it is also necessary to make patterns more closely to prescribed dimensions. Allowing "plenty of finish" so that surface defects could be machined away was never good practice, is now less acceptable than ever.

SCRAP STUDY

There is no such thing as perfection in any manufacturing operation. The foundry is no exception. Competent testing and inspection will always find rejects but these must be held to a minimum if the foundry is to maintain a competitive position with other foundries and with other methods of forming metal parts. In the making of small castings, where a considerable number of molds is to be poured from one crucible or ladle, an overall loss of 5 per cent of the gross casting weight is a reasonable goal which should at least be approached. Failure to do so can easily mean an excess cost of several cents per pound of finished castings.

At this point it may be well to mention a fallacy which still lingers in some foundries. In these it is believed that tests and records are a waste of time and money, that a good foundryman, by the exercise of intelligence and experience, will produce a maximum of good castings, a minimum of rejects. Nothing could be further from the truth. Tests and records are among the foundryman's most important tools. To make best use of them requires the full employment of his skill and experience. Without them he suffers a severe handicap. I have in mind two instances where the study and recording of scrap greatly reduced scrap losses without any change in foundry personnel. In one case losses were reduced from 16 to 6 per cent, in the other from 12 to 3 per cent. Similar cases are numerous.

Troubles which are responsible for unsatisfactory quality may for convenience be divided into six classes:

1) Surface blemishes, which, while they may not interfere with the function of the casting, do spoil its appearance.
2) Misshapen castings not only give a poor appearance but often cannot be machined to the dimensions desired for the finished casting.
3) Castings in which the metal does not completely fill the mold cavity.
4) Castings which have internal shrinkage cavities resulting in lack of strength and pressure tightness.
5) Nonmetallic inclusions embedded in the casting.
6) Castings of substandard structure due to gas porosity or granular coarseness.

There are several definite points where foundry operations should be tested and controlled in order to keep casting defects at a minimum. These are concerned with the quality of the metal, the molding sand and cores and with the mechanics of making and pouring molds.

METAL QUALITY

It is first of all essential to ensure that the quality of metal, as it goes to the melting furnaces, is good. Watchful care should be used to make sure that foundry returns are clean and dry. A little molding sand on gates and sprues does not seem to matter greatly but should not be excessive. Scrap castings should be free from grease and broken tools. Machine turnings should be free from tramp iron and organic substances and must be dry, without oil and cutting compounds. Metal spillings and skimmings should preferably have been cleaned by a concentrating operation and then dried. The importance of not mixing alloys of differing composition has already been mentioned.

Most brass foundries use composition ingot in quantity sufficient to make up the weight castings shipped plus metal losses. While the chemical composition of ingot should not be neglected it is not the only factor which determines ingot quality.

When ingot is purchased in large lots performance tests are sometimes made. Sample castings may be made and fractured. Test bar results are useful. One procedure which may occasionally be worthwhile is to make a series of test blocks, poured at intervals from 2300 F down to about 2050 F. The specific gravity and Brinell hardness of these blocks is then measured. Taking the red brasses and tin bronzes as examples, a specific gravity of 8.9 is as near perfection as one is likely to get. Anything above 8.6 is acceptable; lower values indicate objectionable porosity. If a specific gravity of 8.6 or better is obtained over a temperature range of 150 F the metal is of satisfactory quality.

Interpretation of Brinell readings will vary according to the composition of the alloy and can only be standardized by experience. Exceptionally low read-

ings indicate porosity, high readings brittleness. Deviations from normal in metal which has a good conventional analysis are not uncommon. The reasons for such deviations are a matter of speculation rather than definite knowledge.

A number of other methods for judging ingot quality have been suggested. In the deep etch test a cross section of the ingot is obtained by sawing and the section surface etched by acid. In my opinion the results of this test are not particularly enlightening. In the making of ingot there is no point in trying to obtain a dense, fine-grained structure and the most efficient methods of production are not such as to promote such a structure.

The presence of coarse dendritic crystals in the interior of the ingot is not evidence of inferior quality. Even a gassy ingot does not necessarily produce gassy castings, although it is better to specify that the top of the ingot should show shrinkage. Ingot with puffy tops should be avoided.

The relative merits of rough-top and smooth-top ingot are a matter of controversy. The argument in favor of the smooth-top ingot is that oxides and nonmetallics which might later cause trouble have been skimmed away. In favor of the rough-top it is said that the roughness is mostly copper oxide, a substance which, as a useful ingredient during the melting operation, is often added in the furnace. As a matter of personal experience, I have never been able to decide that there is any real difference in the results obtained with one or the other.

Fuel vs. Electric Furnace

Metal quality can be greatly impaired during melting. The precautions to be observed depend upon the type of furnace in use. With fuel-fired furnaces a slightly oxidizing atmosphere must be maintained but excessive oxidation is to be avoided. In electric furnaces, where a neutral atmosphere is desirable, reducing conditions due to the presence of finely divided carbon must be guarded against. In any furnace type, the presence of water vapor, whether from the metal charge or the furnace lining, can be very damaging.

Fluxes should be used sparingly and for a definite purpose, not as a matter of habit. In theory the best metal is that which is slightly oxidized during melting, then deoxidized before pouring (Fig. 2). In fuel-fired furnaces a slight degree of oxidation is attainable by control of the combustion gases. In electric furnaces, which normally have a stagnant neutral atmosphere, it is sometimes advantageous to use a mildly oxidizing agent such as manganese dioxide, copper oxide or the sulphates of barium or sodium. Their use is called for only when the metal has been exposed to a reducing condition such as the presence of oil, carbon dust or water vapor. All of these fluxes tend to shorten the furnace lining life.

The introduction of air for oxidizing purposes in electric furnaces is not recommended. Fluxes of the type mentioned are also sometimes employed in crucible furnaces. The use of borax in crucible furnaces helps to keep the crucible clean and to accumulate the dross and slag so that they can more readily be skimmed from the metal. Regardless of the type of melting employed the metal should be deoxidized before pouring by about 2 oz of phosphor copper per 100 lb of metal and thoroughly stirred prior to skimming.

As a matter of routine, melting practice can best be judged by examination of castings but, when trouble is encountered, important information can be gained by pouring test blocks which are then subjected to tests for specific gravity, Brinell hardness and fracture.

TEMPERATURE CONTROL

Although the importance of temperature control has been emphasized for many years and almost every brass foundry owns a pyrometer, practice in this respect often falls far short of what it should be. If a single large casting is to be poured, the foundryman usually has considerable leeway. Precise temperature control is of secondary importance as compared with the size and location of gates, risers, chills and insulating pads.

As a rule, however, the brass foundry's principal production consists of small castings, where the objectives are the largest possible number of castings in each mold and the largest possible number of molds poured from each crucible or ladle. Here temperature control is of prime importance.

Under average conditions we may assume that a well-designed, well-gated red brass casting has a temperature pouring range of 150 F. Molds poured at either higher or lower temperatures will produce excessive scrap. Poor gating or design will reduce the range as will substandard metal. During pouring the metal will cool at a rate of 0.8 to 1.0 F per second, depending upon the insulating quality of the crucible or ladle. The time required for pouring will vary from 8 to 10 seconds per mold. It appears then that we can pour only some 15 molds, or possibly 17 or 18 under most favorable conditions, and still keep within the pouring range of the individual casting. Roughly speaking, 10 F is equivalent to one mold and, if we could, we would like to measure the temperature that closely. The best pyrometers available are not

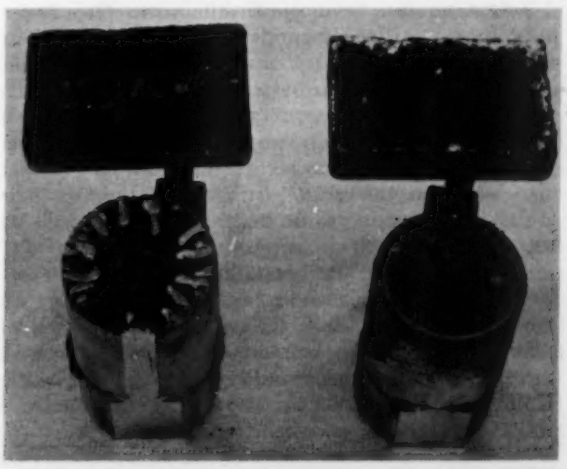


Fig. 2 — The riser on the left was from a heat that was not deoxidized. That on the right was deoxidized. Both were blind risers without vents.

ROUGH CASTING SCRAP						DATE 6-12-56
ITEM	PATTERN NO.	MOLDER NO.	DEFECTIVE NUMBER	DEFECTIVE WEIGHT	CODE NO.	REMARKS
11/2" 602	1012	12	50		1	
CENTERPIECE			42		2	
			36		5	
TOTAL			128	116		

Fig. 3 — Example of typical daily scrap report.

11/2" 602 CENTERPIECE PATTERN 1012			FILE RECORD															
DATE	MOLDER NO.	CSTGS. MADE	GROSS CSTGS.	%C.R. SCRAP	%SHOP SCRAP	TOTAL % SCRAP	DEFECTIVES											
							1	2	5	6	14	18	20	22	M	TOTAL
JUNE 56	12	2018	1996	24.5	10.3	34.8	200	153	138	18	108	50	50		34	696
JULY 56	12	1600	1590	5.2	2.8	8.0	20	12	6	15	20	13	12	20	10	128

Fig. 4 — File record form is permanent record of individual items produced.

quite that accurate although they can be read to 10 F.

SCRAP RECORDS

If we are to keep scrap at a minimum, then, the following conditions must be observed:

1) The quality of the metal must be good.
2) The design of the casting must be good.
3) The casting must be well gated.
4) The pouring temperature range must be known.
5) The pouring temperature must be closely measured and kept within the known range.
6) We must not pour any more molds from one ladle than variations in the above requirements permit.

Perhaps the most difficult problem is to determine the pouring range. This is a matter of trial and error, based on experience and good judgment. Each casting is an individual problem, to be solved by a study of the casting and its arrangement on the pattern plate. Incidentally, this arrangement and the gating are a job for the foundryman rather than for the patternmaker. Once decided upon, the pouring range should be recorded and the record consulted whenever the casting is made. It is obvious that detailed scrap records are here an essential tool.

Before we can have worthwhile scrap records we must have a skilled diagnosis of scrap, made by men who understand the various scrap causes and can identify each for what it is. One procedure which has worked well is to have the sorter throw out all castings which definitely are, or may possibly be, rejects. This accumulation is then inspected by a skilled man who lists the scrap castings according to their faults and sets aside those which can be salvaged by corrective work.

The scrap sorter's figures are combined and tabulated so that each casting item has in the file a card showing its performance. Its history over a period of time provides a basis for decision when a change in gating or pouring temperature is to be considered. Figure 3 illustrates a card on which scrap data are collected; Fig. 4 a permanent record.*

*Figures 3 and 4 are reproduced from *Brass and Bronze Foundry Practice* by H. M. St. John. Courtesy of Penton Publishing Co.

While the collection and study of scrap information constitute an important procedure for scrap control the results are too slow to catch troubles as they occur. For this purpose nothing so far available equals facture tests. The systematic test, developed under the auspices of the AFS Brass & Bronze Division's Research Committee, is useful although its interpretation is somewhat difficult. It principal contribution has been to our basic knowledge.

In the average brass foundry nothing has taken the place of fractured castings, examined in the light of long experience. The color and structure of a freshly fractured casting are difficult to describe for interpretation by a novice but to the veteran they tell a story he can get in no other way. The location of the fracture is important as is the manner in which it is made. A crushing machine is a useful adjunct to the foundry's equipment. Lacking this, hammer and vise will serve.

MOLDS AND MOLDING

Molding sand and the manner in which molds are made are important factors in their effect on casting quality. Up to the date of its publication, the AFS FOUNDRY SAND HANDBOOK covers the matter thoroughly but there have been many recent developments which can only be found in current literature. The present trend in green sand molding is toward harder molds which yield castings of closer dimensional tolerances. The frequent use of a mold hardness tester is indispensable. Whether or not facing sand is used, the properties of that portion of the sand which comes in contact with the pattern must be carefully controlled and periodically tested. One of the most important requirements is to keep this sand free from clay balls and from fragments of metal.

For thousands of years makers of castings used molding sand as they found it in nature. Now this practice is changing, for two reasons. First, natural supplies of suitable sand, with the right proportions of desired grain size and shape, blended with the needed amount of the right kind of clay, are becoming scarce. Next, it has been learned that nature's sands can be improved by blending with each other and with additional elements. The highest standards of quality now prevailing can hardly be achieved in any other way.

Brass and bronze require fine-grained sands. For a first class finish on small castings the base sand should have an AFS fineness of 150 with closely grouped screen size. In practice a certain amount of coarser core sand becomes mixed with the molding sand, thus decreasing the fineness number of the sand in use to a possible 130. The sand grains should preferably be rounded since the tendency of sharp sand to pack makes permeability control difficult. If all grains were perfectly round and of the same size, regardless of what that size might be, the percentage of voids between the grains would be 33⅓ per cent. The more closely the sand is grouped in a few adjacent screen sizes the more nearly we approach this condition.

Finer sands will still have a somewhat lower permeability because of the more tortuous path gasses

must follow in passing through the sand. Any excess of clay or other bonding material will tend to fill the voids and reduce permeability. The addition of a finer sand will also tend to fill voids and decrease permeability but the addition of a coarse sand will not increase permeability unless the addition is so great as to make the coarse sand dominant.

The so-called synthetic sands have many advantages. They permit a control of molding sand properties more complete than is otherwise possible. For this is required a sand almost if not quite free from clay to which a measured quantity of a selected bond and such other substances as may be desired are added. This has become common practice in iron and steel foundries for whose purpose suitable sands are readily available. Some of these sands work very well for the high-nickel nonferrous alloys and even for brass and bronze castings when these are large, say 100 lb or more. With a well-grouped AFS fineness of 100 to 110, a clay content of one or two per cent, a satisfactory finish can be obtained on these and also on manganese bronze.

But for small brass castings sufficiently fine sands with a low clay content are scarce and costly. A few are available having an AFS fineness of about 150 and a clay content of 5 or 6 per cent. These can be used to good advantage adding bentonite as the principal bond. Crushed sandstone, screened to the required size, is also a possibility.

The brass foundryman has one other recourse which, in many cases, is his best bet. He can take a fine sand which is too rich in clay and blend it with used sand which has been passed through a fine screen to remove metal particles. Foundry floor sweepings and sand which has been spilled from poured molds are suitable for this purpose. They will be low in bond and sufficiently dry, two per cent moisture or less, for fine screening. Heap sand which has been dried to this point may also be used.

Facing Sand Use

To an increasing degree the brass foundryman is finding that he can most effectively get the results he wants by using facing sand of such a mixture as described above. This is much more practical than to process the entire bulk of his molding sand, most of which will simply serve as backing and need not have such specialized properties.

For best results a naturally bonded sand should have a permeability of 15 to 20, compressive strength of 6 to 8, tempered to about 6 per cent moisture. A good synthetic sand will show a permeability of 25 to 30, a strength of 10 to 12, moisture content of 4 per cent or less. In either case about one per cent of flour, either cereal or wood, should be added to give the sand elasticity. Other additives, such as soda ash and ethylene glycol, show promise but are still in the development stage. The same may be said of sand tempered with oil or synthetic plastics in place of water. Molding pressure is important. For squeeze or jolt-squeeze machines, the air pressure at the machine should be at least 100 psi.

Care of core quality is a matter which is too often neglected. The foundryman will do well to read Dietert's AFS classic* on this subject but again there are new developments in recent foundry literature. While 100 per cent inspection of cores is seldom advisable, a partial inspection is well worth while, followed in cases of trouble by complete inspection. It is cheaper to throw away cores than it is to throw away castings.

CORE TROUBLES

Core troubles are usually due to two causes, either too porous a surface resulting in "burn-in," or excessive gas causing blows in the casting. Baking time and temperature are critical and must be adjusted to suit the size of cores as well as the character of binder used. Where high permeability is essential a coating may be needed to yield desired smoothness of finish. As another possible source of gas, coatings should be avoided except when absolutely necessary. Thorough baking without the use of too high a temperature and careful drying of cores which have been pasted or coated will help to avoid many of the troubles commonly experienced.

Cores should be used as soon as possible after making to avoid moisture pickup. Since it is convenient to have molds ready to pour at the beginning of the day's work, completed molds from the previous day are often allowed to stand overnight. One hesitates to condemn this practice but it does involve danger of moisture pickup from the molding sand. Certainly unpoured molds should not stand over a holiday or weekend.

Without going into detail it should be added that the likelihood of core trouble is lessened by dielectric baking or by the use of shell cores or carbon dioxide cores. While these methods have drawbacks peculiar to themselves, they are finding increasing use and, in the course of time, may substantially change coreroom practices.

STATISTICAL QUALITY CONTROL

How often should tests be made and how complete should be the inspection? First of all let us say that, so far as possible, no bad castings should ever be allowed to reach the customer. At intermediate stages too much testing may result in a waste of time and money. Statistical quality control, its application in the foundry described in a recent AFS publication, STATISTICAL QUALITY CONTROL FOR FOUNDRIES, points the way toward avoiding this. With respect to any foundry operation it is possible to determine the operation's quality potential. That is to say, how much deviation from a fixed standard is to be expected and can hardly be avoided.

If the test of a sample shows a greater deviation, whether in properties or in number of rejects, then we can assume that there is a definite cause for this excess, something which can be found and corrected. While the mathematics involved are somewhat difficult, the principles governing the needed size of sample and number of tests can readily be understood. Statistical quality control is not generally practiced in the brass foundry but its study will be of benefit to any foundryman.

*H. W. Dietert, MODERN CORE PRACTICES AND THEORIES, American Foundrymen's Society, Des Plaines, Ill.

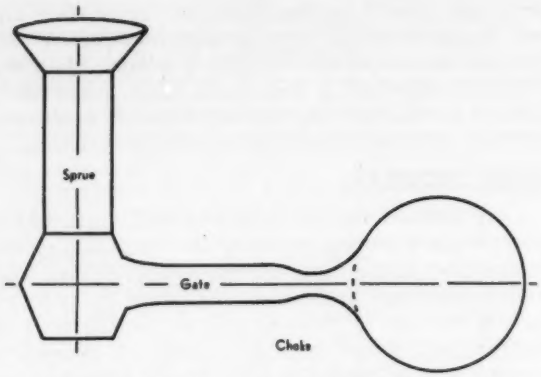


Fig. 5 — Choke gates are in general undesirable since they cause turbulence in the casting cavity and, except for light castings of uniform section, are likely to cut off the feed before the casting is entirely solid. As often used, choke gates are a prevalent cause of scrap. A choke at the bottom of the sprue is preferred.

GATING

The gating of red brass and bronze castings continues to be an art rather than a science (Fig. 5). Scientific progress has been made in the study of mathematical principles involved in the gating of metals such as steel and aluminum which have short freezing ranges, but the formulas derived for these metals are seldom of much use in the brass foundry where long freezing ranges are the rule.

The shrinkage characteristics of the red metals are such that small castings do not ordinarily require risers. On the other hand, the critical distances through which metal will feed from a gate are short so that multiple gating is commonly necessary. In determining the best practical gating one must consider the design of the casting, the arrangement of the castings on the pattern plate and the pouring temperature range which can be used without causing excessive scrap.

In most cases it is advisable to set up a new job with temporary gating designed according to the foundryman's best judgment, with the expectation that at least minor changes will have to be made after trial of the temporary gating. During this stage the pouring temperature should also be given careful study. Once established, the gating should not be changed without positive evidence of necessity. If there appears to be such evidence it should be examined carefully because of the likelihood that metal quality or pouring temperature, rather than gating, is at fault.

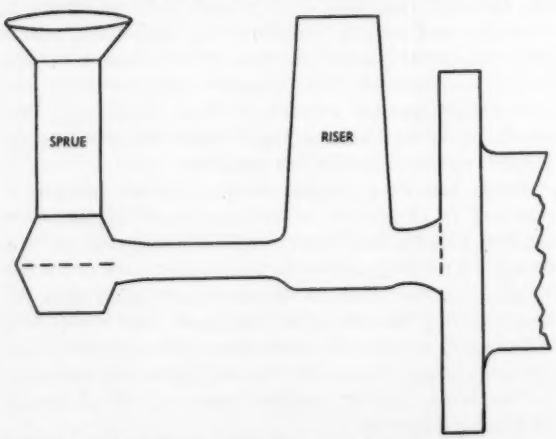


Fig. 6 — Method of locating riser to receive the last and hence the hottest metal, thus insuring adequate feeding.

The foundryman is under pressure to produce the highest possible casting yield, that is, the largest number of castings which can be made from each 100 lbs of metal melted. As a means to this end, nonproductive metal, in the form of gates, risers and sprues, should be reduced to a minimum (Fig. 6). Obviously if one goes too far in this direction excessive scrap will result and the yield of good castings will actually decrease.

The foundryman has a natural tendency to play it safe by making sure that there is plenty of metal. The closer one arrives at the highest possible yield the more necessary it becomes to maintain high metal quality and to exercise precise control over pouring temperatures, since both of these influence the adequacy of casting feeding.

Multiple Gating

Where multiple gating is employed the application of directional solidification becomes complicated and cannot readily be expressed in mathematical terms. It is commonly stated that the cross section of the gate should be 1½ times the casting section which it feeds, but with multiple gates it sometimes happens that one of the gates should have a cross section *smaller* than that of the casting. Where risers are used one finds recommendations for height ranging from 2½ to 1½ times the riser diameter, with approximately 25 per cent of the volume available for feeding purposes.

If insulating sleeves or exothermic compounds are used, the height must remain the same but available volume increases and diameter decreases. One must also remember that a blind riser is less effective than one open to the atmosphere and that the most effective riser of all is one placed between the sprue and the casting. All of which means that the foundryman must use his own judgment and profit by experience.

In most brass-foundry work castings are so small that several can be made in one mold, sometimes a large number. The problem then is to design runner and gates so that all are properly fed. The pattern maker is inclined to exercise his ingenuity in an effort to get the largest possible number of castings into the mold but in so doing he may defeat his own end since, after all, the objective is to get the largest possible number of *good castings.*

If the added castings turn out to be scrap, there has been a loss rather than a gain. Here the foundryman's know-how should be brought into play. He, in his turn, should endeavor to include as many castings as can be satisfactorily fed. Sometimes the possible saving is such as to justify a small increase in scrap.

CASTING DESIGN

The foundryman is not a casting designer and the casting designer is not a foundryman. Co-operation between the two is essential if problems which are

sometimes almost insuperable are to be avoided. The designer usually has in mind machine shop requirements and tries to design the casting in such a way as to make machining as convenient and inexpensive as possible. In doing so he may unwittingly greatly increase foundry costs. Or, in order to decrease the weight of the casting, he may saddle the foundryman with difficulties far more costly than the saving. This is particularly true of brass castings because brass is expensive and the designer does not want to be accused of "giving away" metal.

A complicated casting may offer a real problem to both designer and foundryman but simple castings are often at fault in this respect. Let us take the case of a valve body. Here, in Fig. 7, we have a casting which is almost impossible for the foundry to make. The seat section *S* is $\frac{5}{32}$-in. thick, while metal is saved at *C*, *D* and *E* by reducing the section to $\frac{3}{32}$-in. There is now no way of feeding the seat section. By thickening the section at *C* and *D*, at the *parting line only*, the seat section can be fed without greatly increasing the weight of the casting. This is equivalent to incorporating feeding gates in the body of the casting itself. The foundryman encounters many instances where solutions as simple as this can be found if the designer can be induced to consult him.

Foundryman and Designer

The foundryman can be of help to the designer in a number of ways, including the quality of the casting as well as the cost of making it. His familiarity with the nature of castings and with the properties of the metal he is using will enable him to point out what are likely to be weak spots and suggest means for their elimination. Among the fundamentals are the use of ample fillets at corners and the avoidance of an abrupt change from a thin to a thick section.

Differences in metal properties must be taken into consideration. Familiarity with iron, steel or aluminum may mislead the engineer when he designs for brass or bronze. When the same item is to be made in more than one metal it is frequently necessary to accommodate the design of the casting to fit the differing properties of the metals, either by changing the design or by making the design such that a sound casting can be poured from any of the desired metals. The latter practice is usually more economical than to have a number of different designs and patterns.

The engineer will read in his handbook that $\frac{3}{32}$-in. is the minimum workable thickness for the red alloys, $\frac{1}{8}$-in. for those higher in zinc, but he must remember that these minimum thicknesses are good only for short distances. Where there is to be a considerable area of thin section, these minimums must

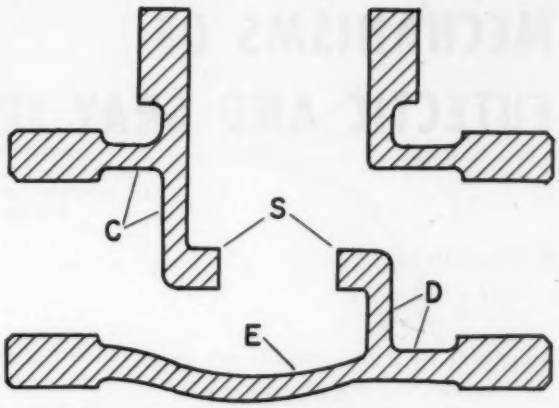


Fig. 7 — Sketch of valve body casting.

be substantially increased. Otherwise the pouring temperature range of the casting is cut down to a point where sound castings cannot economically be made.

The designer must also take an interest in gating since he must make sure that all parts of his casting can be fed with gates which can be fitted into a good pattern layout. For this reason there is some advantage in making a preliminary pattern layout before the design is made final in all details. It may be found that slight changes in design will make possible a better layout. When the casting is to be made in large volume a substantial saving is possible.

SUMMARY

To summarize, the foundryman needs to observe the following control points if he is to profitably serve a satisfied customer:

1) The quality and condition of metal going to the furnaces.
2) The quality and condition of molten metal from the furnaces.
3) The quality and condition of molding sand and molds.
4) The suitability of cores.
5) The soundness and condition of rough castings delivered to the machine shop.
6) The soundness and accuracy of castings delivered to the customer.

In support of this program, the foundryman should play a decisive part in the gating and pattern layout of his castings and act in an advisory capacity with respect to design.

Following such a schedule will cost something in both time and money but will yield returns out of all proportion to the cost.

SOLIDIFICATION MECHANISMS OF EUTECTIC AND GRAY IRON

By Albert De Sy

ABSTRACT

Any metallurgical transformation whose initial and final states are known but whose mechanisms are not established is a phenomenon not under control. Under these conditions, systematic processing to obtain a predetermined result is not possible. Observation does sometimes permit the establishment of correlations between processing variables and their influence on the result. But as long as the mechanism of the transformation is not known precisely, any intervention will be arbitrary, and full possibilities of control will not be realized. When the mechanism is known, the desired result can be obtained with a minimum expenditure of time and effort.

INTRODUCTION

The mechanisms of metallurgical transformations are generally difficult to determine since the transformations mostly occur at high temperatures and on a small scale. It is claimed in some quarters that these problems are of only theoretical interest; however, experience has shown that research into these transformations has resulted in the discovery and improvement of industrial procedures. This research always will be of assistance although it may not pay off for a considerable period, and for this reason enough basic research is not being conducted. For example, some purely basic research conducted at the Belgian Center for Foundry Studies and Research on the mechanism of the graphitization of gray iron[1] has produced results with important industrial applications. These findings have been recently confirmed.[2]

For the next two years, the Belgian Center for Foundry Studies and Research intends to undertake the study of the mechanism of gray iron solidification. This is a fundamental problem of gray iron metallurgy and appears to be well chosen. A study of the solidification of gray iron must consider the following items:

1) Initial or liquid state.
2) Actual solidification when both liquid and solid states coexist.
3) Final solid state existing immediately after solidification.

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The observation of the final state is hardly possible since we only see the structure at ambient temperature after it has been profoundly modified by cooling from the solidification temperature.

The liquid state or solution just above the melting temperature is a condition that is not perfectly understood. This liquid solution is called homogeneous. However, it is necessary to consider the degree of observation and the size of the sample which is termed homogeneous. A pure metal near the point of solidification contains pre-groupings, which are temporary associations caused by the chance movement of atoms or because of affinities between the atoms present in the liquid solution.

When the ternary Fe-C-Si system is considered, attractions between these elements produce a domain such that the Fe-C and Fe-Si pre-groupings predominate at the solidification temperature. It is probable that the latter destroy the former as the silicon content increases. The silicon, by attaching itself to the iron, liberates carbon, thereby increasing the activity of this latter element. This may probably explain the graphitization action of silicon in gray iron.

Solidification is a phenomenon of nucleation and growth. Nucleation is the predominating factor and deserves particular study. Nucleation can occur spontaneously or in a homogeneous manner, although this requires considerable undercooling. Nucleation can also occur from foreign nuclei, called heterogeneous solidification. The latter mode is the usual case, since undercooling values of 0.14 to 0.18 T_s (where T_s is the temperature of solidification expressed in °K) have been observed for homogeneous solidification. It is well known that the undercooling becomes more pronounced the greater the purity of the metal.

HOMOGENEOUS NUCLEATION

Homogeneous nucleation is the formation of a stable nucleus of the solidifying phase, by the assembly of a certain number of atoms of that phase in their crystallographic lattice. This nucleus is effective when it is capable of growing into a crystal. According to Volmer,[3] the effectiveness of a nucleus depends on its dimensions. Considering the solidification of a very small particle, the change in free energy may be expressed:

$$-\Delta F = -\Delta F_v + \sigma S \qquad (1)$$

where:

$-\Delta F$ = change in free energy of the system.
$-\Delta F_v$ = change in free energy resulting from the solidification of this elementary particle (change from liquid to solid state).
σ = surface energy per unit area at the liquid-solid interface.
S = surface area of the solid particle.

It is noted that ΔF_v and σS have opposite signs, since the latter term represents an increase in energy.

When the resulting $-\Delta F$ produces a decrease in free energy, the reaction or the solidification is spontaneous and will proceed. Solidification will not occur, however, when the resulting ΔF is positive.

Considering a particular shape of an elementary particle, for example a sphere, equation (1) may be developed in terms of the radius of the sphere. Assuming that the term ΔF_v is for a unit volume, then for a spherical particle of radius, r:

$$-\Delta F = \frac{-4\pi}{3} r^3 \Delta F_v + 4\pi r^2 \sigma \qquad (2)$$

ΔF_v and σ may be considered constant for a given metal at a given temperature and

$$-\Delta F = f(r)$$

where:

$-\Delta F$ is the algebraic sum of the two terms.

This function is represented graphically in Fig. 1. The function goes through a maximum whose value in terms of the variable r may be found by setting the first derivative equal to zero. Accordingly:

$$r = r_c = \frac{2\sigma}{\Delta F_v} \qquad (3)$$

The value r_c is called the critical radius. The interpretation of the facts in Fig. 1 may be summarized as follows.

Atom Movement

The chance movement of atoms in the liquid metal at or below the liquid-solid equilibrium temperature produces agglomerations of certain numbers of atoms. These agglomerations form at various locations in the melt, but are unstable at first since they form, disappear and re-form elsewhere. Considering an agglomeration of a volume equivalent to a sphere of radius $r<r_c$, the embryo may (depending on the chance movement of atoms in its vicinity) increase in size or may redissolve. Since the curve for $-\Delta F$ vs. r (Fig. 1) shows that any increase in r, when $r<r_c$, results in an increase in free energy, the chance that the agglomeration (embryo) will grow and attain the dimensions of a stable nucleus are practically nonexistent.

However, if the embryo resulting from the chance movement of atoms in the liquid phase is formed with dimensions equivalent to a sphere of radius $r>r_c$, any increase in radius corresponds to a decrease in free energy of the system and the nucleus will probably grow. The chance of continued growth increases as the radius r_s is approached. For this reason, r_c at the maximum point of the curve is called the critical radius. A nucleus which has attained or surpassed the dimension r_s is an absolutely stable or effective nucleus, and will practically always grow to a crystal. Homogeneous nucleation occurs with great difficulty, and only happens when undercooling has become sufficiently pronounced.

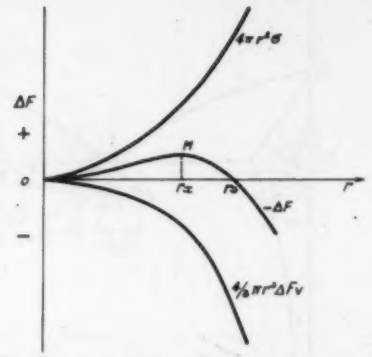


Fig. 1 — Change in free energy with the solidification of a spherical particle of radius r.

HETEROGENEOUS NUCLEATION (FOREIGN NUCLEI)

Industrial metals and alloys generally solidify with a small amount of undercooling. Since this solidification temperature is considerably higher than that required for homogeneous nucleation in pure metals, the role of foreign nuclei must be recognized. These nuclei may occur from reactions or precipitations of the elements present in the melt, or may be suspended solid particles entirely foreign to the metal. The importance of the crystal structure of the particle in this regard must be emphasized.

SINGLE PHASE AND POLYPHASE SYSTEMS

Pure metals and solid solutions are solidified by the nucleation of a single phase. Solidification is controlled by the number of effective nuclei per unit of time and volume, and by the rate of growth of the single solid phase. The preponderant influence of the nucleation factor is proved by the fine grain produced whenever the maximum number of nuclei are produced naturally (by solidification in pure metals with a large amount of undercooling) or artificially (by effective inoculation of metals and alloys).

Polyphase crystallization is more complex, since each phase must be nucleated separately. One phase *A* may solidify without difficulty and without undercooling, whereas another phase *B* may undercool considerably because of lack of effective nuclei. This will cause a lag in solidification that will considerably influence the structure of the alloy. It is also possible that phase *A* will act as a foreign nuclei for the nucleation of phase *B* and impose its orientation upon phase *B*, such as occurs in simultaneous crystallization. To remain within the scope of the title, only eutectic solidification will be considered as the most appropriate example of polyphase solidification.

EUTECTIC SOLIDIFICATION AND RESULTING STRUCTURES

Eutectic solidification, as with any industrial alloy, occurs with a small amount of undercooling

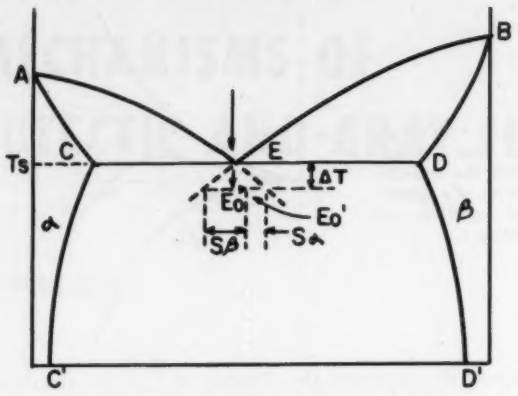


Fig. 2 — Solidification eutectic (arbitrary scale of ΔT).

(Fig. 2). Each of the two phases, α and β, require stable nuclei to solidify. Except for the exceptional simultaneous solidification of the two phases from the melt, or unusual instances when the same nuclei can serve both phases simultaneously, one of the two phases will be nucleated before the other. If the α phase solidifies first at a temperature of $T_s - \Delta T$ (when the liquid reaches point E_o in Fig. 2), then effective nuclei of the α phase develop simultaneously at numerous points in the melt at this temperature.

These nuclei initiate the solidification of α grains, and each is surrounded by a liquid zone impoverished in element *A* and enriched in element *B*. The average concentration of the volume of liquid surrounding each grain of α has been moved from E_o toward E_o'. As the solidification of α continues, the supersaturation of the liquid with respect to β will increase as the composition of the melt moves towards the right until sufficient supersaturation occurs to force solidification of β. Thus, eutectic solidification is initiated, and the first β crystals will be formed in the immediate vicinity of the α crystals.

Based on the foregoing considerations, two possibilities for eutectic solidification can be postulated:

1) Alternate solidification of the two phases of the eutectic occurring with alternate displacement of the composition of the surrounding liquid from left to right, and from right to left of the average composition.
2) Simultaneous solidification of the two phases with a common solidifying front advancing into the liquid, which is supersaturated with respect to the two phases.

Normal eutectic lamellae result from solidification following the second hypothesis; abnormal eutectics more closely follow the first hypothesis.

Normal Eutectics

The mechanism of normal eutectic solidification such as the Al-Zn eutectic and the ferrite-cementite eutectoid is well known. These are clearly instances of simultaneous growth of the two phases advancing in a common front. This mechanism is illustrated for the growth of pearlite into austenite in Fig. 3. It has been proved that pearlite is nucleated by the cementite, and obtains its orientation from the cementite.

This mode of secondary nucleation is applicable to all normal eutectics. The phase that is nucleated first nucleates the second phase, and imposes its orientation upon this second phase. As a result, the growth of normal eutectics obey the laws of M. Straumanis[4] and N. T. Belaiew,[5] which actually are only corollaries of the means of nucleation.

According to the results of the research of Straumanis[4], all the plates of the first phase have the same crystal orientation, and form a constant angle with the plates of the second phase. It is not an exaggeration to refer to a pearlitic grain or a eutectic grain of this type, since it is similar to two single crystals cut in slices and associated as alternating planes in the same crystal structure.

The law of Belaiew[5] is better known to the ferrous metallurgist. This law points out the constant interlamellar distance between the alternating plates of pearlite within the same grain. Furthermore, when the plates or lamellae are straight, it can be verified that the interlamellar distance is a function of the temperature of formation.



Fig. 3 — Crystallization front of the normal pearlite eutectoid. 30,000 ×.

Abnormal Eutectics

Partial or Total Envelopment. E. Scheil[6] cites five examples of abnormal (anomalous) eutectics: Austenite-graphite, Al-Si, Al-Al_3Fe, Zn-Zn_3Sb_2 and Pb-Ag. These five cases present common characteristics.

1) The primary or enveloping phase clearly predominates in mass and volume.
2) The secondary or minor phase has a higher solidification temperature than the primary phase. This difference is large with graphite and austenite and Al and Si, but is small with Zn and Zn_3Sb_2.

The laws of Straumanis and Belaiew do not apply to these anomalous eutectics. This is clearly not a case of simultaneous solidification of the two phases. If the conditions of simultaneous growth are not fulfilled, the importance of the time lag between the solidification of the two phases is of interest.

Thermal analysis of the solidification of abnormal eutectics shows the existence of a eutectic stage with characteristic undercooling, although it is slight and always followed by a rise in temperature. This indicates a delay in solidification of the secondary phase, shows also that the primary phase is not nucleating the secondary phase and shows the lack of nuclei for the secondary phase in the melt. These conditions hold for the five anomalous eutectics listed. In all of these cases, the anomaly results because the major phase is incapable of nucleating the minor phase.

Second Phase Solidification

When solidification of the second phase does begin, it is usually at a relatively small number of locations, indicating a small number of effective nuclei. However, the speed of solidification is rapid, and the secondary phase forms quickly. This rapid solidification produces a local supersaturation of the first phase in the eutectic liquid at the interface of the secondary phase and the liquid. This tends to cause envelopment of the second phase by the first.

The mechanism of this eutectic reaction is illustrated at the *A, B* and *C* stages in Fig. 4 for the solidification of plates or flakes of the second phase. Stage *A* represents the rejection of the first particle of the β phase. As a result of the solidification and growth of the β particle, a concentration gradient is established in the surrounding liquid with its steepest slope perpendicular to the surface of the β plate. This produces a solidification of the α phase, as shown under *B*, and the envelopment of β starts to become apparent.

As the β plate continues to grow, it spreads towards the limits of its zone of influence (limited by the existence of neighboring plates), and its speed of growth decreases permitting greater envelopment by the major phase. The stage of quasi-complete envelopment, marked by stopping the growth of the β plate, is finally achieved in *C* (Fig. 4).

The three phases, liquid, α and β are in contact during the entire period in which the β phase is growing. Cooling curves of this melt would show a hold during this period, since this is a real eutectic solidification. However, the appearance of the solidified structure is considerably different from that of



Fig. 4 — Schematic representation of abnormal eutectic crystallization. *A, B, C:* minor phase (β) in plates.

a normal eutectic which occurs with the simultaneous growth of the two phases. The structure of an anomalous eutectic is illustrated for the Al-Si eutectic in Fig. 5.

This photomicrograph also contains some primary silicon phase present as polyhedral crystals. The form and distribution of the silicon phase in the eutectic resembles the form and distribution of graphite flakes in gray iron.

Minor Phase Spheroid Solidification

If, instead of solidifying in plates or flakes, the minor phase solidifies as spheroids, or in some form without a preferential direction of growth, a more rapid envelopment by the major phase can be visualized. This will produce an abnormal eutectic with total envelopment, and will appear as numerous β particles in an α matrix. If, in the case of anomalous solidification with partial envelopment, the undercooling of the minor phase is forcibly increased, early total envelopment may result.

Undercooling of the silicon phase of the Al-Si eutectic can be accomplished by the addition of sodium. This is the result of poisoning of the foreign nuclei of the minor silicon phase. This modification in structure is universally employed to improve the mechanical properties of this alloy.

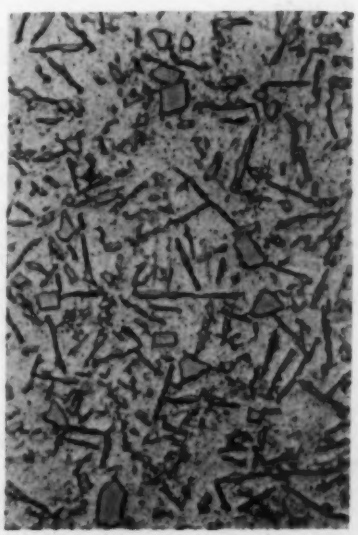


Fig. 5 — Abnormal Al-Si eutectics. 100 ×.

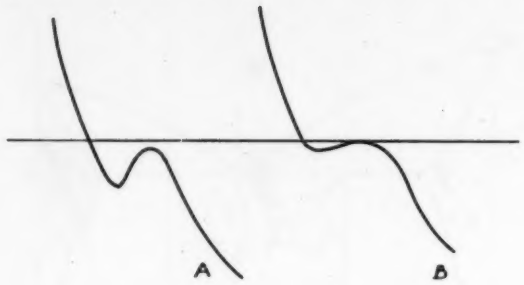


Fig. 6 — Schematic representation of the steps of eutectic solidification of irons *A* and *B*.

EUTECTIC SOLIDIFICATION OF GRAY IRON

(Anomalous Austenite-Graphite Eutectic)

If the cooling curves are traced for

A) a relatively pure, synthetic Fe-C-Si alloy, and
B) an industrial gray iron with comparable C and Si contents,

the cooling curves shown in Figs. 6a and 6b, respectively, are obtained. Examination of the graphite structure of these two irons will show:

A) Graphite of the supercooled type for pure Fe-C-Si alloy.
B) Normal flake graphite (such as Type A) for the industrial gray iron.

These results may be interpreted based on the preceding theory. If the two irons were hypoeutectic to the same degree (had the same carbon equivalent), eutectic solidification was preceded by the rejection of primary austenite. Eutectic solidification is not initiated when the equilibrium eutectic temperature is attained, since the state of nucleation of the eutectic liquid with respect to graphite is insufficient. However, the solidification of the austenite continues. The two irons have behaved in a similar manner up to this point, but the amount of supercooling obtained for the two is different. Even though this difference is small when expressed in degrees centigrade, it produces widely different results during eutectic solidification.

Case A

When the solidification of case *A* is considered, appreciable undercooling is proof of the absence of nuclei capable of initiating the solidification of the graphite. It is probable that the graphitization is started by spontaneous or homogeneous nucleation. This hypothesis, or one requiring the action of less effective foreign nuclei than those which cause the graphite nucleation in the industrial gray iron (Case B), explains the undercooling and the nucleation and simultaneous growth of the great number of nuclei with regular distribution, as observed in these undercooled irons.

The great number of graphite particles implies a small zone of influence of each and rapid envelopment by the major austenitic phase. This eutectic is definitely an anomalous one with quasi-total envelopment. It is definitely not a normal eutectic, and does not follow the laws of Straumanis[4] and Belaiew.[5]

The industrial gray iron *B*, or a pure Fe-C-Si alloy to which sulfur is added according to Boyles[7] demonstrates less undercooling, and solidifies as a eutectic of austenite and flake graphite. It behaves as an anomalous eutectic with partial envelopment similar to the Al-Si eutectic in Fig. 5. However, the real supercooling is greater than the apparent supercooling expressed as the temperature difference between the maximum and minimum of the eutectic arrest. It is considered, as a first approximation, that the mechanism of this austenite-graphite eutectic is that expressed in *A*, *B* and *C* of Fig. 4.

The apparent undercooling of the eutectic in the industrial gray iron (Fig. 6b) is small, and difficult to detect at times. This proves the existence of a good condition of nucleation for the graphite phase in the melt. Nucleation is probably produced by the presence of foreign nuclei which only require slight undercooling of the eutectic liquid to become effective. The effectiveness of sulfur in producing nucleation of flake graphite is noted by W. Patterson.[8] This work points out the frequent association of graphite flakes and the FeS or xMnS•yFeS inclusions.

However, if the industrial gray iron is superheated to a considerable degree, the ability to nucleate the graphite is lost, even though the sulfur content is unchanged and its distribution and appearance remain similar. It seems probable that the superheating may have destroyed or poisoned effective foreign nuclei other than the sulfur. Intervention by sulfur is undeniable, but it is believed to have another explanation.

Sulfur Effect

Research work by K. Grueter and B. Marincek, and more recently work by R. Gautschi[10] have shown that sulfur reduces the surface tension of liquid iron and, in general, the intersurface tensions. However, sulfur is practically insoluble in solid iron, and the sulfur not combined in the sulfides (xMnS•yFeS) is pushed into the liquid phase at the austenite-liquid interface, concentrates at this location (slow diffusion rate) and lowers the local temperature of solidification. At the same time, this concentration of sulfur increases the fluidity and reduces the surface and interface tensions. Accordingly, the nucleation of graphite is facilitated since by equation (1):

$$-\Delta F = -\Delta F_v + \sigma S$$

Therefore, in the presence of sulfur certain foreign nuclei are effective. If the liquid iron is superheated, the sulfur remains but these foreign nuclei are destroyed or poisoned. Graphite does form during the solidification of this superheated liquid iron, but sufficient undercooling is necessary to force either homogeneous nucleation or nucleation from less effective foreign nuclei.

EUTECTIC CELLS

The preceding general considerations are only concerned with the shape and dimensions of the graphite. Another important consideration is the distribution of the graphite, for example the rosette or type *B* classification. Each rosette corresponds to a center

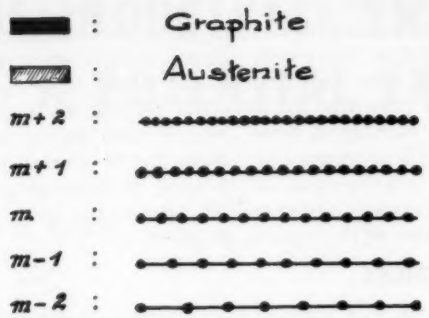


Fig. 7. — Legend of symbols used in Figs. 8a, 8b and 8c.

of crystallization from which a colony develops. These colonies correspond to eutectic cells and they are not numerous. This indicates that the first stage of solidification of the graphite occurred at a small number of effective nuclei.

However, the regularity of dimension and distribution of the colonies indicate that nucleation of the graphite was caused by the supersaturation of the liquid in the vicinity of the solidified austenite, not by general supercooling. In the latter case, the number of nuclei occurring simultaneously is generally great as in the case of supercooled graphite and, to a lesser degree, Type *A* graphite.

Considering a theoretical case of a small volume of perfect radial isotropism from the point of view of temperature, composition, etc., successive stages of radial graphite crystallization starting from a graphite nucleus may be represented by Figs. 8a, 8b and 8c. Figure 7 shows the significance of different items in Fig. 8. The first step considered in Fig. 8a is for the first small graphite particle nearly totally enveloped by austenite.

CARBON DIFFUSION

At this point, the surrounding liquid will be chemically heterogeneous because of the limited diffusion rate of the carbon. The liquid is also physically heterogeneous because of the temperature differences. Considering only the concentration of carbon, it is possible to trace the curves (surfaces) of concentration in the liquid

$m + n \ldots m + 2$, $M + 1$, (positive carbon concentration) where m = average concentration at the limit of the zone of influence,
$m - 1$, $m - 2 \ldots m - n$ (negative carbon concentration, i.e., lower than average). It is evident that the maximum carbon concentration gradient is established around the xx′ axis, and that the maximum concentration is located at the intersection of this axis with the surface of the austenite crystal which tends to envelop the graphite completely.

The next particle of graphite, then, should precipitate at this point of intersection except for the fact that the temperature gradient being at a maximum on this axis is opposed. Thus, it is probable that the next particle of graphite will be nucleated on the xx′ axis at a small distance from the surface of the austenite, and will grow along the xx′ axis of maximum carbon concentration gradient. A similar line of reasoning shows that the next austenite will develop along the yy′ axis.

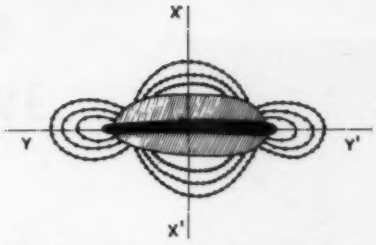


Fig. 8a — The first graphite particle is nearly enveloped by austenite.

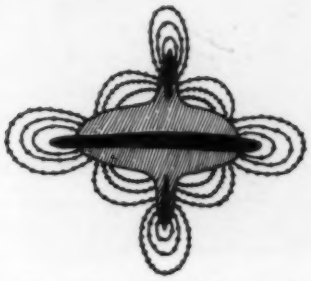


Fig. 8b — The progressive growth of the eutectic cell implies progressive envelopment of new graphite particles by the major austenite phase along the axis xx′.

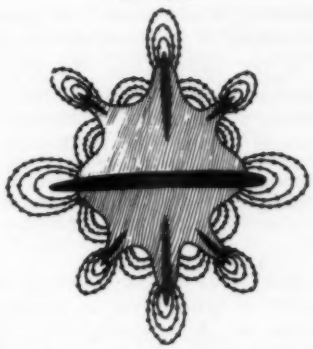


Fig. 8c — Third stage of eutectic solidification development.

As growth of the eutectic proceeds, the new graphite particles will be enveloped by the major austenitic phase growing along the xx′ axis of each graphite flake. This second step is shown in Fig. 8b.

A continuation of this reasoning results in the step shown in Fig. 8c. This is the last drawing that is presented because they become too complicated, and the successive development of the eutectic cell can be easily imagined. The drawings in Figs. 8a, b and c can now be compared to the actual rosette pattern of a eutectic cell shown in Fig. 9. The difference in these structures appears large at first glance, but this can be explained as follows:

1) Perfect radial isotropism was presumed in the drawing for simplicity. In practice, however, this perfect radial isotropism does not exist in either temperature or concentration. This explains why the actual flakes are curved and not perfectly radial in direction.

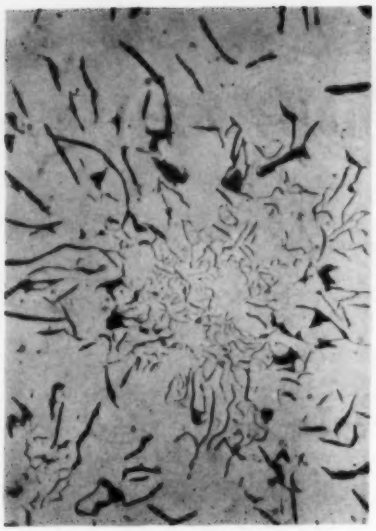


Fig. 9 — Graphite rosette (type B, A.S.T.M.).

2) The rosette shown is not necessarily cut through the center of the eutectic cell.

3) It is noted that the graphite flakes are fine at the center of the cell, increase to a maximum thickness and then decrease in size toward the limits of the rosette. The increase in the dimension is caused by segregation of the sulfur, and the final decrease at the outside of the rosette is the result of interpenetration of the zones of influence of the various rosettes during the final stages of solidification.

4) The photomicrograph of the rosette in Fig. 9 shows a central zone of fine graphite without a distinct radial direction. All of these small flakes formed at the start of eutectic solidification. These flakes take the place of the single large central flake shown in the drawings in Figs. 8a, b and c. The number of these flakes formed during initial solidification may vary from one to several dozen, and is a function of the state of development of the primary austenitic dendrites at the time of nucleation of the eutectic graphite.

REFERENCES

1. (a) A. De Sy, J. Vidts and J. Van Eeghem, "Le mécanisme de la ferritisation de la fonte grise et l'étude de son ralentissement par les éléments Sn et Cu," exchange report of A.T.F.B. at the 21st International Foundry Congress, Florence, Sept. 1954, and *La Fonderie Belge*, No. 12, Dec. 1954.
   (b) A. De Sy and J. Foulon, "La Suppression systematique de la ferrite de formation eutectoid," Report No. 7 at the International Foundry Congress, Florence, Sept. 1954, and *La Fonderie Belge*, No. 12, Dec. 1954.
2. (a) J. A. Davis, D. E. Krause and H. W. Lownie, Jr., "Tin as an Alloy in Gray Cast Iron," AFS TRANSACTIONS, *vol. 65*, p. 590, 1957.
   (b) *Steel*, p. 83, May 6, 1957.
   (c) *Bulletin hebdomadaire d'information et de liaison de FABRIMETAL, no. 583*, p. 683, Sept. 2, 1957.
3. M. Volmer, *Kinetik der Phasenbildung*, Dresden, 1939.
4. M. Straumanis and N. Bracks, *Z. phys. Chem., vol. 30*, 115, 1936, and *vol. 36*, 140, 1938.
5. N. T. Belaiew, *Journal of the Iron and Steel Institute*, 105, 1922.
6. E. Scheil, *Metallforschung, vol. 1*, parts 1 and 2, 1946.
7. A. Boyles, *The Structure of Cast Iron*, A.S.M., Cleveland, 1947.
8. W. Patterson, *Giesserei*, Techn. Wiss. Beihefte, part 6, 1952.
9. K. Grüter and B. Marincek, *Giesserei*, Techn. Wiss. Beihefte, part 12, 1953.
10. R. Gautschi, Doctorate thesis, Zürich, published by P. G. Keller, 1957.

THE INDUSTRIAL ENGINEER AS A POTENTIAL EXECUTIVE

By M. E. Mundel

ABSTRACT

Industrial engineering, broadly conceived, is a built-in management development program in a company. This concept of industrial engineering allows the smaller company to attract men of real ability, will help reduce the management-staff rift, and provides the necessary background for potential managers.

INTRODUCTION

Most recently one of the consulting assignments of the author's company has been with a small jobbing machine shop which specializes in a particular type of work. It is hoped that assistance can be given to make this organization profitable again. A basic lesson is contained in an examination of how it reached its present plight of operating at a loss.

The shop operated for many years doing its particular type of work. There was an abundance of this type of work available and a shortage of such jobbing shops. They did well. As the volume of work grew, some of the plants which were sources of such work grew to a size where they found it profitable to purchase their own specialized machining equipment, and their work no longer went to the jobbing industry. There was recently, from the above cause, about a 10 per cent reduction work available for the industry.

As you might anticipate, the ten per cent reduction did not hit the various jobbers across the board; it hit the high-cost shops and this shop was one of them. Their work almost disappeared over-night. A long hard struggle lies ahead as the type of management is developed that will put this shop back in the black. It is my contention that the basic lesson to be learned is the danger of failing to continually develop a cost-conscious management.

In a way, it is the advent of occurrences of this type, as well as the wishes to avoid them, that has given rise to the enormous growth of executive development plans within the larger companies, and to the growth of executive development courses at various University extension divisions to serve the companies who are not large enough to organize their own plans.

I do not blame the plight of the mentioned company on conditions within the industry; I say that this organization failed to develop executive leadership and use this leadership to keep them from getting into this "bind." I should add, in all fairness, that present management is aggressively attacking the problem and is also developing a management team to safeguard the future.

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EXECUTIVE DEVELOPMENT

But how do you develop executive or managerial personnel? Most business has only recently become concerned with this problem.

> Throughout the latter half of the nineteenth century, which saw the blossoming of industrial capitalism, enterprises were bound to individuals. Although the modern corporation was firmly established in law and in certain sectors of the economy long before the end of the century, the smaller individual and family companies, rather than the large, impersonal corporation, typified business enterprise — even during the second half of the nineteenth century. Under these circumstances there was no special problem connected with the selection and development of executives. The man who made his little business grow into a large institution remained at the head of the organization when it expanded. Eventually, the responsibility for ownership and management was transferred to his son or son-in-law.[1]

The changing nature of our society, and the changing manner of performing our industrial work, have brought the problem of the executive to the foreground. Executive responsibility may no longer be simply handed to an untrained individual like a mere piece of property. The executive who is to receive the responsibility must be trained to receive it or he will more likely be a receiver.

MANAGEMENT RESPONSIBILITY

Management is, after all, the task of designing, predicting the performance of, controlling and providing the facilities for an integrated human-group activity, the related physical facilities and the inter-relationships between the two.

The continually increasing mechanization, the growth of unions and the various additional social restrictions on activities have made continually the task of management more complex. Competition, as it increases, adds to the problem.

> If, as many managements believe, competition is going to be more rigorous in the years ahead, then the supply of future executives becomes more critical. The competitive edge will be in the caliber of a company's executives. But executives for ten years or more in the future must be recruited today.
>
> Most companies recognize this need, and so we have been witnessing a massive talent hunt.[2]

The basic contribution of the manager is the making of the decisions necessary to accomplish assigned or assumed objectives within the task of management.

MANAGERIAL LEVELS

Within any organization, managers operate in three somewhat indistinctly separated "decision areas." These are usually associated with three different "levels" of management, which also have somewhat hazy borders, although in a small organization an individual may "wear several hats." The three decision areas and managerial levels are:

a) Top managers. Those who make decisions concerning the relationship of the organization to external economic, situational and political factors. These decisions lead to broad statements of policy or objective.

b) Middle managers. Those who make decisions concerning how, within the policy framework, to achieve the broad objectives. These decisions result in plans for action and this group is often the one most commonly identified as "operating managers."

c) Supervisory managers. Those who make decisions concerning the day-to-day carrying out of action plans.

In the usual organization, each level of management may have staffs (individuals acting data gathering or data digesting extensions of the managers), particularly when either the complexity of the data on which the decisions are based (decision difficulty), or the number of decisions per unit of time (decision density), exceeds the capacity of the individual with the decision making responsibility.

Staff Duties

These staffs, in general, attempt to alleviate the effect of high decision difficulty, high decision density or a combination of both, by activities such as:

1) Dimensioning management designs by means of work count, work measurement, quality specifications, etc.
2) Routinizing and channeling information to reduce legwork and to compile the necessary managerial data into comprehensible form.
3) Making management designs (such as job methods, procedures, production sequences, etc.) to eliminate minor decisions from clogging organization channels.
4) Monitoring and reporting on conformance of management designs and organization performance, both to assist supervisory managers and to evaluate action plans, as well as to provide a comprehensible experience basis for future policies and plans.
5) Recommend many decisions to top and operating managers based on their (the staffs) concept of the objectives.

And if we consider the foregoing five levels of managerial staff as a hierarchy of management, then we could appropriately add:

6) The manager who makes the decisions.

Industrial Engineering

The first five levels in the hierarchy are frequently referred to as "industrial engineering" functions, although there may be other staffs who also perform some of the same functions in different activity areas. However, these other staffs lack the closeness to production, and production is where the basis of costs really lies.

Those of you who come from small organizations may be saying, "This does not apply to me; I cannot afford so many people." Only the last half of this statement is true. You cannot afford the people, but you have the functions and they must be performed.

Before we continue with this, I made an implication concerning industrial engineering which I think is worth examining at greater length.

In all too many cases, industrial engineering is considered as another name for time study and relegated to the role of a somewhat difficult, clerical function. I firmly believe that those who take this position are denying themselves not only tremendous services but also cutting off one of the vital sources of potential executives.

It is interesting to see what Don Burnham, Vice-Pres. of manufacturing of the Westinghouse Corp., has to say about this:

> You may wonder at this point what I consider industrial engineering to include, and this is a logical question. There is certainly little agreement throughout industry regarding the exact boundary lines of the functional department called Industrial Engineering. This unusual situation itself is primarily attributable to the rapid growth of the function. As certain sub-functions have developed, they have become sufficiently important to be segregated as self-sustaining individual departments.
>
> Compounding the confusion which sometimes exists is the fact that different sub-functions reach this level of importance in different industry or product situations. It is not uncommon to find a plant with separate staff departments for Quality Control, Production Control, Inventory Control, Work Measurement, Methods Engineering, Plant Layout, and others. Such splits, however, do not indicate a weakness in the industrial engineering function. To the contrary, they serve as evidence of the degree to which the function has become indispensable to the manufacture of any products.
>
> You can see, then, that I am considering industrial engineering in the broad sense. That is, it includes all applications of scientific principles to the planning, operating, and control of the complete manufacturing operation.[3]

MANAGEMENT TRAINING

The people who perform these industrial engineering functions are those that really serve as staff to the managerial group, be the group large or small. But not only do they serve the executives, but more important, they "train" so to speak, under them. As the staff grows in ability they come closer and closer to the actual executive function; *viz*, as they move up through the hierarchy previously given. In some cases, the upward movement is carefully planned. We find more and more notes appearing, as follows:

> Are your foremen a little weak on time standards? It's no problem at Worthington Company's Compressor and Engine Division (Buffalo, N. Y.). There every foreman comes up from the ranks. But no

> foreman goes directly from the ranks to the foreman job. He first spends two years as a standards man.[4]

It would appear as though industrial engineering, broadly conceived, is the ideal, built in, management development program that most companies already possess. Further, such a concept of industrial engineering, by providing a greater goal for the engineers, allows even the small organization to attract men of real capability. Further, the use of this staff as the future management potential line, will assist in reducing the rift that sometimes exists between line and staff, thus making both more effective.

DISADVANTAGE

The only drawback to thinking of industrial engineers as future executives seems to be the thought that good engineers, once you get them, will be drawn eventually away from the industrial engineering work and thus, be lost to the function. I think this is somewhat academic, because if the job is a dead end, you will probably never get them in the first place. But more seriously, if the industrial engineer eventually becomes a manager, certainly he is in a position to make use of an aggressive staff. And a continually aggressive management is required if a organization is to stay alive.

CONCLUSION

If we consider the manner in which an effective management operates, then it seems only natural that the present industrial engineers are the potential executives, provided the industrial engineering function is broadly conceived, soundly staffed and properly used. Then, not only will it assist the organization now, but it will also be an executive development program for the future.

REFERENCES

1. Ginzberg, E., "What Makes an Executive," Columbia University Press, New York, p. 4, 1955.
2. *Acme Reporter, No. 3*, Association of Consulting Management Engineers, Inc., N. Y., 1958 Series.
3. Burnham, D. C., "The Industrial Engineer's Job Today," *The Cogwheel, vol. 7*, No. 2, Fall 1958.
4. *Factory Management and Maintenance*, p. 162, March 1959.

ALUMINUM ALLOYS FLUIDITY TEST

fluidity tripled with new mold coating

By M. C. Flemings, H. F. Conrad and H. F. Taylor

ABSTRACT

Still another fluidity test is described herein, to be added to the already large number of such tests developed in past years. This is a relatively simple sand mold type test which is especially well adapted (because of the accuracy and reproducibility of test results) to research and development work, but is equally suitable as a foundry control tool. The test piece combines the desirable simple features of other types and includes a metal screen in the runner (in conjunction with a carefully designed sprue and pouring basin) to regulate entrance velocity and provide clean dross-free metal in the spiral.

The above test piece was used to determine:

1) The effect of pouring temperature and alloy composition on the fluidity of aluminum-copper alloys.
2) The relative fluidities of several commercial aluminum alloys.
3) The effect of mold coatings on the fluidity of aluminum alloys.

Of all mold coatings tested hexachloroethane proved to be the most spectacular in its effect; for example, using a coating 0.006 in. thick, the fluidity of aluminum-4.5 per cent copper (at 1330-1305 F) was improved 200 per cent (e.g. tripled). Among the other coatings tested were carbon-black, graphite, gypsum and zinc carbonate.

It has been concluded that one important function of certain of these coatings is to decrease heat flow (act as insulators). However, other factors may also be important such as reduction of surface tension of the metal and alteration of the character and/or amount of the oxide film which, in aluminum alloys, tends to form at the tip of the flowing stream.

Hexachloroethane improved the fluidity of all alloys tested, including high purity aluminum, 195 alloy (high purity), 356 alloy (high purity), 220 alloy and an aluminum-zinc commercial alloy (40E). Studies using a large, thin, plate casting gave ample, practical proof that the beneficial effect of the coating extends to commercial type castings and is not a laboratory curiosity limited to test spirals.

INTRODUCTION

To meet the design requirements of modern aircraft and missiles, strong, sound aluminum castings must be produced in thinner sections than have heretofore been considered possible. Fabrications are sometimes chosen in preference to castings (even at much greater cost and trouble) simply because castings cannot be made thin, and hence, light enough to meet requirements.

The difficulty in producing thin castings (under 1/8-in.) is due primarily to inadequate metal "fluidity"; the metal solidifying before completely filling thin mold cavities. Metal fluidity is increased in practice by judicious choice of alloy composition, and by providing additional superheat to the melt. However, there are limitations to these techniques, and few foundries can produce aluminum castings with extensive sections much less than 1/8-in. thick.

In July, 1957, a research program was sponsored at Massachusetts Institute of Technology by the U.S. Army Ordnance Dept. through Pitman-Dunn Laboratory, Frankford Arsenal, to obtain a basic understanding of the factors affecting fluidity in aluminum alloys and, if possible, to develop practical methods for increasing fluidity. This paper describes one phase of the initial research; dealing with 1) development of an accurate and reproducible sand mold fluidity test for aluminum alloys, 2) a measurement of the relative fluidities of several typical aluminum alloys using the test method developed, and 3) a study of mold coatings effects on fluidity in sand molds.

A parallel program, also under Frankford Arsenal sponsorship, was conducted to determine how metallurgical factors ("mode of solidification") of aluminum and its alloys affect fluidity. This study has been carried out with the "vacuum fluidity tester" developed by Ragone, Adams and Taylor[1,2] and used at M.I.T. and elsewhere in several earlier researches.[3,4] Results of this aspect of the present fluidity research are published elsewhere[5,6] and will not be presented herein.

FLUIDITY TESTING

Fluidity, in the casting sense, refers to the property of a metal which allows it to flow into a mold. It

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is usually defined as the distance metal flows when poured into a standard fluidity test channel. This channel may be straight or it may be in the form of a spiral, and the cross-section of the channel may be round, half-round, trapezoidal or rectangular.

T. D. West,[7] in 1902, was the earliest investigator to report on the flow characteristics of molten metals cast into sand molds. He poured metal into a horizontal wedge and considered the distance flowed as a measure of fluidity. Ledebure,[8] in 1904, Sexton and Primrose,[8] in 1911 and Moldenke,[9] in 1917, modified the wedge test somewhat.

Ruff[10] ran metal in a long cylindrical channel, and used the length of flow as his value of fluidity. This test was particularly sensitive to errors in leveling. Evans[11] tried an inverted "U" type of test in which he used several vertical sections of various cross-sectional areas fed from a common channel. The heights to which the metal rose in the various sections were a gage of fluidity.

Fluidity Spiral

The familiar fluidity spiral was first tried by Saito and Hayashi[12] in 1919; their method simplified handling and leveling problems. Many investigators have improved upon this spiral type test; the two best known modifications being those of Saeger and Krynitsky[13] for cast iron, and of Taylor, Rominski and Briggs[14] for steel. These spirals are now accepted in America as standards for determining fluidity of ferrous metals.[13,17]

Eastwood and Kempf,[18] and later Sicha and Boehm[19] developed a spiral casting of flat cross-section for studying fluidity of aluminum and its alloys. Only a limited amount of experimental work has been done with this fluidity test, but results to date indicate it is not sufficiently reproducible for research and development work.[16] A "star" type fluidity test piece has been developed by Kondic.[20] It, too, is a relatively recent development which must yet stand the test of time.

A new approach in fluidity testing was developed by Ragone, Adams and Taylor.[1,2] They developed a "vacuum fluidity test" in which the liquid metal was drawn into a pyrex glass tube by means of a vacuum. Variables of this test were determined experimentally and analytically using low melting point alloys. Later investigators have employed this method in precise determinations of the fluidity of magnesium and its alloys,[3] and of aluminum alloys.[4,6]

Several excellent, comprehensive bibliographies of the history of fluidity testing have been prepared; the reader is referred to these for more complete coverage of the subject.[15,16]

FLUIDITY OF ALUMINUM AND ITS ALLOYS

Until recently, interest in metal fluidity has been concentrated largely in the field of ferrous metals. Fluidity tests have been developed and are used commercially as quality control checks to determine the flowing qualities of molten metal. Fluidity tests are used in some ferrous shops in place of temperature measurement.

Fluidity testing of aluminum alloys has been less important as a shop control technique because temperature measurement has appeared to be an adequate gage of the running qualities of a given alloy. Also, such temperature measurement is accurate and reliable, and until recently fluidity tests for aluminum alloys were either quite complicated or yielded results which were difficult to reproduce.[16]

Courty,[21] in 1931, measured the fluidity of various aluminum alloys in a carefully designed spiral test mold using a bottom pouring basin stoppered with two fusible lead plugs. While the arrangement was well designed for the use intended, it does not appear adaptable as a standard fluidity test because of its relative complexity.

In 1939, Eastwood and Kempf[18] developed the fluidity spiral of flat cross-section described above. Cross-section of the channel was 1¾-in. wide by 1/16-in. thick. A simple ¾-in. diameter downsprue was used with an overflow pouring basin. Sicha and Boehm[19] later modified the design by adding a "well" at the bottom of the sprue and adding a piece of tinfoil to serve as a temporary "dam," thereby reducing the influence of any variation in pouring. While these modifications were said to result in improved uniformity of experimental runs, nonetheless, fluidity spiral weights for duplicate runs varied by as much as 15 grams, in an average total spiral weight of only 60 grams.

Investigation Results

Experimental results of the above investigations included 1) a comparison of the relative fluidities of several commercial foundry alloys, and 2) data illustrating that dissolved gas content and remelting had little or no effect on fluidity of aluminum alloys.

Kondic and Kozlowski[22] later developed an apparatus for bottom pouring aluminum spirals in a metal fluidity mold. The metal was melted in situ, and at the desired moment a stopper rod was lifted and the aluminum allowed to flow into the mold cavity. The mold was heated with electrical elements to any testing temperature desired. Fluidity data were obtained for aluminum-silicon alloys from 0 to 16 per cent silicon, illustrating that optimum fluidity is obtained with pure metals and with eutectics. The authors also showed that modified aluminum-silicon alloys possess considerably lower fluidity than unmodified alloys. This latter conclusion has recently been confirmed by Kondic using his new "star" fluidity test pattern.[20]

Vacuum Fluidity Test

Utilizing the "vacuum fluidity test," Ragone and Floreen[4] investigated the fluidity of a series of aluminum-copper alloys, and aluminum-magnesium alloys. Both alloys were investigated in the region of 0 to 33 per cent solute. Results obtained for the aluminum-copper system were qualitatively similar to those obtained by Courty[21] with his quite different fluidity test. In the work of Ragone and Floreen, fluidity was found to vary inversely with solidification range; solidification range was calculated from nonequilibrium considerations in the manner used by Niesse, Flemings and Taylor[3] in their study of fluidity in magnesium alloys.



Fig. 1 — Single spiral fluidity casting with gating system.

DEVELOPMENT OF SAND MOLD FLUIDITY TEST

The "vacuum fluidity test" wherein metal is drawn out of a crucible directly into a glass (or metal) fluidity channel has been found to be ideal for studies of the effect of solidification variables on fluidity. In addition, when glass tubes are used, motion picture photography can be utilized to follow the progress of metal during the test. The major disadvantage of the test, however, is that it does not readily permit study of the effect of mold or mold-metal variables on fluidity. For example, study of the effects of sand grain size, moisture content or mold coatings on fluidity requires the use of a more standard sand mold test. In addition to being able to measure effects of mold and mold-metal variable, a reproducibile sand mold fluidity test would be e pected to closely relate to actual conditions of castin

After carefully evaluating the various test piece developed to date it was decided no single one wa simple enough for extensive testing and also for gi ing test results accurate enough for laboratory inve tigation. Accordingly, it was decided to develop new type test which would be simple and easy to use, and would also:

1) Afford precise control over metallostatic "head," and permit this pressure head to be reached before any metal entered the fluidity spiral.
2) Provide control over metal turbulence as the metal entered the spiral.
3) Filter any dross or other foreign materials from the metal before the metal entered the fluidity spiral and do so without altering the metal "head."
4) Be of flat cross-section to simulate problems encountered in pouring sand castings of thin section in the foundry.

Single Spiral Mold Design

The fluidity test initially developed to meet the above requirements is illustrated in Figs. 1 to 5. Figure 1 shows a cast spiral with pouring basin, sprue, runner and gate attached. The pouring basin is designed with an overflow to maintain constant metal head during pouring, and the sprue is tapered to minimize tubulence and aspiration of air. After passing through the sprue, metal enters a large runner and passes up through a screen to remove dross. It then travels through a vertical gate and into the spiral. The large runner serves to minimize turbulence and to remove dross (by permitting a screen to be molded in); its primary purpose, however, is to permit a time delay so that constant "head" is obtained before any metal enters the fluidity spiral. A somewhat similar arrangement has been described by Prussin and Fitterer.[23]

Figures 2 to 5 provide details of the test equipment. The runner (with screen) is made as a core but the spiral itself is formed entirely in the sand mold (Fig. 2), which for this investigation was green sand. The

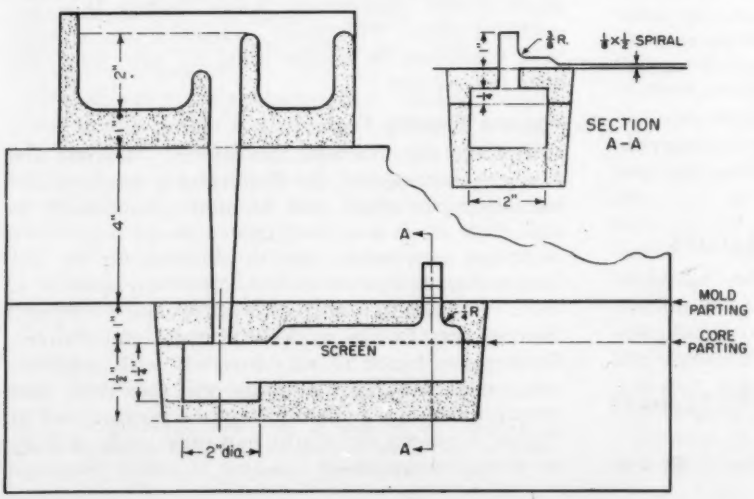


Fig. 2 — Schematic diagram of single fluidity test mold.

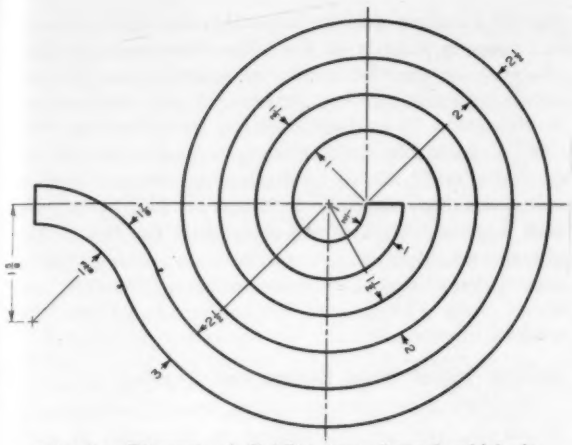


Fig. 3 — Diagram of fluidity test channel which is ⅛-in. thick.

spiral (Fig. 3) is of rectangular cross-section and is 28 in. long x 0.50 in. wide and 0.125 in. thick. Figures 4 and 5 are photographs of the matchplate pattern, dry sand pouring basin and dry sand cores and screen used in the gating system. The screen is of tin plated steel with holes 1/16-in. diameter. Perforations account for approximately 50 per cent of the area of the screen.

Sprue Design

The sprue employed tapers from ¾-in. square at the top to ½-in. square at the base. The area of the sprue base is small enough to permit the pouring basin to be filled to its maximum level before any metal enters the spiral itself. This fact was verified by placing a thermocouple at the entrance to the spiral and observing the height of metal in the pouring basin at the moment flowing metal first contacted the thermocouple. Approximately 2½ sec elapsed from start of pour until the metal reached the spiral entrance.

The test piece described above was found to satisfy all the essential preliminary requirements; the molten metal reached the spiral free from dross, flowing smoothly at a constant rate, and under a constant pressure head. Radiographic examination showed the spiral castings to be entirely dross free.

Single Spiral Mold Accuracy

Accuracy and reproducibility of the test piece were checked by pouring 1) 18 test castings of aluminum-4.5 per cent copper alloy, varying only the pouring temperature, and 2) 20 test castings in which copper content as well as pouring temperature was varied.

In this phase, and throughout the entire study, molds were hand rammed in a 14 in. x 18 in. aluminum snap flask. Molding material used was synthetically bonded green sand (140 N. J. Silica Sand). A ½-in. diameter vent hole was provided at the end of each spiral to allow mold gases to escape. Baked core sand bonded with urea formaldehyde was used to form the pouring basin and runner cores.

Melting was done in a gas-fired crucible furnace. All alloy compositions were prepared using aluminum ingot (99.9 per cent) and 50 per cent aluminum-50 per cent copper master alloy. Master alloy was added shortly after meltdown of the aluminum ingot. No grain refiners were used. Each melt was degassed by bubbling dry nitrogen through the metal for at least ten min at 1300-1350 F (while metal continued to heat in the furnace). Melts were then tapped at 1425 F and checked for gas with the reduced pressure tester.

If any gas was found to be present, the degassing operation was repeated. Samples were taken from each heat for chemical analysis. After tapping, all molds were poured consecutively from the same crucible, with close temperature measurement to obtain the proper pouring temperature.

Temperature Measurement

Since pouring temperature is of such great importance in determining metal fluidity, it is necessary to measure temperatures precisely. In these experiments, a chromel-alumel thermocouple, enclosed in an iron protection tube, was immersed in the crucible at the time of pouring. Melt contamination was prevented by coating the iron tube with a layer of silica-bentonite wash. The temperature of the melt at the instant of pouring was indicated on an electronic recorder with an external cold junction (ice water).

Figure 6 presents the fluidity data obtained for the aluminum-4.5 per cent copper heats, with pouring temperature as the only variable. Fluidity increases



Fig. 4 — Cope half of single spiral pattern plate.

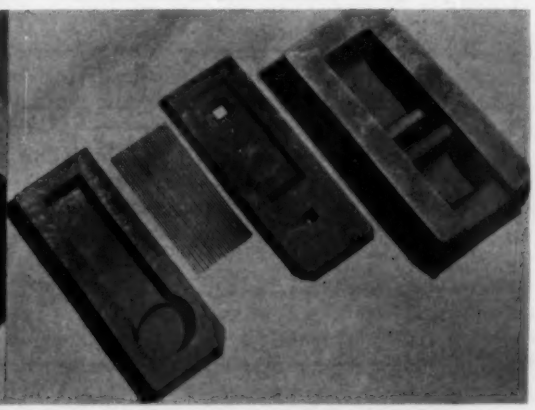


Fig. 5 — Left to right — Bottom runner core, metal screen, top runner core and pouring basin.

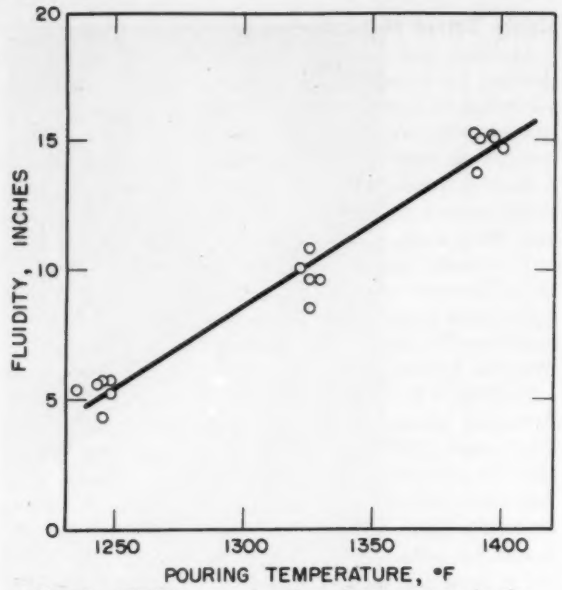


Fig. 6 — Fluidity as a function of temperature in the single spiral fluidity test (showing reproducibility of test) for Al-4.5 per cent Cu alloy. Data from three separate heats.

from about 6 in. for castings poured at 1250 F to about 15 in. for those poured at 1400 F.

Fluidity of aluminum-copper alloys containing from 0 to 20 per cent copper is shown in Fig. 7. Length of flow depends markedly on copper content as well as on superheat. Up to about 1 per cent copper, fluidity decreases with increasing alloy additions; further alloying increases fluidity. For alloys poured at constant superheat, increasing copper contents (above 1 per cent) increase fluidity only slightly. However, for alloys poured at constant temperature, the improvement in fluidity with increasing alloy addition is much more pronounced.

Figure 7 clearly illustrates that both the composition of an alloy and its superheat are important in determining fluidity of the alloy. Composition (alloy analysis) affects the mode of solidification in the flowing stream, and superheat affects the rate of solidification at critical locations in the stream.[24]

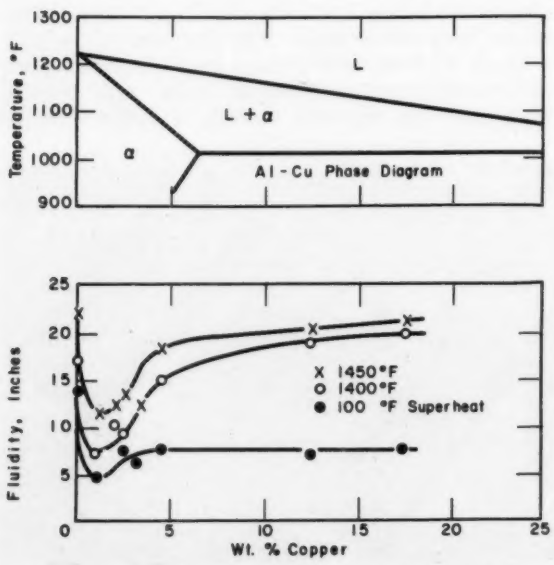


Fig. 7 — Fluidity vs. composition for the aluminum copper system at various pouring temperatures.

The results obtained in the green sand mold fluidity test (Fig. 7) are in qualitative agreement with results previously obtained by Courty[21] and by Floreen and Ragone,[4] both of whom studied fluidity in the aluminum-copper binary system. As mentioned above, Courty employed a bottom pouring crucible and metal mold; Floreen and Ragone employed the "vacuum fluidity test."

Double Spiral Mold Design and Testing

During the later stages of this research program, as increased emphasis was placed on the study of mold coatings, it became evident that a test casting with two identical spirals attached to a single runner would be desirable. This arrangement would permit comparison of mold variables in a single casting. For example, in checking the effect of a particular mold coating, one spiral only would be coated and the other would serve as a reference standard. The double spiral test is shown in Fig. 8.

Essentially, the only differences between the single and double spiral designs are that in the latter two spirals are attached to the runner, and the spirals are in the drag rather than the cope. The second modification was to permit observation of metal flow in the spirals. With the spirals in the drag it is possible to embed glass "viewing plates" in the cope and follow metal flow visually or by motion picture photography. Work on this aspect is currently underway, and has provided interesting and valuable information which will be reported at a later date.

Reproducibility of this spiral was checked, as was that of the single spiral, by pouring 30 test castings of aluminum-4.5 per cent copper alloy at three different precisely determined temperatures. The results are shown in Fig. 9. A straight line has been drawn through the points for reference; the slope is the same as that of the line drawn for the single spirals, Fig. 6. Reproducibility in this test, as in the single spiral is to within about ±1 in. of flow at a given temperature.

Absolute fluidity values obtained in the double spiral were slightly lower at all temperatures than

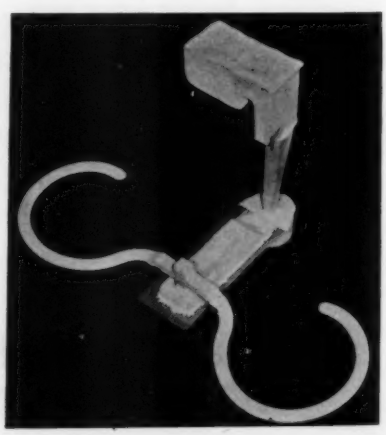


Fig. 8 — Double spiral fluidity casting with gating system.

those of the single spiral (about 1.4 in. less). Correspondence between the length of the right hand and left hand spiral in a single casting was found to be within 1 in., except for one of the 30 castings where the difference was 1.5 in.

Thicknesses of 27 spirals were measured at intervals along their length. Average thickness of all spirals was 0.123 in. with the average thickness of any one spiral deviating from this by not more than about ±0.005 in. Thus, there was no difficulty encountered in obtaining an accurate and reproducible flow-channel.

Consistent variations in thickness were, however, found in the individual spirals. All spirals measured were slightly undersize (0.001-0.003 in.) in their central locations and oversize near their tips (0.002-0.004 in.). At the tip, spirals were usually undersize due to lack of complete filling of the mold cavity. The consistent variations in thickness could not be correlated with variations in the pattern equipment, and it is felt that the variations are a result of the solidification mechanism of the alloy in the spiral.

Melting and molding practice were the same in this series of heats as in the series poured to check reproducibility in the single spiral, with the exception that aluminum-4.5 per cent copper alloy revert from previous heats was employed. The revert (remelted and pigged alloy) was used in amounts up to 80 per cent of the charge. All revert was analyzed before use to assure close control of chemical composition; runners containing the screens were discarded and not remelted.

MOLD TREATMENT EFFECTS ON FLUIDITY OF AN ALUMINUM ALLOY

Work to date on this aspect of the program has consisted of 1) selecting mold coatings or other mold treatments which might be expected to improve fluidity, 2) testing these mold treatments, and 3) interpreting observed results on the basis of thermal, chemical or physical factors. This study is not yet complete, and data contained herein should be considered a progress report of continuing research.

Melting and molding techniques employed in this phase of the study were identical to those described earlier. The double spiral was used for all quantitative work reported herein; in these cases one spiral in each mold was treated and the other left untreated to act as a control. Some of the qualitative observations reported were obtained from research using the single spiral mold.

Treating agents were sprayed or dusted on the mold, depending on the particular agent being applied. Thin pieces of sheet metal were placed at several locations near the mold cavity to be sprayed. An even coating was then applied to the mold face, and the coating thickness (on the sheet metal) measured with a micrometer. Most of the fluidity data reported are from tests employing mold coatings of 0.003 ± 0.0005 in. thick. A study is also described of the effect of varying coating thickness (of one treating agent).

Amorphous Carbon and Graphite Coatings

Many foundrymen have long felt that "blacking" a mold surface (using an oxyacetylene torch) has a beneficial effect on fluidity. However, no quantitative laboratory evaluation of this has been made. In this study, the effects on fluidity of three forms of carbon as mold coatings were examined. Carbon was applied as lamp black (from an oxyacetylene torch), as activated charcoal and as graphite.

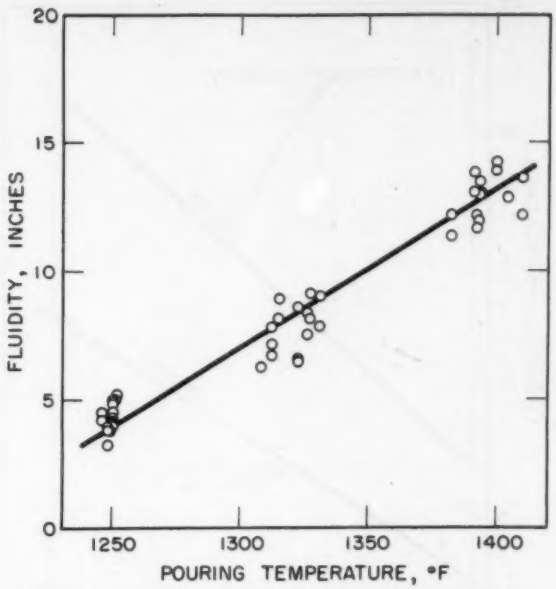


Fig. 9 — Fluidity as a function of temperature in the double spiral fluidity test (showing reproducibility of test) for Al-4.5 per cent Cu alloy. Data from seven separate heats.

Activated charcoal, in a finely divided form, was applied to the sand mold surface in controlled thicknesses by suspending the carbon particles in ether and spraying the mold. The ether volatilized quickly, leaving a uniform layer of activated charcoal particles built up to a thickness of 0.003 in. Molds coated in this manner evidenced an improvement in fluidity of about 40 per cent (i.e., the treated spiral was 1.40 times the length of the untreated control spiral).

Of greater interest are the results found by applying an amorphous carbon layer to the mold face by torching. The improvement in fluidity resulting from a 0.002-0.003 in. layer of lamp black (carbon black) applied in this manner is shown in Fig. 10. Improvement is 130 per cent at 1250 F and 68 per cent at 1400 F.

Some difficulty was experienced in attempting to obtain a uniform coating of the blacking (at 0.003 in.), due to its tendency to pile up and give unevenness; also, at thicknesses over about 0.003 in. the blacking did not adhere well to the mold surface. It would appear that if carbon black is to be used in thicknesses over about 0.002-0.003 in. it should be applied by techniques other than the use of an oxyacetylene torch.

Improvements in fluidity possible with this technique were found to be much greater than those possible with a coating of activated charcoal. Both the carbon black and the activated charcoal are amorphous forms of carbon; however, due to the extremely fine, feathery nature of the lamp black coating, its thermal insulating properties are much

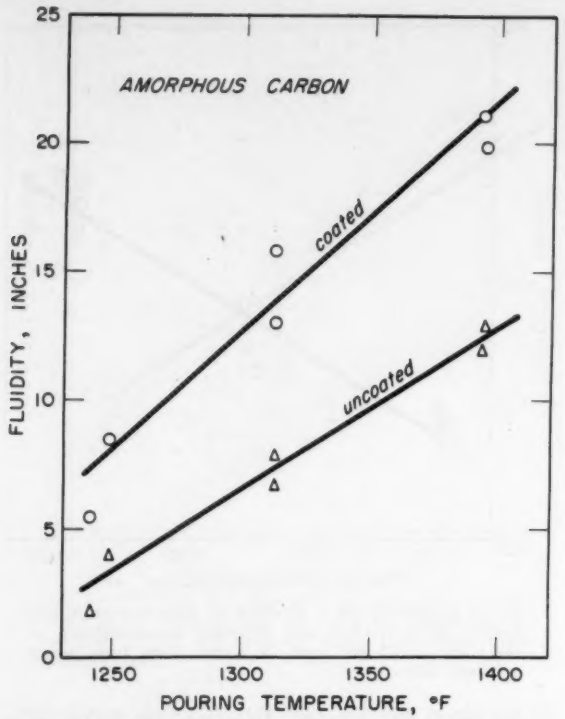


Fig. 10 — Amorphous carbon mold treatment effect on fluidity in the double spiral test for Al-4.5 per cent Cu. Mold treatment — 0.002 in. to 0.003 in. thick coating of amorphous carbon (flame application). Mold treatment was applied to one spiral in each mold; the other spiral was uncoated and was the control spiral.

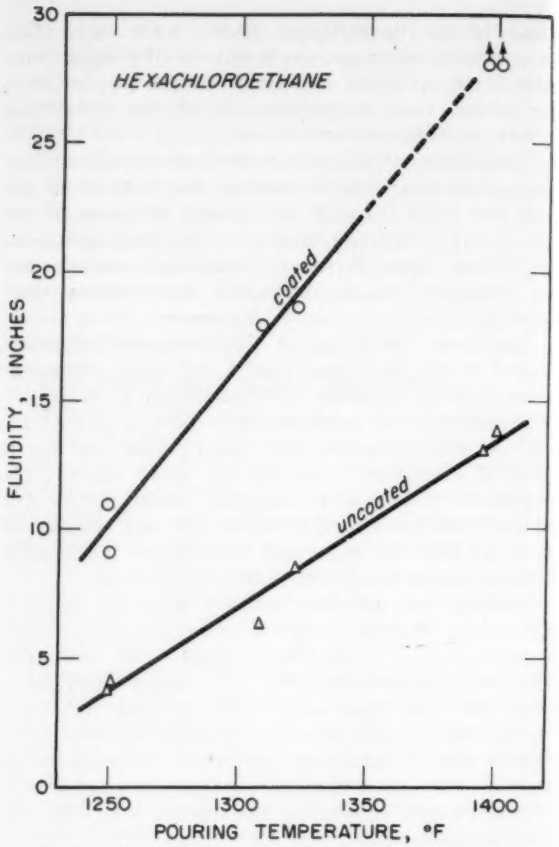


Fig. 11 — Hexachloroethane (C_2Cl_6) mold treatment on fluidity in the double spiral test for Al-4.5 per cent Cu. Mold treatment — 0.003 in. thick coating of hexachloroethane (spray application). Mold treatment applied to one spiral in each mold; other spiral uncoated for control.

greater. The thermal conductivity of lamp black is given as being more than 100 times lower than that of carbon.[25]

To complete this study, a test was made wherein spiral molds were brushed with a layer of graphite in much the same way as is done with sand molds to improve the surface finish of iron castings. No improvement in fluidity was noted in this experiment, probably because of the much higher thermal conductivity of graphite as compared to lamp black.

From these experiments it appears that the insulating qualities of the carbon coating are of primary importance in increasing fluidity. The greater the insulating qualities of a carbon coating, the greater is the fluidity increase obtained from the coating. As a check on this theory, several tests were made dusting the mold surface with a layer of gypsum ($CaSO_4$). Gypsum can be obtained in fine powder form and is highly insulating. The improvement in fluidity resulting from this treatment was substantial, and about the same as that obtained with activated charcoal.

Hexachloroethane Coating

A somewhat different approach to the problem of promoting fluidity in sand molds has been based on the use of an organic compound with a low sublimation temperature. Hexachloroethane (C_2Cl_6) has been successfully used as a spray for a mold coating. The results of the double spiral test using a 0.003 in. coating of hexachloroethane on the treated side of the mold are shown in Fig. 11. The points shown by arrows at the highest temperatures indicate that the entire length of the test spiral was completely filled (28 in.); to adequately examine the improvement in fluidity of this compound at the higher temperatures a more difficult test would be required (e.g., a spiral of a thinner cross-section or greater length).

Improvement effected by the hexachloroethane was 180 per cent at 1250 F and greater than 110 per cent at 1400 F. A series of three spiral test castings from this series is illustrated in Fig. 12.

The improvement in fluidity resulting from the hexachloroethane coating is a strong function of coating thickness, as shown in Fig. 13. A 0.003 in. thick coating improves fluidity roughly 100 per cent, while a coating 0.006 in. thick results in a 200 per cent improvement (at the pouring temperature studied). A 200 per cent improvement is equivalent to tripling fluidity, and even greater improvements with thicker coatings seem feasible. The straight line giving results for the untreated molds simply shows the consistent length of the companion control pieces on the double spiral; the coating thickness on these is constant at zero.

At a coating thickness greater than about 0.003 in. (of hexachloroethane), depressions begin to appear on the fluidity spiral castings. These depressions are not present in the absence of the hexachloroethane, and appear to be simple gas "blows." In subsequent

work using hexachloroethane to improve fluidity in plate-type castings, the gas depressions did not appear.

Mold coatings of hexachloroethane were obtained by dissolving the compound in ether to its maximum solubility and spraying the solution on the mold. The ether evaporated leaving a white coating of the hexachloroethane; no difficulty was experienced obtaining a uniform and adherent coating.

It is of interest to note here that, while the thickness of uncoated spirals averaged 0.123 in., and the thickness of the spirals coated with carbon-black averaged slightly less, the thickness of the spirals coated with hexachloroethane averaged 0.014 in. thicker (0.137 in.). Several factors may account for this. The hexachloroethane coating should be expected to separate the cope and drag at the parting line by the coating thickness (0.003 in. on each half for most runs).

Then, since the coating in the cavity volatilizes, the casting would be expected to be at least 0.006 in. thicker. Another reason may also be that the hexachloroethane permits the metal to fill the cavity more completely, through some change in the surface tension properties of the metal and/or its surroundings.

Mechanism of Improvement Due to Mold Coatings

The improvements in fluidity resulting from amorphous carbon and gypsum coatings have been attributed to the insulating qualities of these compounds. These coatings do not seem likely to affect the surface characteristics of the flowing metal or to enter into any other important metal-mold reaction. Complete proof that the insulating quality of these coatings is the primary factor has not yet been obtained; nonetheless, substantiating evidence is found in the fact that crystalline graphite has little or no effect on fluidity. Crystalline graphite is chemically identical to amorphous carbon, but has relatively high thermal conductivity.

The reason or reasons for the marked effect of hexachloroethane on fluidity is not immediately evident. Hexachloroethane is a chlorinated hydrocarbon of composition C_2Cl_6 which sublimes at 560 F (293 C). Possibilities which might account for its effect on fluidity include:

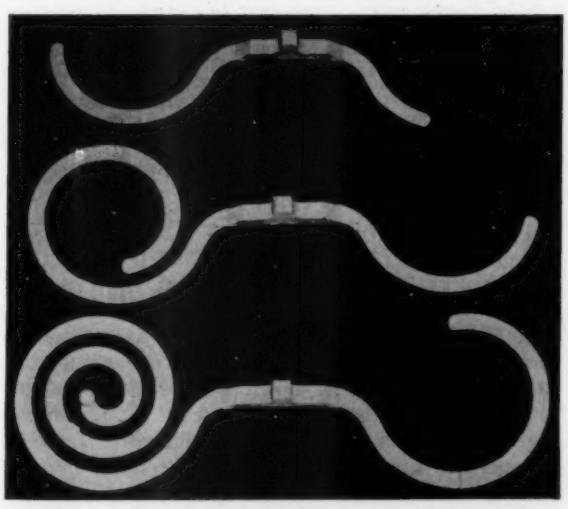


Fig. 12— Double spiral fluidity test castings showing improvement in fluidity due to C_2Cl_6 for Al-4.5 per cent Cu. Left side coated; right side uncoated. Pouring temperature (top to bottom) — 1250 F, 1322 F and 1395 F.

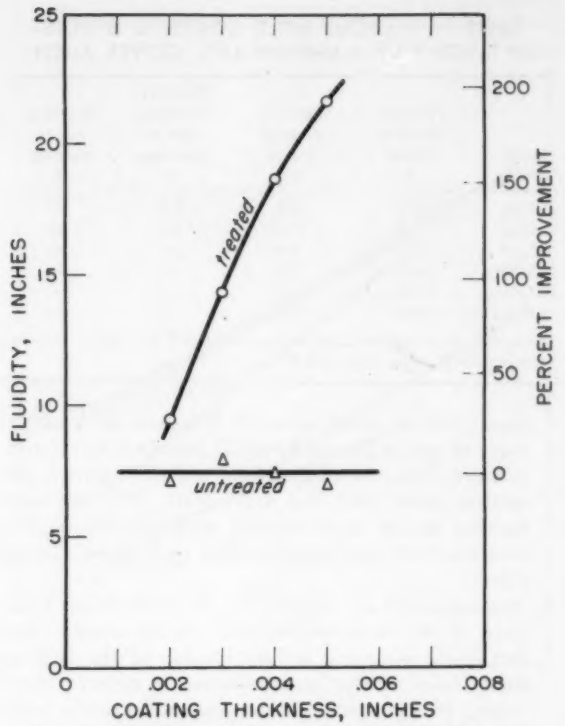


Fig. 13 — Coating thickness effect on fluidity in the double spiral test for Al-4.5 per cent Cu. Mold treatment — hexachloroethane (spray application). Pouring temperature — 1330-1305 F. Mold treatment applied to one spiral in each mold; the other spiral uncoated for control purposes.

1) *Chemical reaction involving alteration of the aluminum oxide skin at the top of the flowing stream.* A typical reaction would be,

$$Al_2O_3 + 3Cl_2 = Al_2Cl_6 + 3/2\ O_2, \text{ at } 1200\ F$$

From available thermodynamic data, the standard free energy of this reaction is an extremely large, positive number,[5] and it is evident the reaction as written will not proceed to any appreciable extent. It is known, however, that chlorine has some physical or chemical effect on aluminum oxide, since chlorine so effectively fluxes aluminum oxide from molten aluminum.

An experiment was made by allowing concentrated chlorine gas to flow through the mold cavity until its presence was noted by emergence of the greenish-brown gas from the sprue. Results seemed to indicate that no simple reaction between the chlorine released by the decomposition of the hexachloroethane and the metal is responsible for the beneficial effects of the compound. Improvement in fluidity resulting from the chlorine treatment was small and not greater than that obtained when a similar experiment was conducted using nitrogen.

2) *Variation of heat transfer conditions between the*

TABLE 1 — VARIOUS MOLD COATINGS EFFECTS ON FLUIDITY OF ALUMINUM-4.5% COPPER ALLOY*

Mold Coating	Fluidity Control Spiral, in.	Fluidity Coated Spiral, in.	Fluidity Increase due to coating, in.	Increase due to coating, %
$ZnCO_3$	8.2	15.1	6.9	85
$CaCO_3$	7.7	14.1	6.4	84
NH_4Cl	8.3	8.9	—	—
NH_4BF_4	8.5	7.3	—	—
NH_4NO_3	7.8	7.9	—	—
$(NH_4)_2CO_3$	7.6	6.8	—	—

*Coatings were applied approximately 0.003 in. thick. Pouring temperatures were 1340-1295 F.

metal and the mold material. It is possible a partial layer of gas at the mold-metal interface could conceivably limit the rate of heat transfer between the molten metal and the mold wall. The increased fluidity in the molds coated with amorphous carbon has been attributed to just such an insulating effect.

3) *Displacement of oxygen by decomposition products of the hexachloroethane.* As an upper limit this could alter the surface tension of the flowing metal from that of aluminum oxide to pure aluminum, or even to that of some aluminum compound exhibiting lower surface tension than pure aluminum.

More detailed discussion and mathematical analysis of the possible importance of the preceding factors are given in the original report on which this paper is based.[5] Results are not, however, conclusive; and it is hoped that analytical and experimental studies now being conducted in the M.I.T. Foundry Laboratory will more clearly delineate the action of hexachloroethane in improving metal fluidity; and that other expedients for enhancing fluidity of molten metals may be found on the basis of the theory developed.

Other Mold Treatments Effects on Fluidity

Several tests were made to study the effect of skin drying on fluidity. Spirals were moderately and heavily skin dried (while retaining the green sand control spiral intact). No improvement was found as a result of the drying.

It will be recalled that throughout most of this study the hexachloroethane has been applied from an ether solution. To determine if the ether itself might have some effect on fluidity, several double molds were sprayed on one side with pure ether. No improvement in fluidity resulted. Other tests have shown that the hexachloroethane improves fluidity whether applied from an ether or alcohol solution, dusted as a dry powder over the mold face or used as a sand additive in facing material.

At the present time, compounds are being selected for further testing if their chemical or physical properties are such that they might affect 1) heat transfer conditions, 2) surface tension of the flowing metal, or 3) surface oxide films of the flowing metal.

Another phase of the study of factors affecting fluid flow of molten metals being given serious consideration is possible alteration of surface tension and oxide forming properties by additives to the metal.

TABLE 2 — HEXACHLOROETHANE EFFECT ON FLUIDITY OF SOME COMMERCIAL ALLOYS AND PURE ALUMINUM (IN THE DOUBLE SPIRAL TEST)*

Alloy	Pouring Temp., F	Fluidity Control Spiral, in.	Fluidity Coated Spiral, in.	Fluidity Increase Due to Coating	Percent Increase Due to Coating
195	1395	13.0	>28.0	—	—
Al-4.5 Cu	1318	6.3	17.9	11.6	184
	1249	3.8	11.0	7.2	190
356	1351	15.5	22.1	6.6	43
Al-7 Si-	1318	11.9	22.1	10.2	86
0.3 Mg	1246	7.2	14.1	6.9	96
220	1372	17.1	>28.0	—	—
Al-10 Mg	1312	10.7	22.9	12.2	114
	1252	10.3	16.5	6.2	60
40E	1350	8.2	13.2	5.0	61
Al-5 Zn-	1299	7.2	13.2	6.0	83
0.5 Mg-	1248	4.0	6.6	2.6	65
0.5 Cr					
99.9 Al	1396	18.7	>28	—	—
	1318	13.8	20.4	6.6	46
	1240	9.9	16.9	7.0	71

*Symbol (>) used when metal ran complete length of fluidity spiral (28 in.). These represent minimum values.

Table 1 lists some representative compounds thus far examined. Zinc carbonate and calcium carbonate effect significant improvements in fluidity; both compounds evolve CO_2 gas at fairly low temperatures which presumably acts as an insulating layer. None of the ammonium compounds tested (and listed in Table 1) have increased fluidity significantly. Most of these compounds are quite hygroscopic, and may even decrease fluidity due to excessive moisture migration to the mold cavity.

FLUIDITY OF SOME COMMERCIAL ALLOYS

Hexachloroethane Effect on Fluidity of These Alloys

The green sand mold fluidity spiral provides a rapid and efficient means for determining the relative fluidities of commercial casting alloys, and for appraising the effects of mold coatings on fluidity. Table 2 and Figs. 14 and 15 summarize results of such studies on five different commercial alloys. The limited number of alloys examined were chosen to be as representative as possible of the wide range of alloys which are ordinarily sand cast. Alloys chosen were pure aluminum (99.9 per cent), 356 alloy (Al-7 per cent Si), 195 alloy (Al-4.5 per cent Cu) and 40E alloy (Al-5 per cent Zn-0.5 per cent Mg-0.5 per cent Cr). Alloys 356 and 195 were of high purity.

In Fig. 14 relative fluidities of the five alloys are plotted versus pouring temperature (no mold coating). The pure metal is seen to be most fluid, and the aluminum-zinc-magnesium alloy least fluid. Data in Fig. 14 agree generally with practical experience in commercial foundries. In Fig. 15, the same data are plotted versus superheat temperature (above the liquidus). This type of presentation clearly indicates the important effect of mode of solidification on fluidity; at constant superheat the pure metal is much more fluid than commercial foundry alloys.

Hexachloroethane mold castings improved fluidity of all the alloys tested (Table 2). The degree of improvement obtained varied from about 43 per cent to 190 per cent depending on the alloy and pouring temperature. All data in this study were for coatings 0.003 in. thick.

Mold Coating Effect on Fluidity in Thin Plate Castings

In some earlier fluidity investigations, data from cast spirals have been found to correlate with actual foundry experiences, while in other studies the correlation has been poor. To assure that results in this investigation would apply to practice, results were checked using a simulated production type casting.

The test casting chosen, shown in Fig. 16, consisted of two ⅛-in. plates, 10 in. wide by 12 in. long. The plates were placed on opposite sides of a large runner and filled from a single tapered spiral. The pouring basin employed was the same as the one used for the spiral castings. Advantages of this design are 1) the extensive thin section plate closely represents actual thin section production castings desired in aircraft and missile fabrication, and 2) the double plate arrangement permits coating one cavity and leaving the other uncoated as a control.

Figures 16 and 17 illustrate the effect of hexachloroethane on fluidity of two aluminum alloys poured into flat plate castings; a fluidity spiral was poured with the plate casting shown in Fig. 17. The casting of Fig. 16 is for 195 alloy and for Fig. 17, 356 alloy. The same tests were made for 220 and 40E alloys with the same striking effects. In all cases one of the plate cavities was coated with hexachloroethane, the other left uncoated. Large improvements resulting from coating the mold are clearly apparent. It has been experience to date that, in this plate casting, the coating permits reductions of pouring temperature of at least 100 to 200 F, while still filling the mold cavity.

The surface appearance of the cast plates was not affected by the mold treatment, except that in one casting poured at a low temperature, depressions from partial misruns or from gas pressure were observed. These depressions, which were often found in fluidity spirals, were entirely absent in plates cast above 1300 F. Radiographic examination of representative flat plate castings showed that the treated plates were entirely free from entrapped gas. Shrinkage was present in some cases in castings made in both the treated and untreated molds.

CONCLUSIONS

1) The green sand mold spiral fluidity tests developed in this investigation have been shown to be reproducible to within about ± 1 in. of metal flow. Slightly higher absolute values of fluidity were obtained in a single spiral design as compared to a double spiral design.
2) Fluidity of an aluminum-4.5 per cent copper alloy increases about 0.6 in. per 10 F increase in pouring temperature (in both the single and double spiral test).
3) For aluminum-copper alloys of varying copper content (0.20 per cent), alloys of approximately 1 per cent copper exhibit minimum fluidity. Fluidity data are presented for these alloys at constant pouring temperature and at constant superheat.

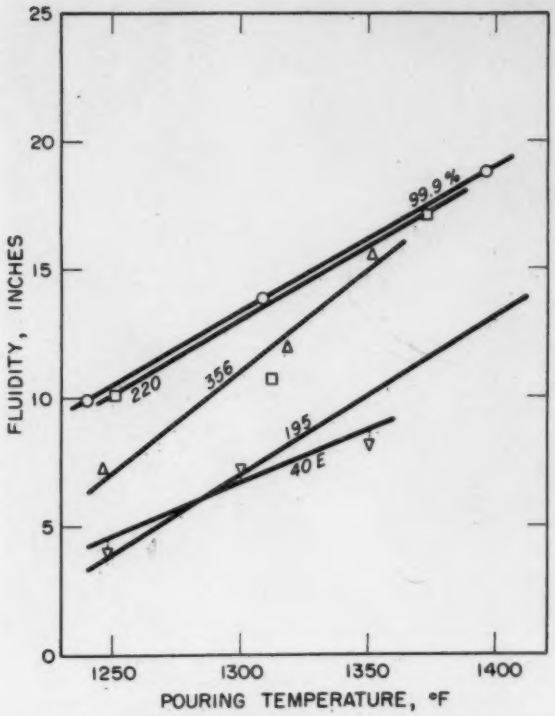


Fig. 14 — Fluidity as a function of pouring temperature of some commercial casting alloys in double spiral test. Data for 195 alloy taken from Fig. 9.

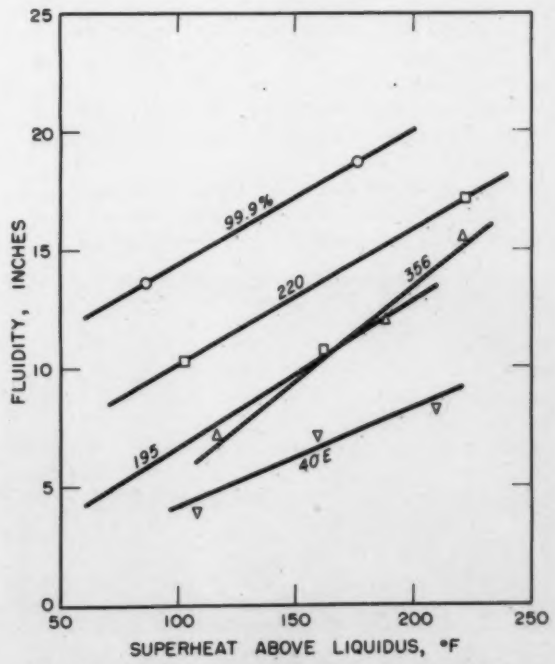


Fig. 15 — Fluidity as a function of superheat above the liquidus of some commercial casting alloys in double spiral test. Data for 195 alloy taken from Fig. 9.

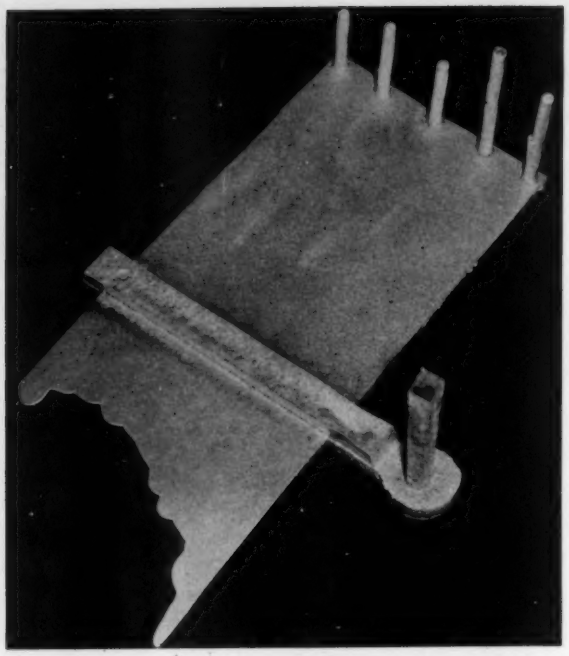


Fig. 16 — Test casting to determine effect of mold treatments on fluidity in plate-like sections. Test casting shown illustrates improvement effected by hexachloroethane treatment on fluidity of 195 alloy. Casting consists of central runner system feeding two identical plates (one on either side) 10 x 12 x ⅛ in. thick. In molding, 5 vent holes ½-in. in diameter are cut at the extremity of each plate. Alloy — 195 (Al-4.5 per cent Cu). Mold treatment — right side coated with 0.003 in. thickness of hexachloroethane, left side uncoated. Pouring temperature — 1400 F.

4) Amorphous carbon mold castings (in a thickness of 0.002-0.003 in., applied by a smoking oxyacetylene torch) improved the fluidity of an aluminum-4.5 per cent copper alloy. Improvement was 130 per cent at 1250 F and 68 per cent at 1400 F.
5) A coating of finely powdered activated charcoal was less effective than torch blacking and a coating of crystalline graphite did not improve fluidity at all.
6) Hexachloroethane mold coatings approximately 0.003 in. thick increased fluidity 190 per cent at 1250 F and at least 120 per cent at 1400 F.

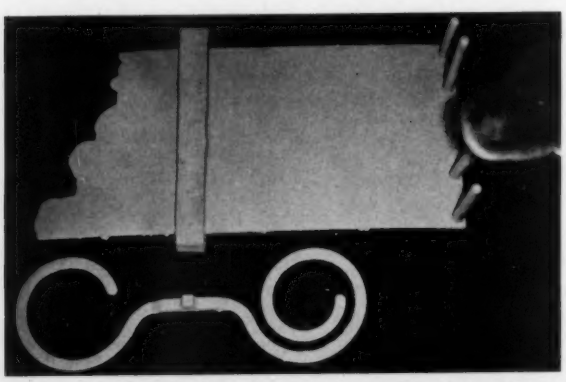


Fig. 17 — Hexachloroethane mold treatment effect on fluidity of 356 alloy. Left side of spiral and plate uncoated; right side coated with 0.003 in. layer of hexachloroethane. Pouring temperature — 1350 F.

7) Improvement in fluidity resulting from hexachloroethane mold coatings was a direct function of coating thickness. At a thickness of 0.006 in., fluidity of aluminum-4.5 per cent copper alloy was approximately tripled (at 1330-1305 F).
8) Improvements in fluidity resulting from other mold coatings (including $CaSO_4$, $ZnCO_3$ and $CaCO_3$) were noted. Insulating properties of the coatings are thought to be important in increasing fluidity but other factors may also contribute. For example, certain of the coatings may alter the surface tension or surface oxide films of the flowing metal.
9) Fluidities of five aluminum alloys were measured in the double spiral test. These were pure aluminum (99.9 per cent), 220 alloy (Al-10 per cent Mg), 356 alloy (Al-7 per cent Si-0.3 per cent Mg, high purity), 195 alloy (Al-4.5 per cent Cu, high purity) and 40E alloy (Al-5 per cent Zn-0.5 per cent Mg-0.5 per cent Cr). Fluidities of the alloys decreased in the order listed.
10) Hexachloroethane mold coatings resulted in substantial improvements in fluidity of all alloys listed above.
11) A flat plate casting was designed to test the effect of mold coatings on fluidity in a simulated production casting. Large increases in fluidity were obtained with hexachloroethane coatings. In the case of aluminum-4.5 per cent copper alloy a 0.003 in. thick coating permitted filling the plate at approximately 1300 F pouring temperature. In the absence of the coating the cavity did not fill completely at 1450 F.

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REFERENCES

1. Ragone, D. V., Adams, C. M. and Taylor, H. F., "Some Factors Affecting Fluidity of Metals," AFS TRANSACTIONS, *vol. 64*, pp. 640-652, 1956.
2. Ragone, D. V., Adams, C. M. and Taylor, H. F., "A New Method for Determining the Effect of Solidification Range on Fluidity," AFS TRANSACTIONS, *vol. 64*, pp. 653-657, 1956.
3. Niesse, J. E., Flemings, M. C. and Taylor, H. F., "The Fluidity of a Series of Magnesium Alloys," AFS TRANSACTIONS, *vol. 65*, 1957.
4. Floreen, S. and Ragone, D. V., "The Fluidity of Some Aluminum Alloys," AFS TRANSACTIONS, *vol. 65*, pp. 391-393, 1957.
5. Foundry Section, Metals Processing Div., Dept. of Met., M.I.T., "Research on Parameters Influencing Fluidity in Aluminum Base Alloys," Final Report, Department of the Army, Ordnance Corps, Contract No. DA-19-020-507-ORD-4503, July, 1958.
6. Feliu, Matas, S., Flemings, M. C. and Taylor, H. F., "Effect

of Mode of Solidification on the Fluidity of Aluminum-Tin Alloys," To be presented 26th International Foundry Congress, 1959.
7. West, T. D., *Metallurgy of Cast Iron*, Cleveland, 1902.
8. Curry, C., "La Mise au Point de L'Eprouvette de la Coulabilite," Association Technique de Founderie, *Fonderie Moderne, 18*, pp. 171, 1924. Test Bars to Establish the Fluidity Qualities of Cast Iron, Exchange paper, Association Technique de Fonderie de France, AFS TRANSACTIONS, *vol. 35*, pp. 289-306, 1927.
9. Moldenke, R., *The Principles of Iron Founding*, New York, 1917.
10. Ruff, W., "The Running Quality of Liquid Malleable Iron and Steel," Iron and Steel Institute, *Carnegie Scholarship Memoirs, vol. 25*, pp. 1-39, 1936.
11. Evans, E. P., "The Fluidity of Molten Cast Iron," The British Cast Iron Research Association, *Journal of Research and Development, vol. 4*, 2, Oct. 1951.
12. Saito, D. and Hayashi, K., "Investigation of the Fluidity of Metals and their Alloys," *Memoirs of the College of Engineering*, Kyoto Imperial University, *vol. 2*, pp. 83-100, 1919; *vol. 4*, pp. 165-178, 1924.
13. Saeger, C. M., Jr. and Krynitsky, A. L., "A Practical Method for Studying the Running Quality of a Metal Cast in Foundry Molds," AFA TRANSACTIONS, *vol. 39*, pp. 513-540, 1931.
14. Taylor, H. F., Rominski, E. A and Briggs, E. W., "Fluidity of Ingot Iron asd Carbon and Alloy Cast Steels," AFS TRANSACTIONS, *vol. 49*, 1-93, 1941.
15. Clark, K. L., "Fluidity Testing of Foundry Alloys," AFS TRANSACTIONS, *vol. 54*, pp. 37-48, 1946.
16. Krynitsky, A. L., "Progress Made in Fluidity Testing of Molten Metals During the Last Ten Years," AFS TRANSACTIONS, *vol. 61*, pp. 399-411, 1953.
17. Briggs, C. S., "Fluidity of Metals," *Metals Handbook*, pp. 199-204, 1948.
18. Eastwood, L. W. and Kempf, L. W., "The Measurement of Fluidity of Aluminum Casting Alloys," AFA TRANSACTIONS, *vol. 47*, pp. 571-582, 1939.
19. Sicha, W. E. and Boehm, R. C., "A Fluidity Test for Aluminum Casting Alloys," AFS TRANSACTIONS, *vol. 56*, pp. 502-507, 1948.
20. Kondic, V., "Fluidity Testing Developments," *Foundry Trade Journal*, pp. 31-36, Jan. 12, 1956.
21. Courty, Andre, "Contribution a l'etude de la coulabilite," *Revue de Metallurgie, Memoirs, vol. 28*, March, pp. 169-182, April, pp. 194-208. 1931.
22. Kondic, V. and Kozlowski, H. J., "Fundamental Characteristics of Casting Fluidity," *Journal Institute of Metals, vol. 16*, pp. 665-678, 1949.
23. Prussin, S. A. and Fitterer, G. R., "Some Requirements for Successful Fluidity Testing," AFS TRANSACTIONS, *vol. 66*, pp. 143-150, 1958.
24. Niesse, J., Flemings, M. C. and Taylor, H. F., "Application of Theory in Understanding Fluidity of Metals," AFS TRANSACTIONS, *vol. 67*, 1959.
25. *Handbook of Chemistry and Physics*, 40th Ed., Chemical Rubber Publishing Co., Cleveland, 1958.

PATTERN ENGINEERING MATERIALS

new horizons

By J. Edward Olson

ABSTRACT

A new rubber-like plastic pattern material was tested to see if it would withstand the pressures and abrasion of normal molding operations. The results of these tests are given. Disadvantages of this material are also reported.

INTRODUCTION

The word *horizon* is defined as the bounds of observation or experience. The title implies that there has been an extension of these bounds so that we may anticipate, or realize, an extension in the life of pattern equipment.

The need for new pattern materials is evidenced by the high initial cost, short life and continuing high cost of maintenance and repair of present high production patterns. One need only look at the rapid growth of the die casting and plastics industry to realize that foundrymen must reduce costs in order to maintain their position in the present competitive market.

Pattern cost is reflected in casting cost, therefore, most foundrymen would like to have patterns that are low in cost and high in life expectancy. The initial cost of original or master pattern equipment will probably maintain its present cost relationship. This reasoning indicates that the immediate problem is to find materials that will substantially increase pattern life and reduce maintenance costs.

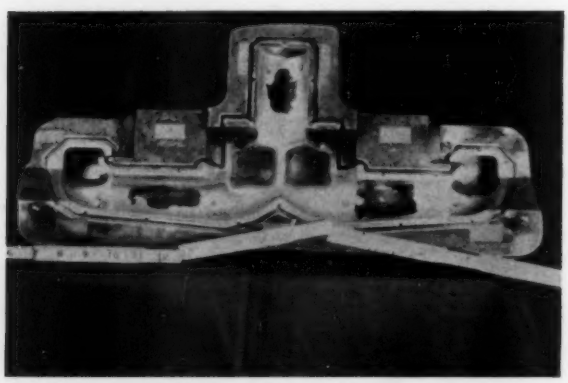


Fig. 1 — Fast plastic repair made under blow tube areas in core box.

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Using the cause and effect analytical approach indicates that more progress has been made in treating the effect than eliminating the cause of poor pattern life. The nature of foundry molding operations such as jolt, squeeze, ram, blow, sling and vibrate contribute to the wearing away of pattern equipment. Since the immediate future does not indicate any radical change in molding technique, the solution to the problem will probably be found in:

a) Development of a better method of repairing existing pattern equipment.

b) Development of a pattern material that will last longer than existing materials.

NEW PRODUCTS

Problem A

An investigation into (a) indicated conventional materials such as plastic wood, epoxy putties, brazing, welding and soldering left much to be desired. This dissatisfaction, and the search for improvement, resulted in the development of a new product.

With this new material, repairs can be made rapidly by use of a plastic compound that cures in a short time. A hardener powder is mixed with the plastic compound and spread on the repair area. Depending on the amount of hardener powder used, the material sets in 5 to 10 min to a hard, dense repair. After hardening, the material may be sanded, milled, drilled, tapped, filled and chiseled without injury or dulling of tools. The finish will accept conventional pattern coatings such as shellac or lacquer, and can be used effectively without coatings.

The compound contains additives that have good slip characteristics which aid in drawing the core or pattern. Other areas of use include increasing fillet size, adding draft and finish and making fast pattern changes. The low cost of the material, the speed with which it can be used, the good wearing characteristics and the reduction in repair time indicate this type of material meets the requirement set forth in (a) (Fig. 1).

Problem B

The approach to the problems in (b) included a review of conventional pattern materials (wood, aluminum, magnesium, iron, steel, phenolics, epoxies,

foamed plastics and masonite). While sand blast tests indicated some materials were more abrasion resistant than others, all the materials fell short of the long life requirements. Patterns made from polyvinyl chloride plastic were evaluated and rejected due to the hardening effect of kerosene and parting agent sprays. Rubber patterns were tested, and while they were highly abrasion resistant there was too much deformation when the rubber was of the ideal hardness to resist abrasion. Also, when the rubber was hard enough to resist deformation the abrasion resistant characteristics were lost.

Other known plastics also were evaluated, and finally one material was selected. This was a rubber-like plastic material having the physical characteristics given in Table 1.

Chemical Characteristics

Resistance to kerosene and fuel oils of this material is excellent. Acids, alkalis and highly reacted chemicals (such as methylene chloride, benzene, carbon tetrachloride, toluene and alcohol) will either destroy the material or cause severe swelling. Although water absorption is almost negligible, there is a possibility of slight swelling in water or in high humidity atmospheres.

TEST PROCEDURES

Abrasion Resistance Tests

The test equipment used was a laboratory type of sand blast unit, a gravity fed sand supply feeding a sand blast gun and an air line pressure of 100 psi. Comparative tests were made with the standard materials mentioned previously, with the results showing that the plastic rubber lasts from 3 to 5 times longer than the other materials tested.

Additional tests of severe conditions not normally experienced in the molding process showed good results. A nozzle was made of the plastic rubber and placed in a grit blast unit. After 1/8-hr of continuous blasting, the nozzle was removed and measured. The inside diameter of the original hole measured 5/16-in. After the test, the hole had increased to 3/8-in., indicating a wear of 1/32-in. on a side.

In another test a section of the material was held directly under the discharge end of a sandslinger. No wear effect was observed.

Impact Resistance

Under the best of conditions pattern equipment may be damaged due to improper handling. The resulting dents, scratches, cracks or marks become transferred to the mold and finally into the casting.

Using a 2 lb ball peen hammer, a 1/2-in. thick piece of plastic rubber was given repeated blows up to the physical limits of the person using the hammer. Close inspection of the product indicated no damaging effects.

File Test

Using an 8 in. bastard file, a 1 in. diameter rod of the material was tested with the file held in the flat position. Under normal pressures, there was no apparent stock removal after repeated filing. However, by using the sharp edge of the file, it is possible to penetrate the material.

TABLE 1 — PHYSICAL AND CHEMICAL PROPERTIES RANGE, SOLID PLASTIC-RUBBER ELASTOMERS

Abrasion resistance	exceptional
Tensile strength, psi	4000 to 6500
Elong. at break, %	400 to 600
Tear strength, psi	300 to 750
Compression modulus at 10% deflection, psi	100 to 1000
Hardness, shore A	60 to 98
Modulus of elasticity, psi	at 300%, 600 to 3000
Compression set resistance	good to outstanding
Heat resistance	to 250 F, continuous duty to 350 F, intermittent duty
Specific gravity	1.24

Molding Tests

The first molding tests involved the use of tubular plastic rubber measuring 3 in. inside diameter, 3 1/2 in. outside diameter and 3 in. long.

Test A. Conventional core sand was hand rammed into the tube. After filling the tube it was rapped lightly, and although there was no draft in this piece the plastic rubber drew perfectly. No release agent was used, yet there was no evidence of sand adhering to the inside diameter of the plastic rubber mold.

Test B was a repetition of test A, with the exception that CO_2 sand was used. The green core was drawn without rapping. The draw was perfect, and there was no evidence of sticking. The second part of this test was similar, but this time the core was gassed in the mold. Again, the draw was perfect and there was no evidence of sticking.

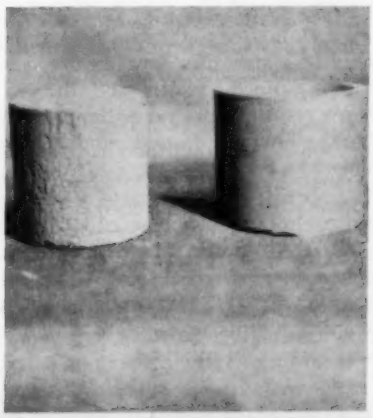


Fig. 2 — Preliminary molding tests in CO_2, core sand and green sand.

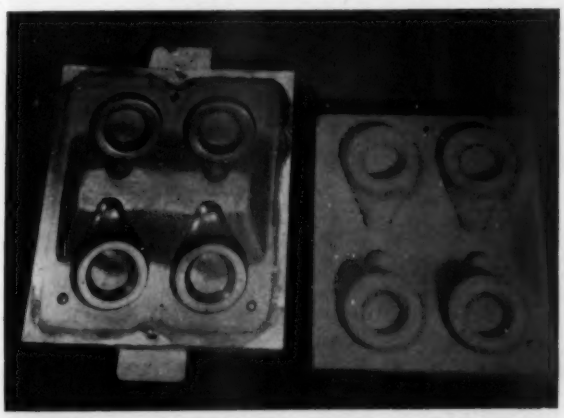


Fig. 3 — Further molding tests in CO_2, core sand and green sand.



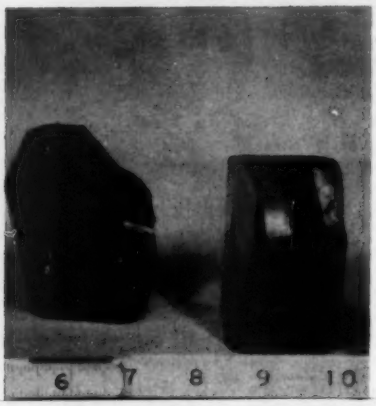


Fig. 4a (above) and 4b (left) — Molding automotive cores with plastic rubber inserts.

Test C involved the use of standard green sand, as used in a gray iron foundry. Sand was rammed in the mold, rapped slightly and drawn. Again, a perfect draw resulted, and no evidence of sand sticking in the mold. The mold used, and the resultant core, are shown in Fig. 2.

Test D involved the use of a pattern which simulated a matchplate. The base of this matchplate was made of cast aluminum which had been coped out allowing a ⅛-in. space between the coped out area and the master. Plastic rubber had been molded and bonded to this matchplate which affected a continuous surface ⅛-in. thick.

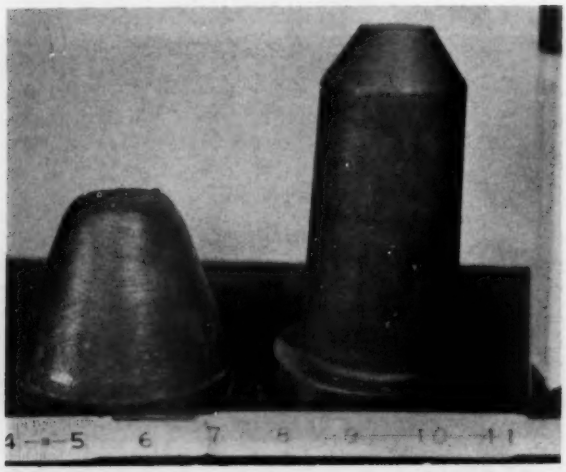


Fig. 5 — Molding feeder — risers.

The molding test on this unit first involved the use of CO_2 sand (Fig. 3). A frame which simulated a flask was set around the pattern, the mold was filled, rammed, struck off and drawn. The draw was as it was for the previous tests, and there was no apparent sticking of the CO_2 sand. Additional molds were made under the same conditions with equal results.

This same mold was taken on a jolt squeeze machine using conventional green sand. The mold was filled, jolted and squeezed. After roll over and draw it was noted that there was some sticking sand. By spraying a standard parting agent, sticking was overcome. A serendipity effect was realized in that the mold hardness was increased approximately 20 per cent by using this plastic rubber mold as compared to an aluminum matchplate. This hardener ram condition was due to the alternating deformation of the plastic rubber during the jolt and squeeze operations.

This same mold was taken under a standard sand-slinger and molds were made. It was observed that the molds were much harder than normal. The pattern drew perfectly, and there was no evidence of sand sticking.

Test E involved the use of plastic rubber inserts in an automotive cylinder head, water jacket core box. The conventional method involved the use of cast iron inserts which were unsatisfactory due to their

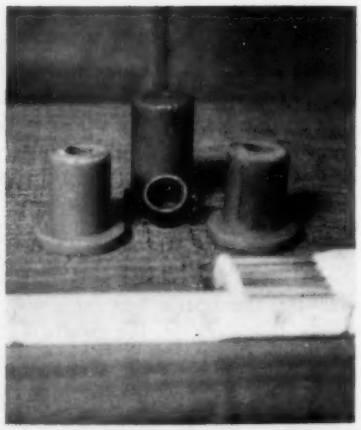


Fig. 6 — Core box blow tubes in plastic rubber.

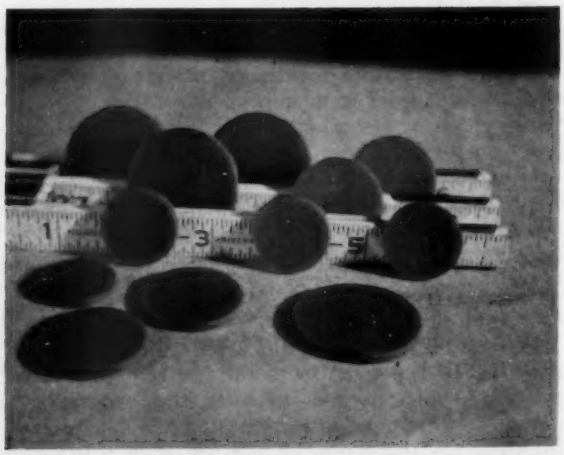


Fig. 7 — Reducing wear under the blow tubes with plastic rubber inserts.

poor abrasion resistance. This type of insert showed wear after only 500 core blowing cycles. Plastic rubber inserts were made forming the core for the area immediately above the combustion chamber. While not conclusive, the tests as of this date indicate that the plastic rubber is lasting longer than the cast iron, as illustrated in Figs. 4a and 4b.

Test F involved the use of a plastic rubber part to form riser-feeder mold cavities in the jolt squeeze molding process, as shown in Fig. 5.

This process was accomplished by filling and jolting the mold. Steel forms or patterns the shape of the feeder-riser were attached to the squeezer head. As the mold was squeezed the feeder pockets were formed. This type of molding technique resulted in heavy abrasion at the lead edge of the forming pattern (called a terracing or snow plough effect). Understandably, the life of the steel parts was short.

This same part was then made in plastic rubber, and while the results are not conclusive indications are that the plastic rubber will outlast the steel.

Pilot Pin Problem

Another problem eliminated was in a pilot pin used to guide the feeder pattern during the squeeze operation. The grinding action of the sand between the hole in the feeder pattern and the steel pilot pin cause rapid wear to the hole and pin. This resulted in a fin in the mold at the bottom of the feeder cavity which resulted in sand inclusions or improper feeding action. The use of plastic rubber eliminated this due to its wiping action on the pilot pin.

Another effect of the use of this rubber in the above mentioned instance was that during the squeeze operation the plastic rubber pattern was permitted to bottom and slightly deform. As the pressure was relieved on the squeeze operation it returned to its original shape thereby giving clearance during the draw and effecting a hard ram in the mold.

Test G. Based on the results obtained in the grit blast test, tests were begun to determine the efficiency of the plastic rubber as a blow tube material. This application is pictured in Fig. 6.

A production test lasting five weeks on a two-shift basis showed no wear to the blow tube after 34,000 blows. The only physical change noted was in the length of the tube which had increased 1 32-in. This did not present a problem as the excess was removed by grinding and the tube returned to productive use.

Test H. The area receiving the severest wear other than the blow tube is probably that directly under the blow tube (Fig. 7). Inserts of plastic rubber were made approximately ¼-in. thick and cemented using a special adhesive in the affected area. Thicker inserts for contoured areas were also made. The contours may be ground using high-speed hand-held grinding tools. Long life and stick-free operation are indicated by preliminary tests.

Test I. Previous work was done using synthetic rubber-like materials to eliminate wear at the flask bearing surface on matchplates, cope and drags and flasks. Stray sand particles lying under the flask bearing surface are an excellent abrasive medium during the jolt and squeeze operations. With the conventional aluminum to steel contact, and in the presence of sand during the jolt squeeze operation, the action is so severe as to cause deep grooves in the match or cope and drag plate and eventual destruction of the flask bearing surface.

Although synthetic rubber has done a satisfactory job in increasing the life of patterns and flasks, indications are that plastic rubber will further increase life expectancy. Figures 8a and 8b illustrate this application.

Adhesives

One of the major problems encountered in the tests was the development of a suitable adhesive for binding the plastic rubber to metal. This problem has been overcome, and a suitable adhesive has been developed which is economical and simple to use.

DISADVANTAGES

While the preceding tests are not conclusive, indications are sufficient to warrant an accelerated program of high production testing. Such a program is now under way in all of the areas mentioned.

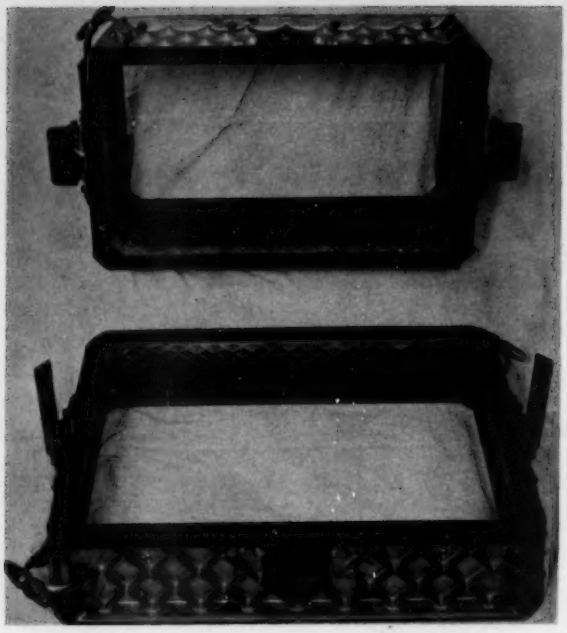
The main disadvantage of plastic rubber rests in the hands of the fabricator of the raw plastic rubber material. The handling of some of the additives is hazardous. Some of the materials are volatile and are conducive to producing allergies and/or asthmatic conditions in persons sensitive to these materials. The material is also a skin irritant, and is somewhat active on the respiratory system and mucous membranes. Extreme care must be exercised in their handling at all times. However, the hazard can be reduced measurably by the use of adequate ventilation.

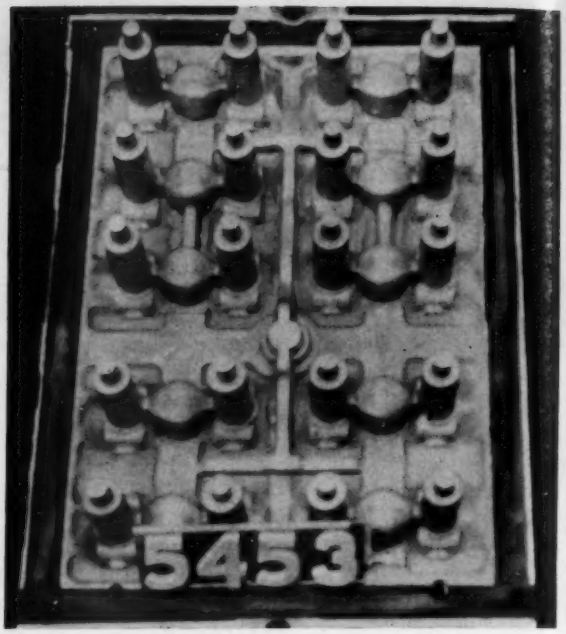
The method and the materials demand extreme care in formulating and processing. The complexity of the process involves heat, vacuum, precise proportioning, close time limits and extremely fast handling when the batch is ready. Wrong mold temperature or an excessive amount of release agent, as well as a near miss in any of the mentioned areas, will result in scrap parts. Changes in relative humidity can also have a negative effect on the quality of the end product.

While these problems seem somewhat insurmountable at times, work in the area of simplification holds promise for the future.

Shrink and Growth

In addition to the above disadvantages, there is one major area that will seriously affect the use of plastic rubber as a pattern material. This is the problem of shrink and growth of the cured product. Changes in formulation, mass and configuration can make problems in shrinkage and growth. Because of this and the other reasons previously mentioned, it is suggested that the initial studies and evaluations by pattern shops and foundries be performed by an experienced fabricator of plastic rubber. Such a service, which might be compared to that of a matchplate manufacturer, is available.





Figs. 8a and 8b — Application on the flask bearing surface (only required on one surface).

CONCLUSIONS

While the above tests are not conclusive, they were sufficiently successful to indicate that plastic rubber has a definite future as a pattern engineering material. Test results seem to indicate that reductions in repair and maintenance costs may be achieved by its use.

NOTE: The uses of plastic rubber as described in this paper or in the uses as: patterns, core boxes, inserts, blow tubes, vents, flasks, blow plates and wear strips, are covered by patents applied for and pending.

INTRODUCTION OF TENTATIVE HOT SHELL DEFORMATION TEST

Report of AFS Sand Div., Shell Mold and Core Committee 8-N prepared by Roderick J. Cowles

ABSTRACT

Literature references are cited which establish that instrumentation and test methods have been developed to evaluate the deformation changes which occur near the surface of conventionally formed mold walls subjected to molten metal. According to these references, these "hot deformation" data have been correlated with the quality of cast surfaces produced by molds of duplicate composition. The results are a guide to the selection of molding materials, molding methods and casting design to develop optimum castings.

Recent surveys and experimental studies conducted by AFS Committee 8-N have emphasized the importance of mold wall movement and shell cracking caused by molten metal poured into shell molds. Production of optimum surface shell molded castings evidently depends considerably upon a knowledge of the hot deformation characteristics of shell mold compositions, just as it has for conventional molds.

Initial results have shown the value of a new test method which applies heat at a rapid rate to one side of shell mold specimens simultaneously loaded transversely. Changes in deflection, conditions of break and rapid deformation are correlated with time, temperature and load application. There is an initial movement towards the heat application in opposition to the load, followed by a sagging between supports in the direction of and proportional to the load. Thermal expansion of the refractory grains differentially heated on one side of the specimen is the reason for the arching against the load. There is a similar occurrence in actual casting practice.

An analysis of initial test data indicates that variables in the shell mold product such as mold wall thickness; refractory grain composition; size, size distribution and configuration of the grain; type and quantity of phenolic binder applied as a powder mix or as a precoat on the refractory grain; degree of cure and the effect of additives will all require consideration in correlating hot deformation results with quality of casting surfaces. This information has been used as a guide for establishing tentative standard equipment and test methods for controlling hot deformation in production, and as a tool for research studies of shell molds.

INTRODUCTION

The improvement of casting quality related to the requirements of performance and appearance is a never ending objective of the foundry technologist. An important phase of this promotion for better castings has been the increase in development studies to determine what occurs at the mold surface, and the refractory binder structure behind the surface, when hot molten metal is poured and cooled to solidification against the mold wall.

Numerous studies and reports[1,2,3,4,5,6] supported by reliable data have assisted in clarifying the causes of "rattails," "buckles," "scabs," "veining" and "cuts and washes," which are the result of mold changes when subjected to high temperatures of molten metal. The knowledge obtained through instrumentation in such studies has been and will continue to be valuable for alleviating these problems in foundry practice.

Free expansion of the mold composition with temperature changes; variations in mold strength properties with gradient heating and isothermal heating; dimensional deformations developed as stress-strain data and expressed as modulus of elasticity; hot deformation data[2] correlated in such factors as "hot deformation rate" and "hot toughness" have all contributed to determine improvements in molding materials and casting design which minimize the problems associated with high temperature at the mold wall.

These studies have all been applied to conventional green sand and dry sand molds with little attention to the basically different sand-phenolic resin shell molds. A recent shell molding survey conducted by 8-N Committee[7] emphasized that more problems of shell production, such as mold cracking and surface defects, could be attributed to mold-wall movement and thermal shock as the shell is subjected to the hot molten metal, than any other cause for scrap castings. Simultaneous with this survey, an experimental "Study of High Temperature Properties of Shell Molds"[8] was carried out.

The results again showed the need for more knowledge of the changes in the shell mold structure when subjected to molten metal conditions. The high temperature shell expansion information obtained in this study[8] showed the value of such test data in correlating casting defects with the shell mold properties.

The committee members further responded with a preliminary evaluation of instrumentation, test methods and demonstrated data, in order to establish a Tentative Hot Shell Deformation Test. This

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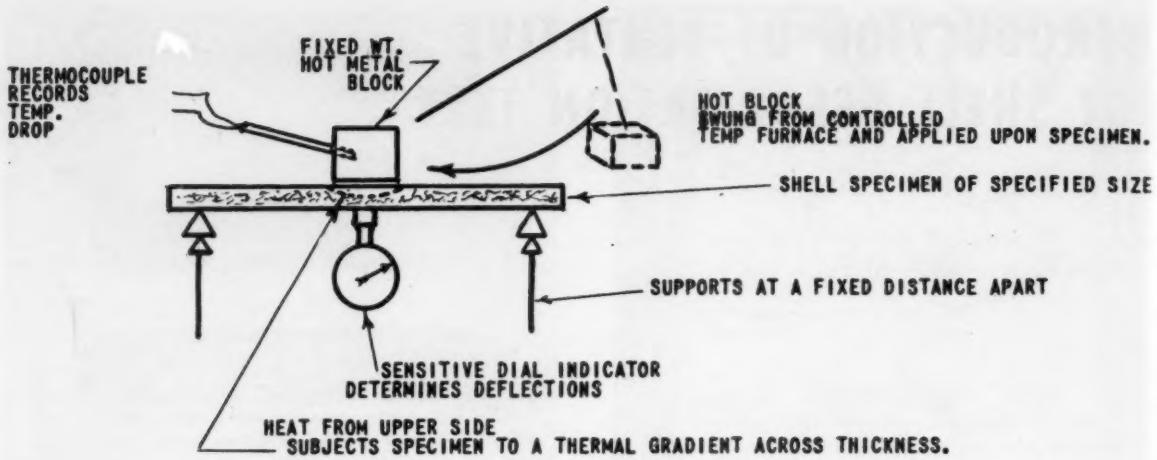


Fig. 1 — Principle of operation of instruments A and B for hot shell deformation evaluation. Data are charted as dimensional deflections with changes in temperature (instrument A) or time (instrument B). The variables of the test to be either fixed or recorded are 1) weight of metal block, 2) area of block adjacent to specimen, 3) temperature of block when applied to specimen, 4) dimensional deflections with drop in temperature of heat block (instrument A) or with time of heat application (instrument B), 5) rate of heat conductivity (a time-temperature relationship), 6) thickness and width of shell mold specimen, and 7) distance between supports.

test is developed as one possible source of data which can be correlated with shell mold composition and structure, guiding the elimination of casting imperfections.

The test is designed to show hot deformation similar with the actual occurrence of molten metal heat applied to one side of a cold shell mold wall. This is more representative of actual casting conditions, and provides additional information to that obtained by the conventional free expansion of heated cylindrical specimens.

TEST METHODS STUDY FOR DETERMINING SHELL MOLD THERMAL PROPERTIES

At least five different manufacturing organizations with representatives on Committee 8-N, submitted hot deformation data for shell mold specimens obtained by testing three resin coated sands. The coated sands were prepared at one source, properly sampled and distributed for individual evaluation. No particular testing equipment or method of test was designated for determining data pertaining to hot shell deformation.

The results were fundamentally alike, and established certain characteristics of shell deformation due to high temperature heat applied to one side of a precured shell specimen.

Three of the test instruments, the principle of their operation and the representative charts of data determined by these tests, are used here to portray this general approach to hot deformation evaluation of shell molds. Figures 3, 4 and 5 show the three instruments considered and are designated respectively as A, B and C. The principles of their operation are diagrammatically given in Figs. 1 and 2, and their data represented in the charts of Figs. 6 and 7.

In every test the shell specimen deformation is initially in opposition to the heated metal block and to the direction of applied load. In all cases the specimens were 1 in. wide and ¼-in. in thickness, with

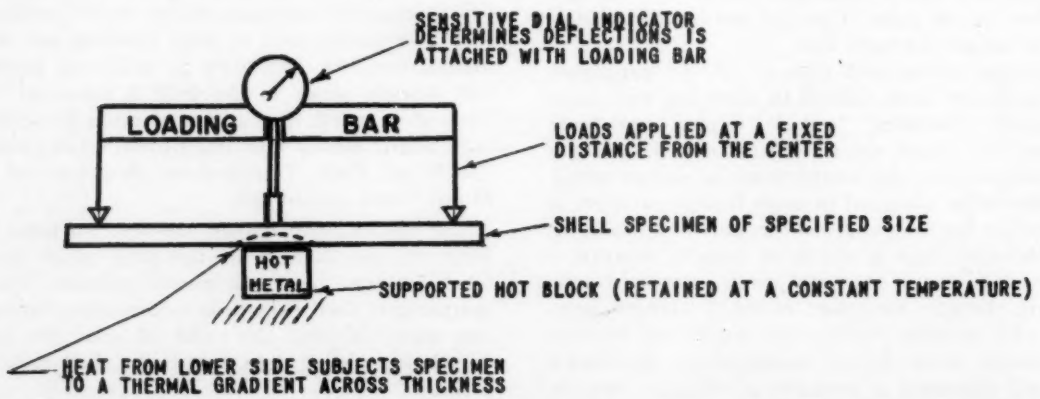
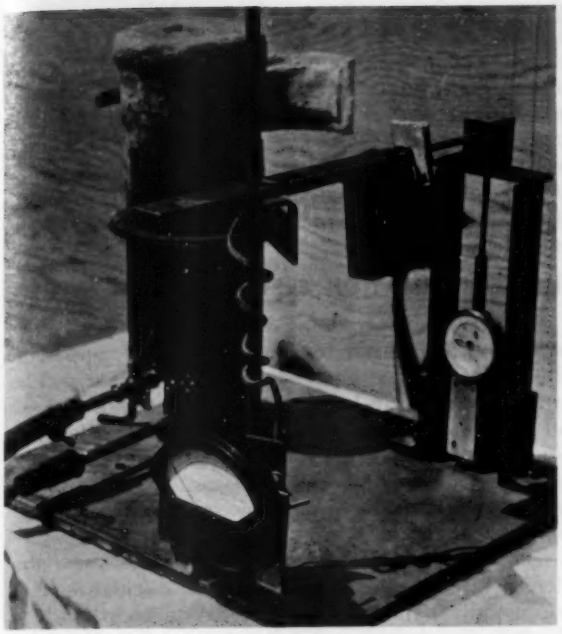
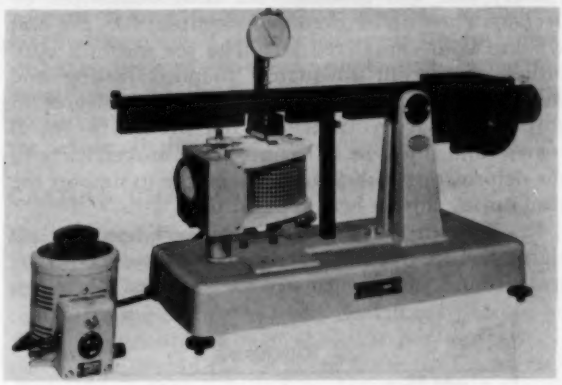


Fig. 2 — Principle of operation of instrument C for hot shell deformation evaluation. Data are charted as dimensional deflections with time of heat application. Variables of the test are 1) weight of load applied, 2) area of block adjacent to specimen, 3) temperature of block when applied to specimen, 4) dimensional deflections with time of heat application, 5) rate of heat conductivity (a time-temperature relationship), 6) thickness and width of shell mold specimen, and 7) distance between supports.







Figs. 3, 4 and 5 — The three test instruments considered for the hot shell deformation evaluation test for which their operation principles are given in Figs. 1 and 2. Figure 3 (above), instrument *A*. Figure 4, (upper right), instrument *B*. Figure 5, (right) instrument *C*. Data charts for these instruments are given in Figs. 6 and 7.

distances between supports varying between 3 to 3.375 in. and loads from 258 grams to 1.5 lb. This arching of the sample toward the heated side disregarding the opposing load stress is generally explained as an expansion of the sand grains as they are heated. The heat conductivity through the sand and resin bonded structure is slow enough to cause selective expansion of the grains first subjected to the high temperatures.

The size of grains, their configuration, their degree of packing and the intrinsic property of thermal coefficient of expansion for the refractory grain in use is expected to vary the degree of this movement for any specified conditions of test. In every case studied, there was an initial expansion of the heated side.

As the specimen reaches a uniform temperature the refractory grains become uniformly expanded and the resin binder is weakened by thermal effects. This results in a load response causing the specimen to sag down between the supports. The distance of movement related to time will vary according to 1) the load applied, 2) the total heat transmitted to the sample, 3) the rate of heat transmittal, 4) the variables of the refractory such as its coefficient of thermal expansion, size, size distribution, and grain configuration, 5) heat flow properties of the resin binder, 6) the degree of resin cure, 7) the quantity of resin used, 8) degree and rate of resin degradation with heat application, and 9) the effect of any additives or included foreign materials.

This interrelation of multiple variables always increases the problem of evaluation and complicates the correlation of test data with use results. This hot deformation test is another valuable tool for providing data which can be used with other test information to better unravel and predict the variations in casting product quality with the materials and fabrication methods used to produce the shell molds.

EXAMPLE OF HOT DEFORMATION STUDIES

The constructive criticism of any evaluation method is best developed by experience in production and development application. Instrument *A* has been in use for more than 6 years as a shell molding production control tool. Instrument *C* is a later modification, which allows improved precision and greater flexibility of test for research studies. Figure 7 is a chart to demonstrate the value of data obtained with instrument *A*. This is a plot of dimensional deformation changes with drop in temperature of the heating block. This is a time dependent variable and is better correlated by instrument *C* where the temperature is retained constant and the deformation plotted against time of heat application.

The general differences shown in Fig. 7, however, are quite adequate to analyze differences in shell performance. There are, of course, different evaluation objectives when pouring bronze compared with steel as well as considering differences in casting sizes and configurations.

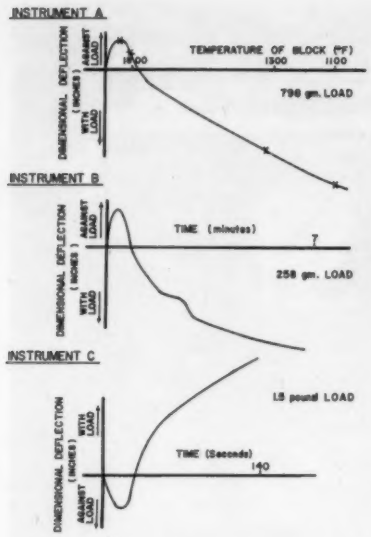


Fig. 6 — Representative data charts for instruments A, B and C for hot shell deformation evaluation.

The *A-3* sand resin combination is low in hot rigidity and will probably not hold dimension, especially when undercured as in sample *A-3*. *B-6* and *C-6* are both considered adequate for medium sized bronze casting in adequately planned (gating and risering) shell molds. Evidently, a certain degree of cure is necessary to contribute hot strength to the *B* coated sand since the *B-3* samples are broken early in the deformation rather than continuing to support the load during further heat application.

Results such as *C-6* may be approaching a strength rigidity under conditions of heat which cause cracking in the mold and hot tears in the castings. These test data together with tensile strength, transverse strength, free expansion data, bulk density of the molding composition and the knowledge that this is a round grain sand which is more easily close packed, can be analyzed in relation to actual shell performance. The effect of selected additives or changes in refractory grain composition on the hot deformation results, plotted for analysis as in Fig. 7, will establish the value of this test for obviating such problems as mold cracking and hot tears.

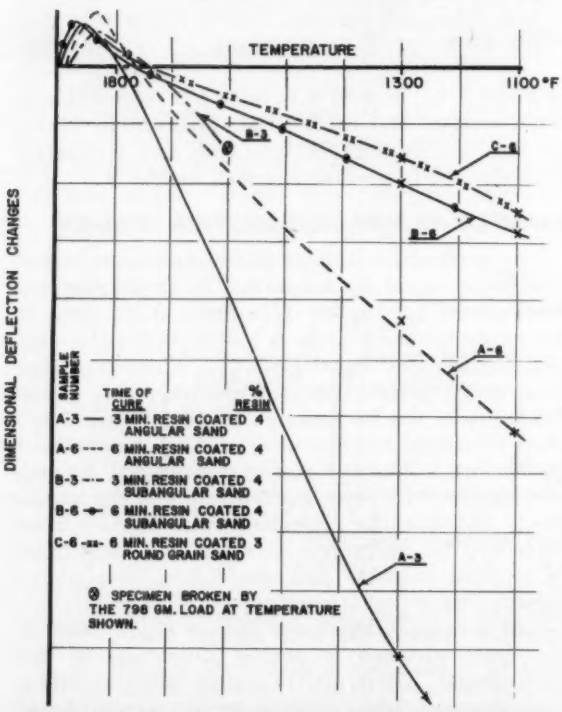


Fig. 7 — Plot showing variations of deformations due to differences in shell specimens. Data for this graph obtained with instrument A.

HOT DEFORMATION TEST POTENTIAL VALUE

The value of extending this fundamental test instrument for research studies becomes apparent. Instrument *C* can be used in numerous ways. The applied weight can be kept low during the first 15 sec of heat application to emphasize sensitive differences in the refractory grain expansion with heat. Following this movement, which opposes the load application, an increase in applied weight would best determine hot shell rigidity and strength.

The heat deformation of equal samples under different loads would provide data for calculating elastic modulus, and with rupture data a factor such as hot toughness of the shell can be determined. These approaches can be studied under various isothermal conditions to further extend the knowledge of pouring metals of different solidification temperatures into shell molds.

The thickness of the shell, and the manner of supporting the shell during casting, will further determine the relative importance of initial refractory grain expansion and subsequent hot sag deformation.

SUMMARY OBJECTIVE

This is not a report of completed development studies in hot deformation of shell molds, but rather an introduction of a valuable testing tool for studying such production problems as shell cracking and hot tearing of castings. Committee 8-N is presenting a tentative test based on these introductory instruments, methods and data examples. Further refinements and knowledge of the test will develop as use experience is correlated with committee work, with outside research studies and with individual publications.

REFERENCES

1. H. W. Dietert, V. M. Rowell and A. L. Graham, "High Temperature Sand Tests," Official AFS Exchange Paper, MODERN CASTINGS, p. 36, Nov. 1956.
2. H. W. Dietert and T. E. Barlow, "Hot Deformation of Molding Sand," AFS TRANSACTIONS, *vol. 66*, p. 7, 1958.
3. N. C. Howells, R. E. Morey and H. F. Bishop, "Properties of Molding Sands Under Conditions of Gradient Heating," AFS TRANSACTIONS, *vol. 65*, p. 402, 1957.
4. N. C. Howells, R. S. Morey and H. F. Bishop, "Properties of Molding Sands Under Conditions of Gradient Heating," Met. Div., Metal Processing Branch, Naval Research Laboratory, Report 4857, 1956.
5. W. B. Parkes and R. G. Godding, "Behavior of Molding Sands at High Temperatures," *Foundry Trade Journal*, p. 473, Oct. 1955.
6. H. W. Dietert, "Processing Molding Sand," AFS TRANSACTIONS, *vol. 62*, p. 1, 1954.
7. AFS Committee 8-N, "Shell Molding Survey," AFS TRANSACTIONS, *vol. 66*, p. 559, 1958.
8. R. A. Rabe, "Study of High Temperature Properties of Shell Molds," AFS TRANSACTIONS, *vol. 66*, p. 484, 1958.

Mn-V-Mo AGE-HARDENING AUSTENITIC STEEL

foundry characteristics

By N. C. Howells and E. A. Lange

ABSTRACT

Successful founding of the new Mn-V-Mo age-hardenable austenitic steels depends upon the use of special techniques with regard to melting, pouring, mold materials and heat treatment. Melting procedures for minimizing hydrogen content of the molten metal and principles of gating design are described. Sand mixtures with a zircon sand base, containing 5 per cent silica sand and 1 per cent core oil as special additives, promote high quality casting surfaces. Tensile properties of the metal can be controlled by the aging phase of the heat treatment; 145,000 psi tensile strength, 110,000 psi yield strength and 12 per cent elongation are attained with an aging treatment of 8 hr at 1300 F.

INTRODUCTION

Widespread use of conventional nonmagnetic casting alloys aboard minesweepers is precluded by their cost, limited mechanical properties or in some cases their physical properties. In order to more extensively utilize the economical and versatile casting process for producing complex shapes, naval engineers require an austenitic steel with a yield strength in the range of 100,000 psi.

The initial phase of an investigation, aimed at developing austenitic steels with high-strength properties, was concerned with the adaptability of recently developed wrought alloys to the casting process.[1] Two wrought alloys were successfully cast and heat treated to high strength levels, but both of the alloys utilized nickel as a primary stabilizer for the austenitic structure. A new family of austenitic alloys containing a minimum amount of nickel was subsequently developed for the purpose of providing a material for emergency conditions requiring conservation of nickel.

The new alloys are essentially manganese stabilized, austenitic steels with age-hardening characteristics promoted by vanadium and molybdenum. Alloys in the composition range 0-2.5% Cr; 14-18% Mn; 0.25-0.60% C; 0.25-0.7% V and 0.4-3.5% Mo can be age-hardened to the range 250 to 500 Bhn and remain nonmagnetic (permeability less than 1.2). Although the manganese content of these alloys is only slightly higher than the manganese content of Hadfield steel, the age-hardenable alloys have much lower carbon contents, and special foundry techniques are required to obtain sound castings with good surface finish.

The heat treatment for the new alloys is also more involved than the heat treatment for Hadfield steels. The development of effective foundry techniques for producing sound castings with good surface finish, and the effects of time and temperature during the aging phase of the heat treatment on the tensile properties of an alloy with a nominal composition, are detailed in this report.

CASTING CHARACTERISTICS

Melting and Pouring

In melting Mn-V-Mo alloys, the first consideration should be given to the fact that steels with 14 per cent manganese are effectively deoxidized because of the high amount of manganese. It has been established that deoxidized or killed heats have a greater affinity for hydrogen than an unkilled heat.[2,3] Reference is made to experimental work with acid open hearth melting where the hydrogen content of the steel increased 1 to 3 ppm in a few minutes after blocking the heat. This situation is not analogous to the induction melting used here, but, in a qualitative sense, caution in the handling of fully killed steels is definitely indicated.

Table 1 contains the list of charge materials and Table 2 the log for a typical heat. The steel was melted in a high-frequency 350 lb induction furnace, and about 1½ hr were required to bring the metal to pouring conditions. Dry argon gas was piped through a loosely fitted cover into the crucible at a rate of flow generous enough to maintain a substantial amount of argon in the furnace atmosphere. A carbon boil was produced just prior to adding the alloy elements to purge hydrogen in the initial-charge armco iron.

A vigorous boil was maintained for several minutes by adding 0.10 to 0.15 per cent carbon, and allowing it to react to about 0.03 per cent carbon. An addition of ferrosilicon was then used to kill the boil. Although the inert gas atmosphere was maintained primarily to minimize hydrogen pickup, it also reduced the manganese loss, which was 17 per cent without the

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TABLE 1 — LIST OF MATERIALS FOR A 300-LB MELT

Alloy	Weight lb	Weight oz
Armco iron (0.03% C, 0.01% Si)	231	1/2
Ferrosilicon (50% Si, 50% Fe)	1	13
Regular ferromanganese (6.7% C, 80% Mn, 13% Fe)	12	8
Electrolytic manganese	39	1/2
Nickel	9	0
Ferromolybdenum (59% Mo, 41% Fe)	10	4
Ferrovanadium (53% V, 46% Fe)	3	8

TABLE 2 — HEAT LOG FOR 300-LB MELT

Time, pm	Action
1:00	Furnace turned on. Charge of nickel plus armco iron, 90-kw power, "low" tap.
1:20	Furnace changed to "next to low" tap. Inert gas introduced into furnace (argon).
1:35	Furnace changed to "next to high" tap.
1:45	Furnace changed to "high" tap.
2:15	Slag removed from metal. Inert gas turned off.
2:15	Carbon boil caused by adding about 0.15% carbon and boiling away carbon to 0.03%.
2:20	Ferrosilicon added. Ferromanganese and the electrolytic manganese added.
2:25	Ferromolybdenum added, followed by ferrovanadium.
2:25	Protective slag cover added. Inert gas (argon) reintroduced into furnace.
2:35	Metal tapped into ladle at 2900 F. Molds poured.

protective atmosphere and 13 per cent with the protective atmosphere. Following the alloy additions, a basic slag cover was maintained on the metal in order to minimize heat loss and gas pickup.

The metal was tapped at 2900 F into a preheated teapot ladle, and the castings were poured at 2800 F. Figure 1 gives a general view of the gating system used for the castings in these experiments. The principle features of the gating system are the choke at the sprue base for controlling the mass flow rate (lb/sec), an enlarged runner bar for limiting the linear flow rate (in./sec) and an ingate entering the riser. Both the sprue and the ingate are larger in cross-section than the choke. The dimensions of the gating system were as follows:

Part	Area (sq in.)	Ratio to Choke Area
Choke	9/16	1
Sprue	11/16	1.2:1
Ingate	13/16	1.45:1
Runner Bar	1 3/4	3:1

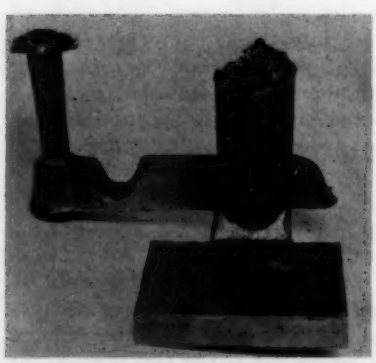


Fig. 1 — View of 4 x 6 x 1 in. plate with riser and gating used for tensile property and surface quality experiments.

The total casting weight was about 25 lb and required a pouring time of 7½ sec. This represented a flow rate of 3.3 lb/sec on a weight basis, or 12 cu in./sec on a volume basis. The average linear flow of the metal through the runner was calculated to be 7 in./sec. Since the enlarged portion of the runner bar was 4½ in. long, the metal was in this portion of the gate a little more than ½ sec. It was found that a flow for this length of time within the runner bar is effective for trapping incidental slag which has entered the sprue with the molten steel.

Mold Materials

Initially, sand mixtures for molds for the new Mn-V-Mo steel were patterned after conventional sand mixtures for Hadfield steel, i.e., silica sand containing additives to provide for a reducing atmosphere. However, severe mold-metal reaction occurred when the Mn-V-Mo steel was poured into molds of sand containing such additives as core oil, seacoal, graphite, pitch and cereal. The increased mold reactivity of the Mn-V-Mo steel was attributed to its higher manganese content and its lower carbon content, which necessitated a higher pouring temperature.

Zircon sand was then used, and while no mold-metal reaction occurred with this more refractory material, a lap type of defect was formed, as illustrated in Fig. 2. The laps were apparently caused by a tenacious manganese oxide film which formed on the surface of the molten metal and restrained the flow momentarily, forming a defect with a cold-shut appearance. After 1 per cent core oil was added to the zircon sand to provide a reducing atmosphere, the surface laps generally disappeared (Fig. 3).

A consistent, protective environment is difficult to achieve with atmospheric casting, and the lap defect occasionally recurred. A small amount of silica sand was then added to the zircon base sand with the aim of decreasing the surface tension of the oxide film. It turned out that an addition of 5 per cent silica sand was sufficient to consistently eliminate the lap defect (Fig. 4). When higher percentages of silica sand were used (10 to 15 per cent), the mold-metal reaction was sufficient to cause a rough surface. The silica portion of the mold material was also incorporated



Fig. 2 — Lap type of defect formed on plate casting when sand of the following composition was used; 100 lb 140-mesh zircon sand, 0.5 lb western bentonite, 1.5 lb southern bentonite, 0.5 lb corn flour and 2.1 per cent moisture before baking.



Fig. 3 — Surface laps generally disappeared from plate casting when sand of the following mixture was used; 100 lb 140-mesh zircon sand, 1 lb core oil, 0.5 lb western bentonite, 1.5 lb southern bentonite, 0.5 lb corn flour and 2 per cent moisture before baking.



Fig. 4 — The surface lap defect was consistently eliminated from plate casting when sand of the following mixture was used; 95 lb 140-mesh zircon sand, 5 lb silica sand (no. 80), 1 lb core oil, 0.5 lb western bentonite, 1.5 lb southern bentonite, 0.5 lb corn flour and 2.2 per cent moisture before baking.

in a wash rather than throughout the body of the mold material.

A higher percentage (10 per cent) of silica could be used in the wash to make it more reactive and still achieve a smooth casting surface, because a limited amount of silica was available. The surface quality obtained from a zircon sand mold containing 1 per cent core oil, and coated with wash containing 90 per cent zircon flour and 10 per cent silica flour, is illustrated in Fig. 5.

Risering

The production of sound castings free from shrinkage involves two prime considerations:

1) The proper size of riser, which means that there must be sufficient molten metal stored within the riser to furnish feed metal to compensate for shrinkage.
2) Temperature gradients must be such that a continuous channel is maintained from the riser to the advancing front of completely solid metal throughout the solidification period.

Although the research on the shrinkage of this steel is limited, the results of one series of experiments with a plate 4 x 6 x 1 in. can be used as a guide. Using the Naval Research Laboratory formula which had been developed for Class B steel,[4] the upper limit for the riser volume was a 2½-in. riser 2¼ in. high. However, when a 2½-in riser was used and poured to a 6-in. height, shrinkage was evident in the casting. A 3-in. diameter riser resulted in a borderline case, some castings exhibiting shrinkage and some not. In actuality, the feeding distance may have been too long for this composition.

HEAT TREATMENT AND TENSILE PROPERTIES

The feeding distance was 4 in., which is 4 times the thickness T of the plate. The specified feeding distance for the Class B steel is 4½xT, or 4½ in. for a 1-in. thick plate.[5] Thus, it appears that this alloy is more difficult to feed than the Class B steel, and consequent precautions should be taken.

The heat-treating procedure for hardening these steels in its basic form resembles that for the majority of age-hardening alloys. The metal is first solution treated at a high temperature in order to dissolve the hardening phase into the matrix. When this is accomplished, the metal is quenched to produce a supersaturated condition. An aging treatment at some intermediate temperature then produces the hardening by reprecipitating the hardening phase. The cooling rate from the aging temperature is not critical, and air cooling is generally satisfactory.

Thus, the Mn-V-Mo alloys differ from low-alloy steels, since alloy steels are softest after quenching from the solution temperature and hardest after heating to what is ordinarily considered to be a high tempering temperature. This investigation was primarily concerned with the aging treatment, because the balance between high yield strength and ductility is primarily controlled by the aging treatment. A few preliminary tests to establish the solution treatment had shown that no appreciable benefits could be derived with long holding periods at 2100 F, so all specimens were held for 1 hr at 2100 F and quenched in water.

Figure 6 is a graph of hardness versus aging time for a Mn-V-Mo steel with the nominal composition. The relative positions of the curves for the three temperatures in Fig. 6 are basically similar for the vast majority of the hardness curves that were ob-



Fig 5 — Plate casting made in mold from sand mixture of Fig. 3, and coated with the following mold wash; 2000 gm 400-mesh zircon flour, 200 gm silica flour, 37.5 gm western bentonite, 20 gm dextrin plus water to thin.

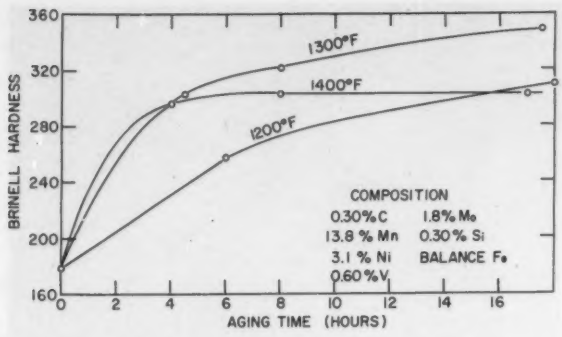


Fig. 6 — (left) — Time and temperature during aging treatment effect on hardness of a precipitation hardening austenitic steel.

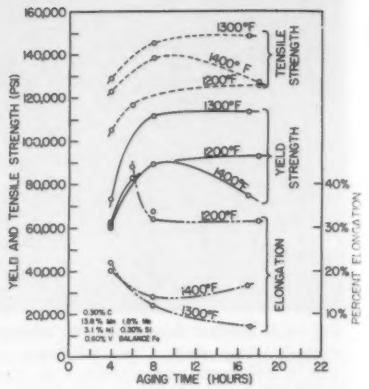


Fig. 7 — (right) — Aging conditions influence on tensile properties of a precipitation-hardening austenitic steel.

tained for steels with other compositions. However, the maximum hardness attainable with a standardized solution and aging treatment is strongly influenced by the composition, most of all by the carbon, slightly less by the vanadium and to a lesser degree by the molybdenum.

Aging Treatment Effect

The influence of the aging treatment on the tensile properties is shown in Fig. 7. The pattern of the tensile properties follows that of the hardness curves, that is, the 1300 F tensile and yield strength curves reach the highest values; the 1400 F values are intermediate; and the 1200 F tensile properties gradually increase and exceed the maximum properties attained at 1400 F. From these data aging temperatures between 1200 and 1300 F can be recommended for best control of properties.

However, there is an advantage to using temperatures approaching 1300 F when maximum yield strength is desired, because the maximum yield strength is reached in the shortest time at 1300 F with little danger of overaging. After aging 8 hr at 1300 F, the tensile properties of the sample steel were 145,000 psi tensile strength, 110,000 psi yield strength and 12 per cent elongation. If higher ductility (20 to 30 per cent) is desired, an aging temperature between 1200 and 1250 F is indicated.

The Mn-V age-hardenable austenitic steels are machinable when techniques comparable to those used for the austenitic stainless steels are used, i.e., carbide tools, low speeds and high feed rates. Machine operations should be performed after the casting has been age-hardened. Metal in the as-cast condition can only be cut by abrasive methods, and after solution annealing it has a tendency to load the tool and gall.

SUMMARY

The following foundry process conditions are recommended for melting and casting Mn-V-Mo age-hardening austenitic steel:

1) Minimize hydrogen content by using clean, dry materials, a carbon boil and a dry inert gas atmosphere for induction melting or protective reducing slag for arc melting.
2) Pouring temperatures should be high, 2775 to 2850 F.
3) The gating system should have a runner choke and sufficient runner size to permit the metal to remain in the runner for ½ sec before entering the mold.
4) The following type of sand mixture should be used for molding material:

Sand	Additives (% of sand)
95% Zircon	1% core oil
5% Silica	0.5% western bentonite
	1.5% southern bentonite
	0.5% corn flour
	2.2% water

5) Risers should be slightly larger and feeding distances slightly shorter than those calculated by N.R.L. methods for carbon steel.

The following heat treatment schedule is recommended for hardening Mn-V-Mo age-hardening austenitic steel:

1) Solution anneal 1 hr at 2100 F.
2) Water quench.
3) For maximum yield strength with nominal ductility age harden for 6 hr at 1300 F. For controlled yield strength and high ductility age harden at 1200 to 1250 F.

Castings in the age-hardened condition can be machined with techniques comparable to those for austenitic steels, i.e., carbide tools with slow speeds and heavy feeds.

ACKNOWLEDGMENT

This work is part of a broad investigation of cast austenitic steels conducted at the request of Mr. G. Sorkin, Code 343, Bureau of Ships, Navy Department.

REFERENCES

1. Lange, E. A., Howells, N. C. and Bukowski, A. T., "Cast Age Hardenable Austenitic Steels," AFS TRANSACTIONS, *66*, 519-525 (1958).
2. Van Voris, F. E., "Controlling C_1 N_1H_2, O in Steel Melting," MODERN CASTINGS, *34* (No. 10), 28-30 (1958).
3. Kerlie, W. L. and Richards, J. H., "Origin and Elimination of Hydrogen in Basic Open-Hearth Steels," *Jour. of Metals, 9*, 1541-1548 (1957).
4. Bishop, H. F., Myskowski, E. T. and Pellini, W. S., "Simplified Method for Determining Riser Dimensions," AFS TRANSACTIONS, *63*, 271-281 (1955).
5. Bishop, H. F. and Pellini, W. S., "The Contributions of Riser and Chill-Edge Effects to the Soundness of Cast Steel Plates," AFS TRANSACTIONS, *58*, 185 (1950).

TIME STUDY AND METHODS TRAINING FOR SUPERVISORS

By John Taylor

ABSTRACT

The need for supervisor education in time study and methods is discussed. A suggested program lasting for six or seven meetings is given.

INTRODUCTION

Time study, motion analysis and incentive systems are valuable management tools. However, for best results they must be properly used. Progressive foundry managements thus have come to realize that such tools will be used most effectively, and the business will profit most when foremen and supervisors are trained to know the real details of motion analysis, time study and the basic principles involved in the application of incentives to the activities they supervise.

Time and again it has been demonstrated that when foremen realize, through training, what can be accomplished by a good motion analysis study, or an activity study on some particular operation, they look for and find other improvements which also can be made.

In addition, foremen who have received proper training relative to the department incentive plans answer the questions of operators and union stewards and prevent dissatisfaction. In being able to answer such questions the prestige of the foreman is increased as well as the workers' respect for him. He then, of course, is in better position to "sell" the system to them. In addition, management benefits from having supervisors trained to use the proper tools for accurate estimating.

SUPERVISOR EDUCATION

Educating supervisors on this subject is neither costly nor time consuming. Usually it is not necessary to take the foreman away from the job for an extended period in order to give him an intensive training on time study and incentives, although this has been done in some of the larger foundries with excellent results. Where such training can be accomplished, the man returns to the job with a broader knowledge of management requirements, plus an inspiration to look for details which ordinarily would not be noticed.

No foreman should be expected to take time studies in his department, but the results of the training will save time for both the time study and personnel departments, and the foreman is more likely to institute methods improvements on his own initiative.

WEEKLY MEETINGS

When it is not possible to provide a training program over an extended period, which frequently is the case, weekly meetings can help to broaden the knowledge of the foreman and enable him to do a much better job. It also will provide a needed stimulant in the attitude toward management in general, and the realization that he is a part of it. Foremen not only are the men on the "front line" but also the logical source for future executives. Therefore, top management can strengthen its position by increasing the knowledge of the supervisory group.

Meetings of approximately one hour's duration can be held each week, either on a general discussion basis or with a qualified leader. In meetings of this kind, too much time should not be spent on fundamental details, but rather on a broad view of the subject.

SUGGESTED PROGRAM

A suggested program, for approximately six or seven meetings might be as follows:

1) Origin of time study; several kinds of incentive systems.
2) Hazards of estimated prices; the necessity for time studies; value of methods analysis; how these subjects are taught in colleges.
3) Recognition of allowances for fatigue, personal, supplemental, balance and incentive; the use of levelling factors and why they were developed.
4) How a time study is made; actually taking one or two simple studies in the meeting room.
5) The use of studies to build standard data; development of constant and variable values; compiling data for operations.
6) Actually setting some time standard by standard data; the many uses of standard data.
7) Checking accuracy of the data by "proof sheets," the completed standard data or "specification sheet."

CONCLUSIONS

Foreman training is essential to the installation and successful use of a new time study, methods or

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incentive program to eliminate delays, misunderstandings and even mistrust of the engineer or the system. These reactions do exist and management should make every possible effort to dispel them and the reasons for them, through a properly planned training program in the fundamentals involved.

People are not skeptical of a good program or of a good tool when properly trained in its use. This is true in all walks of life, and it has been demonstrated that when applied to foundry supervision the problems of time study and incentives diminish or disappear.

DUCTILE IRON CASTINGS VERSUS CARBON STEEL FORGINGS AND WELDMENTS

some comments and recent conversions

By J. L. Salbaing

ABSTRACT

The advantages of ductile iron castings over carbon steel forgings and weldments from the viewpoint of metal structure, isotropy and directionality, strength and production are considered. Case histories of parts which were made by the former processes and converted to ductile iron castings are given.

INTRODUCTION

Product designers and purchasing agents are confronted daily with the proper selection of the most economical and most efficient metal shaping and fabricating methods. Much has been said or written on this subject during recent years, as evidence seems to point out that the foundry industry has been placed in a defensive position. Few foundrymen will disagree with a comment made recently in a trade magazine by Norman F. Hindle:

> "The foundry industry points with pride to the fact that casting is probably the most ancient (and possibly the most economical) method of metal shaping. Let us not forget that the same makes it the most vulnerable to loss of market by new methods of metal forming."

The promoters of these more recently developed methods have used aggressive tactics to penetrate the foundry's market. Using diversified means and constant efforts, they succeeded in striking the minds of the people responsible for the design and production of component parts.

The discovery of ductile iron in 1943 was the first major advance in ferrous metallurgy since Boyden made blackheart malleable iron. Magnesium bearing irons have helped foundrymen reverse the trend towards carbon steel forgings and carbon steel weldments. Whether used in conjunction with sand molding, or newer molding techniques, ductile irons lend themselves perfectly to the production of a number of component parts of high strength and quality at lower finished cost. Ductile iron castings are replacing an increasing number of parts previously made as forgings or weldments.

First, a comparative discussion of ductile iron casting, carbon steel forging and carbon steel weldments from metallurgical, structural and production standpoints, then, a series of examples showing the actual benefits derived when a part was cast in ductile iron will be presented.

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DUCTILE IRON CASTINGS VS. FORGED AND WELDED PARTS

No attempt will be made to cover all respective advantages or disadvantages of any of the three fabricating methods. This discussion will be as a reminder of some fundamental considerations, the knowledge of which might encourage the foundryman to carry on his "sales efforts" successfully.

Metal Structures

Wrought products are said to be affected to a lesser extent by metal impurities than castings. Looking at the solidifying process of a liquid metal, of any alloy composition, we see that impurities present in the melt having a lower freezing temperature than that of the base metal solidify last. The greatest portion of the melt being already solid, these impurities are forced between the solid crystals where they turn from liquid to solid. In general, they are nonmetallic and brittle. If the casting thus obtained is stressed, these inclusions usually would be the initiating point of rupture in the material.

The various working processes to which a wrought metal is subjected from its original form as ingot or billet, will tend to break up the continuity or the size of these brittle inclusions present at the base metal grain boundaries.

Theoretically, a wrought shape should exhibit a superior ductility along the main axis of forming. However, the validity of such an influence depends upon many factors. Most important among these would be the presence of impurities and the extent of their deposits. A metallurgically clean cast metal would certainly eliminate most of these disadvantages, and foundrymen possess in the magnesium bearing irons a series of metallurgically clean metals. In order to consistently obtain a structure where the graphite is in a true spheroidal shape, the base metal composition should be maintained within rigid limits. Elements such as antimony, arsenic and bismuth, or metallics such as selenium, tin and titanium are not tolerated beyond certain maxima.

In addition to the metal purity derived from the selection of furnace charge raw materials to melt the ductile iron base metal, the precise and particular molding techniques required to cast ductile iron play their part. The gating systems should be designed with care to eliminate dross or slag inclusions in the mold. They also should provide a minimum of turbulence in the rapid flow necessary to fill the mold quickly. A 1000-lb casting should be poured in approximately 20 sec, while a 10,000-lb casting should be poured in a maximum time of 65 sec. These requirements oblige the foundryman to use the best known practices in gating the molds. Surface defects will show unless the molding sand is prepared properly.

The above requirements in quality of raw material and care of molding operations, without which a true and generalized spheroidal graphite structure could not be formed, decrease the advantages of the wrought metal for structural applications.

Isotropy and Directionality

The theory of elasticity is the basic concept upon which the knowledge of strength of materials is built. The many formulas used every day for the structural analysis of machine tools, engine parts or other structures, are a direct derivation from this principle. However, the basic hypothesis upon which its validity lies is seldom questioned: Is the material homogeneous and isotropic? Or, in other words, what assurance has the designer that the stresses developed on a plane perpendicular to the direction of the applied force will remain the same when the force is rotated to any other direction? If the member being analyzed possesses this property, its actual performance under static loadings will be approaching that established by calculations.

The goal of foundrymen is to obtain castings exhibiting such a structure. It is recognized that the process of casting is, among all other methods, the one most likely to achieve this required equi-directional structure through the minimum amount of operations. Obstacles lie in the way, however. Foundrymen must succeed in regulating the cooling rate of the liquid metal in the mold so that the crystalline structure will be independent of the metal section. Foundrymen must also use specific techniques to eliminate the unknown thermal stresses remaining within the casting after it has reached room temperature.

In recent years new molding methods have been developed to enable foundrymen to achieve this goal. Progressive solidification is a well understood phenomenon, and casting designs are modified frequently to permit it. The undesirable residual stresses are progressively eliminated either through changes in the design or through molding technique and heat treatment.

It has been stated that if foundrymen are capable of pouring a casting so as to reach progressive solidification, prevent hot spots and use clean metal and procedures, they may obtain a crystalline structure, uniform, homogeneous and entirely satisfactory for all purposes.

In recent years developments in strength of materials have focused the importance of such factors as the stress raisers on the ability of a structure to resist the applied loads in a predictable manner. Surface imperfections and abrupt changes in section thicknesses create points or areas where the stresses developed are seldom predictable in both intensity and direction by means of the normal methods of structural analysis. However, one fact is certain—the more homogeneous and isotropic the material is, the less chance for these stresses to develop and reach, in some directions, values beyond the material's ultimate strength.

It is then clear that materials exhibiting directionality in their properties, as unavoidably all wrought materials do, will be more susceptible to the stress raisers effects. The importance of directionality is, in the author's opinion, too often neglected. It is not uncommon to see gears being cut on a production basis from a steel plate. This is hardly justified from a structural standpoint, since the mechanical properties will vary from one tooth to the other.

The designer should be aware that a specimen taken in the plane of rolling, and in the direction of the fibers of an aluminum plate, shows as expected 20 per cent elongation. However, this ductility will be reduced to 8 per cent when the specimen is taken in a plane perpendicular to the rolling direction and at an angle of 90 degrees to the plane of rolling.

Finally, while the metallurgical history of a casting is short and well known, the similar history of a wrought part is long and unknown in most cases. Foundrymen, in their sales efforts, should take notice of the following comment made by G. Sachs:

> "The actual properties of a metal depend considerably upon its method of processing or its previous history. For a wrought material, this previous history is not readily defined, as it usually consists of a number of hot working, cold working and annealing operations, each of which can be varied in several respects. The effects of many of these factors are not well understood. Changes in the processing conditions, therefore, frequently result in a product that appears distinctly different from that obtained before the changes were made."

Strength

Since the discovery of ductile iron, foundrymen have enjoyed an enviable position in the structural component field. Never before has their position been so strong, because no cast ferrous material could hardly compete from a combined strength and cost standpoint with carbon steel fabricated shapes. It was not a difficult task for the promoters of weldments or forgings to prove the structural inadequacies of one material or the cost of the other.

The various examples presented later point out the possibilities and achievements of ductile iron castings in applications where strength is the paramount design requirement.

The forging industry has been justly concerned with the ductile iron casting potentialities. In an address presented at a meeting of The Drop Forging Association in 1957, Harry W. McQuaid said the following:

> "What has kept the steel forging so far ahead of the steel or iron casting in the eye of the machine

> designer is its great ability to permit a high degree of deformation without fracture. As long as this condition existed, the threat of the casting to the forging was not especially strong. During the past ten years, there has been introduced into the picture a new type of iron casting known as "nodular" or "ductile iron"' which shows definitely improvement in physical properties over the ordinary cast iron part. To the drop forger, this might mean an an important invasion of his field and a possible loss of important tonnage to the foundry."

This potential loss of tonnage also applies to weldments. However, the author would like to sound a word of caution at this point. In their anxiety to compete with weldments from a strength and structural standpoint, foundrymen should not forget that the welding industry has done a commendable job in the field of applied strength of materials. It suffices to look at the voluminous amount of research performed, and literature published by this industry, in an attempt to clarify some particular structural problems faced by the designers. Whether it is the establishment of design criterions for welded connections or the study of a composite beam under various loadings, this continuous concern to provide tools for the designers in the field of experimental stress analysis in structures sometimes not too closely related to welding has paid off handsomely by the gain of markets and the stature of the industry.

Proposed Studies

Because ductile iron is a true structural metal, similar programs should be carried out by the foundries interested in furthering its use. Frequently used cast shapes should be systematically studied to arrive at optimum designs, for the minimum weight and lowest final cost. Although some foundries have undertaken such programs of experimental stress analysis on their own to help their customers reach more satisfactory designs, this practice is not yet generalized. To be of value, such results should be available to all designers and structural engineers and not to a few.

Consequently, it is the author's opinion that these studies should be carried out on an industry-wide scale. The tools are brittle varnishes, photo-elastic coatings, fringe pattern analysis, magnetic particles for qualitative analysis; photo-elastic measurements, strain gages, x-ray analysis, etc., for quantitative analysis. The results would be invaluable to the machine designer.

It is felt that this would do much to show the superiority of ductile iron castings where there is no joint to limit the overall strength. A welded structure is no better than its joints. Besides the residual stresses, and the somehow unpredictable distribution of loading stresses, a welded joint depends greatly upon a human factor. Such discontinuities as sharp corners, notches, weld beads, hard spots, slag inclusions and porosity, which cause obstructions to the flow of stress, are common sources of failure in weldments subjected to repetitive loadings. These points should be a source of concern for the designer contemplating a welded design.

A properly designed and properly cast ductile iron casting eliminates these dangers. Foundrymen are aware of this, but the designer should be continuously kept informed as well as unequivocally convinced.

Production

Ten years of production experience have proved that ductile iron can be used for long or short production runs for parts that shall be machined or not.

Even for short runs it appears that the cost of pattern investment has been overemphasized to such an extent as to detract the production engineers from even thinking in terms of castings. Actually, experience indicates that the patterns, when made of the proper material, selected to last for a known quantity of castings, do not affect the casting unit cost to the extent of being out of proportion with weldments.

The various steps and operations required to complete a weldment gas cutting, positioning, welder or automatic welding units, roll bending, press forming, stress relieving and material handling between each operation have to be accounted for in the final cost analysis of a welded component. They certainly add up to an impressive list, particularly when long production runs are contemplated.

Weldments do not lend themselves too well to machining operations, while ductile iron castings can show savings over other materials. The machining difficulties encountered in welded assemblies are too well known to be developed in detail here. Mention should only be made that the weldment residual stresses cause distortions of the members. The removal of some surface metal will cause the redistribution of these induced stresses. Consequently, the part will go out of shape after machining, sometimes making it impossible to obtain the desired match or alignment.

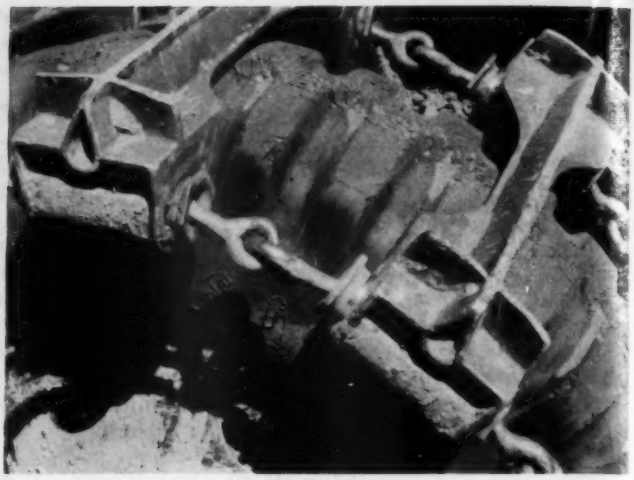
From a production standpoint, it is evident that the cost of a forging die is not economically justified unless the total quantity of parts numbers high in the hundreds. For instance, a large mobile equipment manufacturer has set up for the engineering department the following rule—all parts which are expected to be run yearly in quantity less than 800 shall be castings. Above 800, the costs between a casting and a forging should be examined.

Cost Reduction

Helping foundrymen are the newer molding methods that were mentioned previously. Combined with ductile iron, they permit foundrymen to produce components of high dimensional integrity. By combining the reduction in excess metal to excellent machinability, ductile iron castings enable the producer to achieve certain cost reductions. The author's company knows of an automotive concern that saved 30 per cent of its original cost in switching its production of forged steel gears to shell molded ductile iron castings (grade 120-90-02). The test bars taken from castings purposely broken showed a complete consistency in mechanical properties. The lowest tensile strength registered was 127,000 psi.

The cost reduction obtained in the production of automobile crankshafts by some manufacturers here and abroad has been achieved with ductile iron





Figs. 1 and 2 (left to right) — Installation of single track shoe wrapped around tandem rear wheels for conversion of truck into half-track vehicle, used for bad weather conditions. Right view shows closeup of installation.

castings. To date, some 20,000,000 crankshafts have been made successfully. Yet all the possible combinations of design ideas and production methods toward lowest finished cost have not been exhausted.

An example is a 50 lb finished weight ductile iron V-8 engine crankshaft. A substantial weight reduction could be accomplished by using a hollow design such as the one made by suppliers for a German car. The geometry of this casting, obtained through precise founding techniques, would be close to the finished dimensions. A subsequent short cycled drop forging or hot coining would further bring the casting down to within a few thousanths of an inch of the final shape. Grinding of the bearing surfaces would complete the part.

This suggested procedure, bold in its approach, is aimed at a vital component of an automobile engine. If iron castings are to hold out against the white metals in the automotive and other fields, it will be through the continuous use of imagination to obtain the maximum performance from the metal properties.

Hot working a ductile iron casting is not a new idea. Harry W. McQuaid proposed it to the Drop Forgers, in an address previously mentioned in this paper, as follows:

> "I would like to submit for your careful consideration in the forging field, the possibility of using cast ductile iron blanks which need the very minimum of hot working to convert them into real forgings. These forgings would be, if properly cooled, of excellent machinability and physically satisfactory . . . (ductile iron) conversion from a cast to a forged product is not difficult to visualize and with enough incentive from the large consumer of forgings and castings, this conversion will follow without fail."

From an industrial engineering point of view, it is also interesting to compare the energy required to finish a part by forging to that required to make a casting. Jean Fauquemberque, in *Fonderie,* Oct., 1958, reports a figure of 0.7 kwh/lb for electrically melted castings against between 1.4 to 1.7 kwh/lb for a forging. It is worth while to mention this comparative cost aspect as it might be of some interest when contemplating the installation of facilities to produce large quantities of parts.

CASE HISTORIES

Case 1 — Single Track Shoe

This part is a patented design. The track shoe is an element of a strand of track, wrapped around the tandem rear wheels of a truck to convert it into a half-tracked vehicle (Figs. 1 and 2). The shoes (Fig. 3, right) are joined together by dumbbells (Fig. 4) which engage in a two position as-cast bayonet-type socket. Approximately 45 shoe elements are required for a standard vehicle.

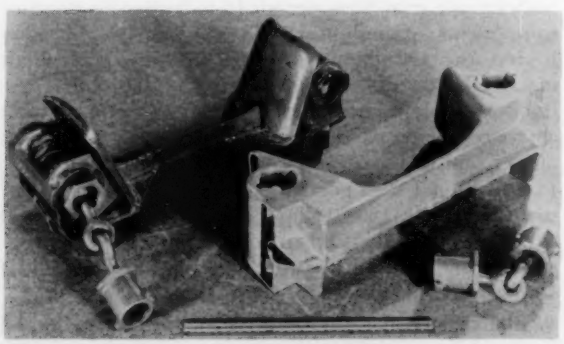


Fig. 3 — Single track shoes shown installed in Figs. 1 and 2.

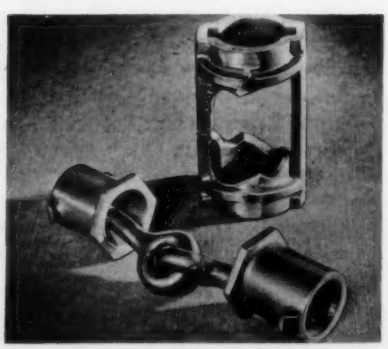


Fig. 4 — Shoes (Figs. 1, 2 and 3) are joined together by dumbbells which engage in a two position as-cast bayonet-type socket.

The part was previously made by welding two ductile iron castings to two mild and medium carbon steel components, stress relieved, hard-face welded on the bottom beam and flame hardened on cam surfaces. Because of the severe service requirements, the complete part was converted to a ductile iron casting with a subsequent improvement in physical properties, a weight saving of 15 per cent and a total cost reduction of 45.5 per cent.

The bushings are shell molded, and thus require a minimum of finishing to clean the spherical seat. This is accomplished by one man at the rate of 40 castings per hr. The castings are supplied in a normalized condition, while the ramps where the bushings engage, are flame hardened. A detailed comparison between the two types of fabrication is as follows:

	Weight and Cost	
	Weldment (Fig. 3, left)	Casting (Fig. 3, right, Fig. 5)
Weight, lb.	20	17
Cost of raw material	4.56	4.85
Fabrication labor & overhead	9.18	0
Assembly, painting & bundling	0.75	0.75
Heat Treatment: stress relieving, oil quench & draw	2.00	3.40
Total cost	16.49	9.00

The added advantages found with the castings are:

1) By the decrease in the number of operations, the parts can be produced quicker as castings than as weldments.
2) No fabrication set-up equipment is required with casting.
3) The casting offers much more flexibility to the choice of possible heat treatments, as opposed to the cumbersome steps which had to be used to reach the required properties in the past – normalizing, oil quenching and drawing, flame hardening, hard face welding.
4) The reduction in weight obtained in the redesign is of prime importance to the military users as track has to be carried on the vehicle at the expense of payload.

The life of the casting over the weldment has been found to be at least 45 per cent longer through the endurance tests on the Canadian Army proving grounds in Ottawa.

This part is an example of how the properties of ductile iron can be combined with advanced molding techniques to succeed in a cost reduction coupled with better overall performance.

The quality of castings already produced and on order justify the statement made previously that castings lend themselves perfectly to long production runs with consistent properties.

Case 2 — Parts for a Portable Drill and Pump Unit

This complete unit (Fig. 6) is manufactured by a geophysical research company. It was developed for use in inaccessible locations found in the jungles or mountainous regions of South America. Weight and reliability are the paramount design requirements, since this equipment is transported on the backs of native porters, llamas or by helicopter.

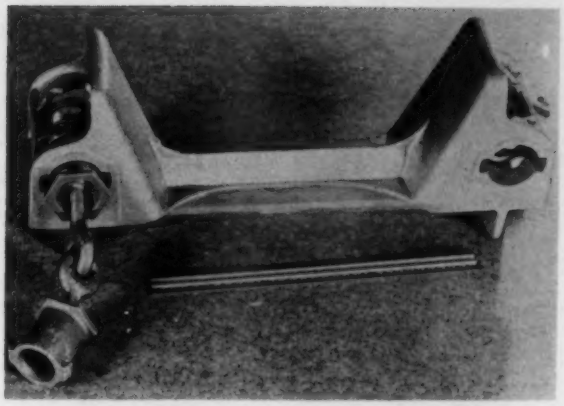


Fig. 5 — Single track shoe casting showing installation of parts in Fig. 4.

The original design was fabricated as a weldment, shown on Fig. 7 (pump section only). A conversion later involved the following parts as shown on Figs. 8 and 9:

1) Pump frame.
2) Cylinder heads (one per unit).
3) Cylinder (one suction and one discharge per unit).
4) Connection rod (two per unit).
5) Drill rotary head casting (Fig. 9).

The advantages of this conversion are numerous:

a) Better appearance of parts.
b) Faster production.
c) Pressure tightness and high strength-to-weight ratio.
d) Lower production cost. 30 per cent saved in machining cost, 20 per cent in fabrication cost.
e) Better corrosion resistance, particularly important in tough service conditions.
f) Better wear and erosion resistance.
g) Less vibration.

Case 3 — Dry Manifold for Gas Engine

This part was made by a company for its own use, weight approximately 460 lb. Previously made in low carbon steel weldment, the cost of the completely

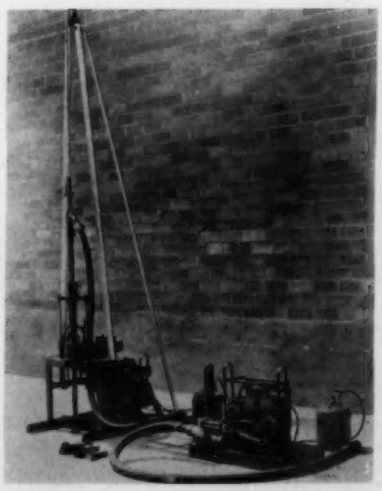


Fig. 6 — Portable drill and pump unit, parts of which were made of ductile iron castings (Figs. 8 and 9).

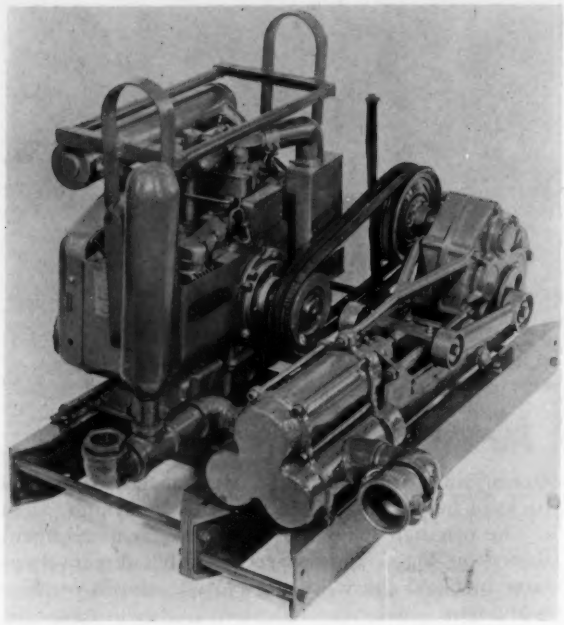


Fig. 7 — Original design of pump and drill unit (Fig. 6) made as a fabricated weldment (pump section only).

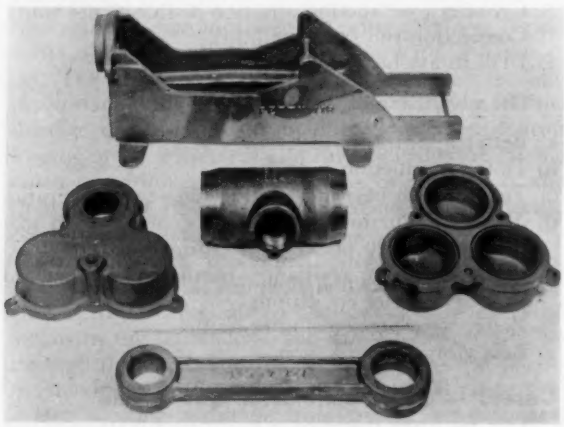


Fig. 8 — Pump frame, cylinder heads, cylinder and connecting rod of unit (Fig. 6) which were converted to ductile iron castings.

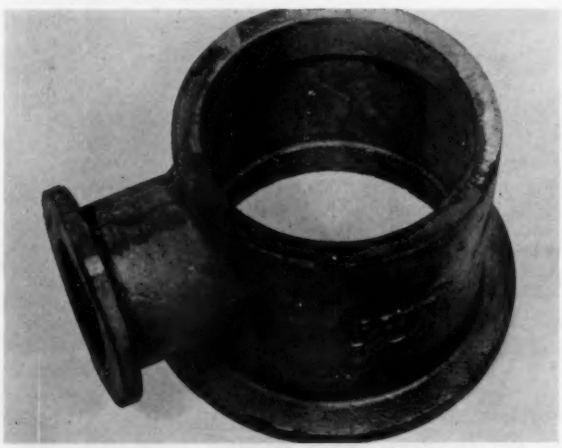


Fig. 9 — Drill rotary head casting made of ductile iron for same unit (Fig. 6).

finished part, machined, was $540.00. When converted to a ductile iron casting (Fig. 10), the finished cost dropped to $287.00, or a reduction of 47 per cent. Besides this cost reduction, the use of casting permitted a quicker delivery of the part than when made by welding.

Case 4 — 12 in. Diameter Sprocket

This sprocket (Fig. 11) is used on a combine that pulls a rake for harvesting grains. This sprocket is operated at relatively slow speed, from 300 to 400 rpm, and is bolted to a steel plate for support. Made originally from a ½-in. thick plate, the teeth were machined. It was later changed to a grade 80-60-03 ductile iron casting, with teeth accurately cast.

The total cost of an aluminum pattern and follow board amounted to $90.00. The sprocket cost was $7.00 when made from steelplate, $1.91 when cast. The total saving to the customer on approximately 2,000 sprockets produced thus far amounts to $10,180.00.

Case 5 — Ejection Cam

This cam was originally made from a 2½-in. thick steel plate, finish weight 427 lb. The machining time consisting of milling, drilling and fitting amounted to a total of 34.3 hr. The total cost of the

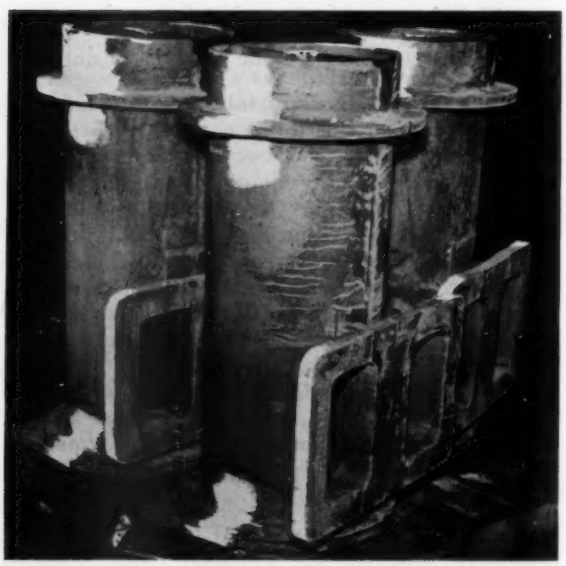


Fig. 10 — Dry manifold gas engine casting, cost of which was reduced 47 per cent.

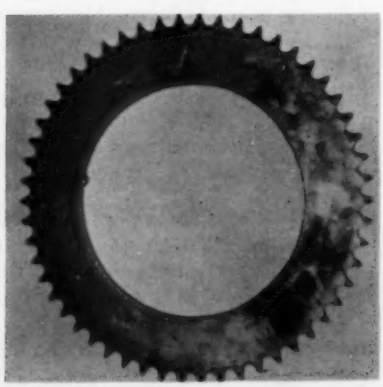


Fig. 11 — Diameter sprocket used on a combine that pulls a rake for harvesting grain.

part including material, machining, flame hardened $\frac{1}{16}$-in. deep to 52-58 Rockwell C, was $161.97.

This cam was redesigned as a ductile iron casting (grade 80-60-03), and the weight reduced to 151 lb through judicious use of ribs. The machining time was reduced to 21.3 hr, a time reduction of 38 per cent. The total casting cost, including flame hardening to 52-58 Rockwell C was reduced to $107.05; a saving of 34 per cent. The press operation is reported to be smoother through lowering the cam weight and the better vibration absorbing properties of ductile iron.

Case 6 — Frame for Paper Press

This manufactured frame was originally made with welded steel plates, as shown on Fig. 12. By converting the same to a high strength (100,000 psi ultimate tensile minimum) ductile iron casting, the following was achieved (Fig. 13):

1) Cost saving 25.8 per cent.
2) Ease of machining.
3) Improvement in strength.
4) Better appearance of the finished product.
5) Incorporation in the casting of such auxiliaries as supports, bosses, etc.

Case 7 — Cleat-hopper Track

These cleats are components of a track used to pick up parts in a hopper bin and deliver them to a loading track. Each cleat is approximately 6½-in. long, and was originally made from two pieces (1¾-in. x ½-in. and 1¾-in. x ¼-in. of steel bar stock welded together, the cross-section being an "L" shape. The breakdown of the operations necessary to complete the steel welded design follows:

1) Saw stock to length.
2) Mill ends to 30 degree angle.
3) Mill weld relief for welding to the track.
4) Weld 2 pieces together.
5) Mill three holding welds.
6) Sand blast to remove scale.
7) Heat treat, case harden to 58-60 Rockwell C.
8) Weld to hopper track.

This part was converted to a ductile iron casting (grade 70-50-15) with an as-cast Brinnell of 220. All machining operations were eliminated and the production steps reduced to three:

1) Casting the part.
2) Blast cleaning.
3) Welding to hopper track.

The cost of the weldment was $3.00 per unit against $1.34 for the casting, including pattern cost.

Case 8 — Adjustable Beam Locking Device

This small part is used to hold cross removable beams in special railroad gondola cars to haul steel strip coils. They were originally made by bending three pieces of steel plate, welding them in five separate places and drilling two holes. The weight was 14 lb (Fig. 14).

This part was redesigned and cast in grade 60-45-10 ductile iron. The total weight was reduced to 13 lb (Fig. 15). The castings performed satisfactorily and the cost was reduced from $3.64 as made in the owner's shop, to $3.15 made in another foundry. The schedule of deliveries could be met easily by the foundry in one day, while fabrication of the same quantity of parts required several days.

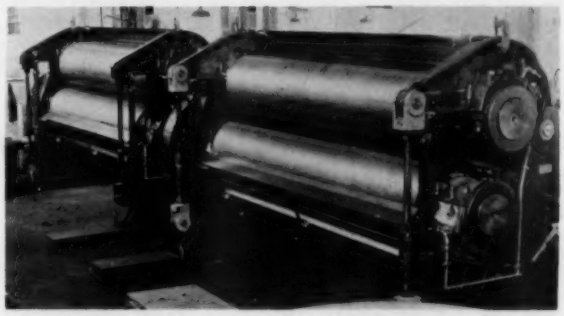


Fig. 12 — Paper press frame originally made of welded steel plates.

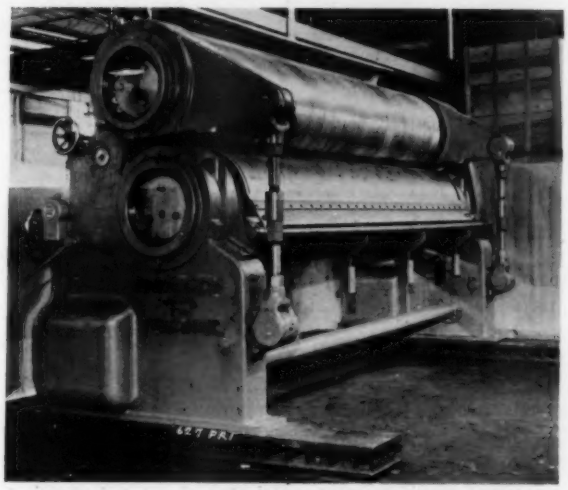


Fig. 13 — Same press frame shown in Fig. 12 made of high strength (100,000 psi UTS) ductile iron.

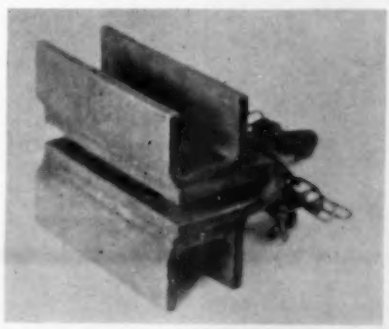


Fig. 14 — Adjustable beam walking device made by bending three pieces of steel plate, welding them in five separate pieces and drilling two holes. Weight, 14 lb.

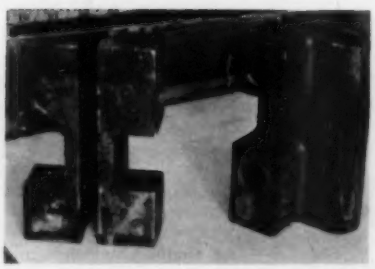


Fig. 15 — Redesigned part shown in Fig. 14 cast in ductile iron. Weight, 13 lb.

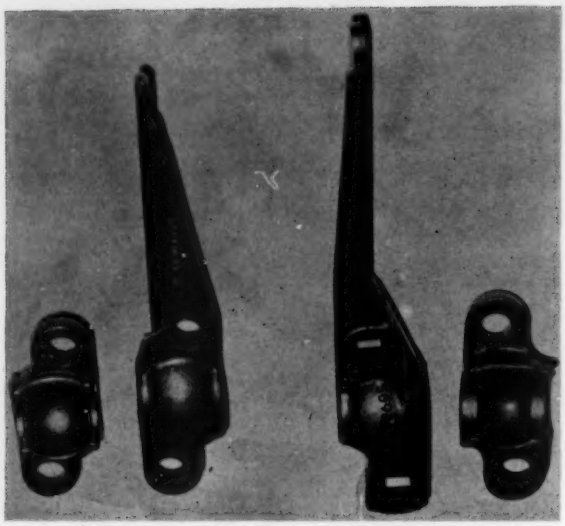


Fig. 16 — Knife heads used on reciprocating cutting mechanisms for grass and grain cutting machines. Left, heat treated forgings. Right, redesigned part made in ductile iron, heat treated for good wear resistance.

Case 9 — Knife Heads

This knife head assembly is used on reciprocating cutting mechanisms for grass and grain cutting machines (Fig. 16). For years these heads were made as forgings and heat treated. They were redesigned to be cast in ductile iron and subsequently heat treated for good wear resistance. The performance of the castings has been excellent, and the cost of the unit made up with castings is less than when made up of forgings.

This case illustrates a condition which is to be currently found. It involves a change in design which justifies the shift from one method of fabrication to another. Once a part is established as a forging, it would hardly be economical to discard the serviceable die equipment.



Fig. 17 — Governor weight weighing approximately ½ lb.



Fig. 18 — Gas engine crankshaft made in ductile iron.

Case 10 — Governor Weight

This part (Fig. 17) weighing approximately ½-lb was previously made as a forging of about the same weight. It was found more economical when redesigned to produce it as a ductile iron casting hardened for better wear resistance. The service performance was also found to be superior to the forging.

Case 11 — Gas Engine Crankshaft

This crankshaft (Fig. 18) is incorporated in a completely new 200 hp-475 rpm gas engine. This new engine can utilize the standard power take-off unit or be connected directly to compressors, pumps or other oil field equipment. Because of the performance of ductile iron cast shafts in older engines, and its bearing properties and damping capacity, ductile iron was selected for this new engine. The pattern cost was far less than the cost of dies to forge this shaft from steel, thereby resulting in a considerable cost reduction.

Case 12 — Electric Cable Pipe Roller Supports

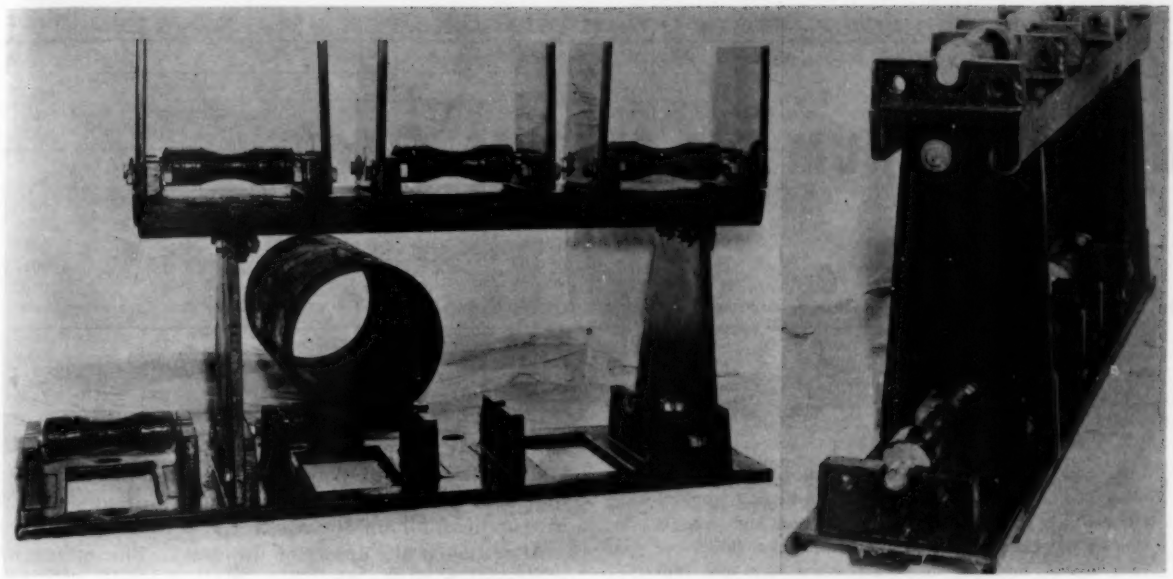
Although this example is not a true conversion, it is felt that it should be included as part of this paper for it is the result of a fabrication method study during the design stage. The company was faced with the problem of installing special new electrical conduits (Figs. 19, 20) in a practically inaccessible location of the Queensboro Bridge. The conduit supports were to possess a good corrosion resistance as well as high strength to resist the various stresses induced by the bridge movements. A cost study indicated that a ductile iron casting was not only more economical than a welded steel assembly, but also answered all the requirements of this installation. The complete casting including rollers were made of grade 60-45-10 ductile iron annealed under an extremely rigid quality control. A preliminary radiographic inspection of the castings was made to determine the gating system which would result in a no-residual stress as-cast condition.

Subsequently, a radiographic inspection was required on at least one casting of each of the first, the middle and the last ladle of the day's melt. Tension tests specimens from 1 in. keel blocks were also taken from the same ladle and heat treated with the castings before being machined and tested.

This case, the author's company believes, is an example showing that a ductile iron casting can be produced in large quantity (close to 1,000 complete supports) with a near-perfect metal structure.

Case 13 — Coupling Bar

This part serves as a connection between the tractor and an implement (Fig. 21). The forging was welded to a support member which is bolted to a tractor. It was redesigned as a ductile iron casting to combine two components into one, thus eliminating one assembly operation and permitting incorporation of cored holes in this new design. This redesign resulted in a lower composite unit cost with equal performance.



Figs. 19 and 20 (left and right) — Two views of ductile iron electric cable pipe roller supports installed on Queensboro bridge.

ACKNOWLEDGMENTS

The writer wishes to thank the staff of The International Nickel Co., Ductile Iron Div., particularly its director, Mr. D. J. Reese, for his continuous help and encouragement; Mr. K. D. Millis of the D&R Div., International Nickel Co., for his personal assistance during the preparation of this paper.

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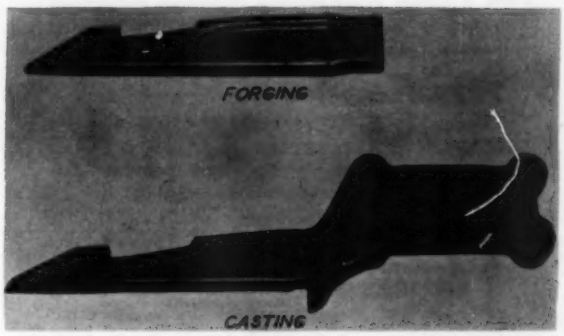


Fig. 21 — Coupling bar used as a connection between the tractor and an implement, redesigned from forging (top) to ductile iron casting (bottom).

NEW ALUMINUM DIE-CASTING ALLOY

By J. H. Moorman and E. V. Blackmun

ABSTRACT

A new aluminum die-casting alloy,* designated alloy 364, has been developed for structural applications which require tensile strength and high elongation or toughness. The nominal composition is 8.5 per cent silicon, 0.30 per cent magnesium, 0.30 per cent chromium and 0.03 per cent beryllium. In the as-cast condition, die castings of alloy 364 provide a combination of tensile properties and elongation comparable to those of alloy 218, and are suitable for structural applications requiring toughness.

The artificially aged (-T5) temper provides good elongation and a high yield strength. Annealed alloy 364 can be joined to other parts by rolling, spinning or riveting without cracking.

The casting characteristics, physical properties, corrosion resistance and machinability of alloy 364 are similar to other Al-Si die-casting alloys, such as alloy SG100B (commercial designation 360). The castability is good, permitting the die casting of thin walled or complicated parts which are difficult to cast in alloy G8A (commercial designation 218). Its use in structural applications or parts which require drastic forming can increase the applications for die castings.

INTRODUCTION

Aluminum die-casting alloys provide various combinations of mechanical or physical properties and castability to fit engineering design, casting requirements and cost. When maximum ductility and strength are required the binary aluminum–8 per cent magnesium alloy G8A (commercial 218) is suitable, except where its relatively poor castability causes poor quality castings, casting problems and higher costs. In spite of its superior properties, machinability and corrosion resistance, the use of alloy G8A die castings is limited to simple shapes.

The new die-casting alloy 364 is the result of investigations to develop an alloy which has the excellent mechanical properties of alloy G8A and better castability. Another objective is to provide an alloy which will undergo an increase in tensile properties and retain good elongation when subjected to a low temperature thermal aging treatment. In view of the known, desirable properties and castability of aluminum-silicon alloys, these alloys were used as a base in searching for an improved alloy.

Alloy 364 is the result of this work. The nominal composition of the new alloy is 8.5 per cent silicon, 0.30 per cent magnesium, 0.30 per cent chromium and 0.03 per cent beryllium. The common impurities, copper, iron, manganese, zinc, nickel and tin are controlled at the normal levels used in commercial die-casting alloys.

PROCEDURE FOR CASTING AND TESTING

In preparing melts or heats, alloying and melting practices are similar to those employed with alloy SG100B (commercial 360). The chromium and beryllium contents require slightly higher melting and holding temperatures, to obtain rapid melting and to prevent precipitation of the highest melting point compounds in the alloy. Metal holding and casting temperatures are 1200 to 1225 F. Frequent stirring will prevent precipitation at local cold spots in the furnaces.

Standard A.S.T.M. die-cast specimens were used to determine mechanical and physical properties. Properties of die castings from production dies were measured with static breakdown, impact and bending tests.

An artificial aging treatment consisting of 10 hr at 340 F is used to produce the high yield strength 364-T5 temper. Annealing for 2 hr at 650 F produces

TABLE 1 — TYPICAL MECHANICAL PROPERTIES[1] OF DIE-CASTING ALLOYS

Commercial Alloy and Temper	A.S.T.M. Designation	Tensile Str., ksi	Yield Str. (offset —0.2% ksi	Elong., % in 2 in.	Endurance Limit,[2] ksi	Shear Str., ksi
364-F	—	43	23	7.5	18	26
364-T5[3]	—	46	33	4.0	—	—
364 annealed[4]	—	27	13	11.0	—	—
218-F	G8A	45	27	8.0	20	29
360-F	SG100B	47	25	3.0	19	30
380-F	SC84B	48	24	3.0	21	31

1. Tensile properties are average values obtained with A.S.T.M. standard ¼-in. diameter round specimens produced on a cold chamber machine.
2. R. R. Moore type specimen, 500,000,000 cycles.
3. Artificially aged 10 hr at 340 F.
4. Annealed 2 hr at 650 F.

*Covered by U.S. Patent 2,823,995 issued to Aluminum Co. of America.

J. H. MOORMAN is Asst. Chief, Chicago Section, Alcoa Rsch. Labs., Aluminum Co. of America, Chicago, and E. V. BLACKMUN is Asst. Chief Met., Fabricating Div., Aluminum Co. of America, Pittsburgh.

TABLE 2 — MECHANICAL PROPERTIES OF ALLOY 364 AFTER AGING AT VARIOUS TEMPERATURES

Aging Time, Hr	Room Temperature Properties[1] After Aging at: 150 F Tensile Str., ksi	150 F Yield Str., ksi	150 F Elong., % in 2 in.	250 F Tensile Str., ksi	250 F Yield Str., ksi	250 F Elong., % in 2 in.	300 F Tensile Str., ksi	300 F Yield Str., ksi	300 F Elong., % in 2 in.
½	40.5	21.6	5.4	41.7	22.9	6.0	39.7	23.1	5.1
3	39.7	22.2	4.5	41.7	23.0	6.8	41.9	26.7	4.8
24	41.5	23.2	5.8	42.4	25.3	5.0	49.5	36.0	4.6
72	41.9	22.8	7.0	44.2	29.5	4.1	48.1	35.7	3.4
360	43.0	23.4	7.3	49.3	35.6	4.6	44.9	30.5	4.8
2400	45.4	26.9	7.9	46.9	34.3	4.5	42.7	28.2	5.6
10,000	46.0	27.7	6.0	45.4	28.8	6.5	41.7	26.1	5.6

1. Averages of four tests on standard A.S.T.M. ¼-in. diameter cold chamber die-cast test bars.

the high ductility required for joining the die castings to other parts by rolling, riveting or spinning.

MECHANICAL AND PHYSICAL PROPERTIES

The mechanical properties of alloy 364 are of particular interest when the yield strength and elongation are compared to alloys G8A, SG100B and SC84B. A comparison of typical mechanical properties is shown in Table 1. Alloy 364-F die castings provide high tensile properties and the excellent toughness or ductility, shown by a typical elongation of 7.5 per cent. Artificial aging to the -T5 temper increases yield strength 10 ksi while retaining a desirable degree of toughness at 4.0 per cent elongation. In the annealed condition alloy 364 can be joined to other parts by rolling, spinning or riveting operations without cracking the die castings.

In view of the use of aluminum alloy die castings under conditions of intermittent or continuous exposure to moderately elevated temperatures, the effect of aging up to 10,000 hr at 150, 250 and 300 F was determined with standard A.S.T.M. test bars. The room temperature mechanical properties are shown in Table 2 after aging for the various periods of time.

The physical properties and other characteristics of alloy 364 are listed in Table 3.

CASTING CHARACTERISTICS AND HEAT TREATMENT

Die castings of alloy 364 are made commercially using production dies and machines. Castability is rated as good, being comparable to that of alloy SG100B and superior to alloy G8A. Alloy 364 die castings are used in the as-cast (-F), artificially aged (-T5) and annealed tempers. As shown in Tables 1 and 2, various combinations of mechanical properties can be obtained with these tempers.

APPLICATIONS

Alloy 364-F is particularly suitable for die-cast parts which require high tensile strength and ductility or impact resistance. The static load and impact test results in Table 4 show the excellent strength and ductility of 364-F alloy die castings. In one series of tests, a lever arm part was tested to failure in static load and impact testing machines. The static test setup is shown in Fig. 1. This part is subject to high bending and impact loads in service. The test results and service of this part show the superior ductility of alloy 364.

The second series of tests in Table 4 compare

TABLE 3 — PHYSICAL PROPERTIES AND CHARACTERISTICS OF ALLOY 364

Specific gravity	2.63
Weight/cu in.	0.095
Electrical conductivity, per cent I.A.C.S., -F Temper	30
Thermal conductivity at 25 C, C.G.S. units, -F Temper	0.29
Average coefficient of thermal expansion/deg. Fahr.	
Range -58 to 68 F	10.8
Range 68 to 212 F	11.6
Range 68 to 392 F	12.2
Range 68 to 572 F	12.7
Approximate melting range	1035-1115 F
Resistance to hot cracking	Excellent
Pressure tightness	Good
Die filling capacity	Excellent
Resistance to corrosion	Good
Machinability	Good

TABLE 4 — STATIC LOAD AND IMPACT PROPERTIES OF ALLOY 364 DIE CASTINGS

Part	Alloy Designation: Commercial	Alloy Designation: A.S.T.M.	Static Breakdown Test: Load at Failure —lb	Static Breakdown Test: Deflection at Failure —in.	Impact Test: No. of Blows to Failure
Lever Arm	364-F	—	317	1.02	18
	218-F	G8A	308	1.03	10
I-Beam Bracket	364-F	—	3600	0.217	18
	364-T5	—	4250	0.081	10
	218-F	G8A	4150	0.219	15
	360-F	SG100B	3580	—	6
	380-F	SC84B	3610	—	5

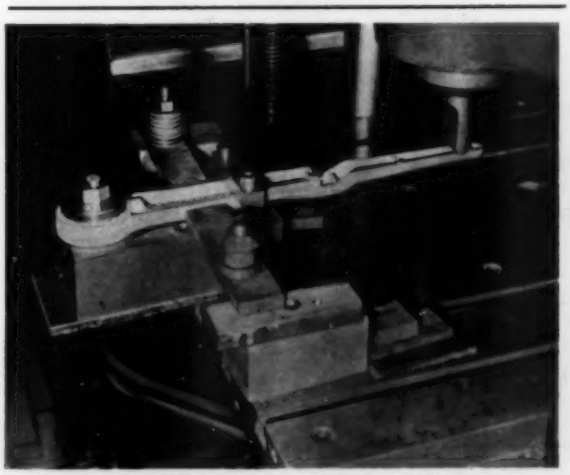


Fig. 1 — Breakdown testing of lever arm part made of alloy 364.

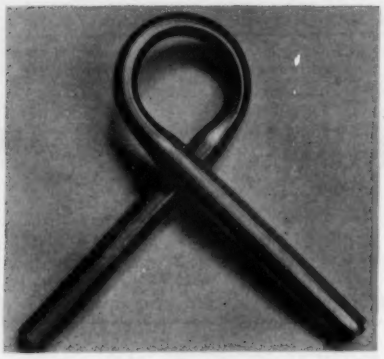


Fig. 2 — Bending deformation of alloy 364 showing extent to which the aluminum alloy can be deformed.

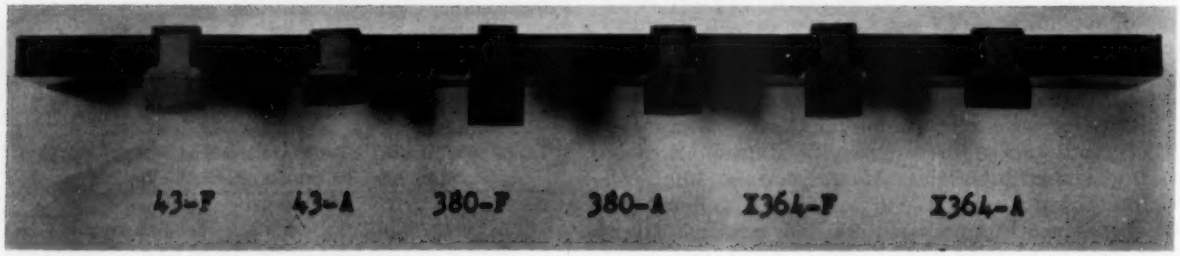
static breakdown and impact test results on the I-beam section of a structural bracket, illustrating the excellent properties of alloy 364. In both cases, the parts are produced in alloy 364. Although alloy G8A die castings show similar test results, high production scrap resulted from shrink cracks and poor castability.

For applications which require assembly of die castings by rolling, spinning or riveting, annealed alloy 364 is of considerable interest. Tensile properties in Table I and bent test bar in Fig. 2 show the extent to which alloy 364 annealed 2 hr at 650 F can be deformed. A standard test bar has been bent approximately 360 degrees around a radius equal to two times the diameter. Similar bends have been made over a radius of 1½ diameters. Another forming application of alloy 364 is illustrated by Fig. 3.

All of the rivets shown were machined from die-cast tensile test bars and cold headed between flat steel dies with a force of 4500 lb/rivet. Although the three alloys formed satisfactorily in the test, forming alloy SC84B caused cracks to develop in production castings, and poor castability limits alloy S5C to simple parts. Alloy 364 die castings are rolled, spun or riveted with satisfactory production results.

Alloy 364-F is recommended for die castings requiring high strength and ductility of impact resistance. In the -T5 aged temper, the alloy provides high tensile strength, good elongation and high yield strength.

Fig. 3 — Rivets formed from grip ends of aluminum alloy die casting test bars. All rivets were machined from die-cast test bars and cold headed between flat steel dies with a force of 4500 lb/rivet.



SOME PRINCIPLES FOR PRODUCING SOUND Al-7Mg ALLOY CASTINGS

By W. H. Johnson and J. G. Kura

ABSTRACT

Principles were developed for the production of sound horizontal castings of the Al-7Mg alloy. They were developed from a study of the relationship that exists between the thermal gradients at the center line of the casting and various conditions such as the size of the casting as well as the location of gates, risers and chills. Numerous embedded thermocouples established the thermal history during solidification of the castings. The experimental work was performed on elementary shapes — plates and bars.

It was found that the feeding range of a side riser decreased as the section thickness increased. Unsoundness occurred when the thermal gradients fell below 5 F per in. The feeding range of an end riser was increased by 50 per cent or more by using wedge-shaped chills on the drag surface of the casting. Chills placed at the end of the casting opposite the riser did not increase the feeding range of the riser. Marked improvement in tensile properties were obtained on plates that were cast with wedge-shaped chills, as compared with similar unchilled plates that were also sound.

Adverse temperature gradients were created by a top riser on horizontal plate castings which were 2 in. thick. As a result, shrinkage occurred in the vicinity of the riser. An insulating sleeve on a top riser increased the degree of soundness somewhat, but did not increase the feeding range of the riser. A marked improvement in soundness was obtained, however, when multiple gates were used with a top riser.

INTRODUCTION

The solidification of castings has been studied by numerous investigators.[1-7] Much has been learned concerning the solidification characteristics of both ferrous and nonferrous alloys.[1] The maximum distance that risers will feed the solidification shrinkage in ferrous castings has been determined and, as a result, formulas for the correct dimensioning of risers to be used on certain ferrous castings[2,3] have been evolved. However, complete determination of feeding ranges in nonferrous casting alloys has not been made. At best, the size of nonferrous risers can only be approximated from the charts and formulas available for certain ferrous alloys.

Therefore, it was considered desirable to develop principles governing the location of gates, risers and chills on nonferrous castings. Such information should result in improved casting soundness.

In the past, procedures were worked out empirically when castings of a new design were first produced. In recent years, however, practical methods have been evolved for estimating in advance the correct dimensions of risers to feed many types of castings of low-carbon grades of steel,[2,3] ductile iron[4] and some grades of stainless and heat-resistant steels.[5] Similar research on the risering of nonferrous alloys and ferrous alloys, other than those that were mentioned, has been quite limited to date.

To determine the principles of thermal dynamics that produce sound nonferrous castings for armor, research was sponsored at Battelle Memorial Institute by the Ordnance Tank Automotive Command, Detroit Arsenal, Center Line, Mich., under Contract No. DA-33-ORD-2434, and was guided by the Research Committee of the Light Metals Div., American Foundrymen's Society.* The Al-7Mg alloy was selected as representative of a typical light nonferrous alloy for armor purposes. It was well suited to the investigation because it is known to produce gross shrinkage when inadequately fed. In addition, the alloy has good welding characteristics which would make it desirable for final adoption by the Ordnance Corps as nonferrous armor plate.

Thermal Dynamics Investigation

Essentially, complex castings are combinations of plate and bar shapes of different sizes. Therefore, one series of experiments was designed to investigate the thermal dynamics of plate castings; the other series covered bar castings. The term "thermal dynamics" as used here refers to the study of how thermal gradients in a casting originate, progress and are affected and controlled by changes in the location of the gates, risers, and chills.

The effects of the following variables on the thermal gradients within these castings during solidification were investigated:

W. H. JOHNSON is Senior Met. and J. G. KURA is Division Chief, Battelle Memorial Institute, Columbus, Ohio.

*The Research Committee of the Light Metals Div., AFS, consisted of the following members: W. E. Sicha, *Chairman;* W. Bonsack, D. J. Henry, W. J. Klayer, J. G. Mezoff, C. E. Nelson, T. D. Stay, and R. F. Thomson. G. P. Messenger, who represented Detroit Arsenal, S. Lipson, who represented Frankford Arsenal, and R. W. Zillman and C. B. Jenni, all served as ex officio members of the committee.

1) Riser position.
   a) Side riser.
   b) Top riser.
2) Single and multiple gates.
3) Chill location.
4) Gate location.

The experiments were designed to study the effect that only one riser has on the soundness of the casting. In commercial foundry practice, multiple risers would be used on castings larger than the ones that were studied. Where the large commercial casting can be broken down to multiple units of the experimental castings, it becomes apparent that combining multiples of the simple castings would constitute the commercial casting. Thus, the principles of thermal dynamics of casting the simple unit shapes could be applied intelligently to a large complex shape.

The present investigation is restricted to a study of individual horizontal plates and bars. For a precise application of principles of thermal dynamics to complex shapes, it is necessary to have knowledge of the interaction that exists during solidification between different section thicknesses. Knowledge of the effect of different orientations of adjoining sections on the thermal dynamics is also needed. These are areas for future investigation.

The thermocouple technique[1] was chosen as the most suitable for studying the thermal dynamics of casting. Briefly, the technique consists of obtaining cooling curves from thermocouples placed at strategic locations in a solidifying casting, and establishing the complete history of movement of the freezing front. A strip-chart temperature recorder was used which is capable of recording temperature readings from 16 thermocouples within 30 sec.

GENERAL CONSIDERATIONS

In the past, the gating and risering procedure to produce sound castings of a new design was developed by trial and error. Later cumbersome mathematical methods, based on empirical relationships were employed for ferrous castings. As a result of recent investigations,[3] however, these methods[6,7] are no longer necessary. Now the proper riser dimension is readily obtained through the use of a simple shape factor. This shape factor is

$$\frac{L + W}{T},$$

where

L = length of casting, in.
W = width of casting, in.
T = thickness of casting, in.

The original data obtained by Bishop, et al.,[3] in establishing the empirical relationship between the shape factor and the ratio of riser volume to casting volume for low-carbon steel are shown by the dotted area in Fig. 1. Trend lines for various ductile irons,[4] and for some highly alloyed steels,[5] are also shown in Fig. 1. The straight-line relationships between the shape factor and the R_v/C_v ratio for these ferrous alloys establish the boundaries between sound and

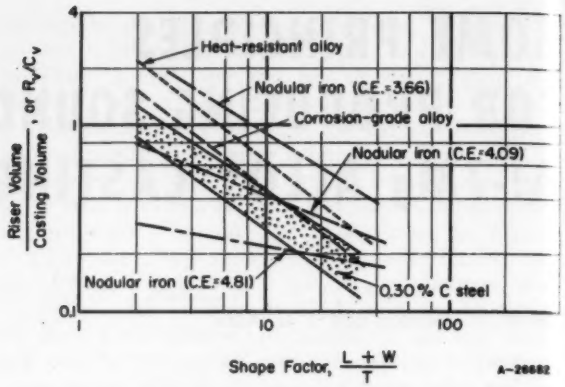


Fig. 1 — Relationship between known shape factors and R_v/C_v ratios.

unsound castings. Castings with shape factors and R_v/C_v ratios above the trend lines would be sound. For example, to make a sound casting from ductile iron with a carbon equivalent of 3.66, the minimum R_v/C_v ratio would be 1 for a shape factor of 8.

To date, the correct dimensions of risers that will feed a casting adequately have been established for ferrous alloys only. Of particular interest is the fact that a straight-line relationship exists between the shape factor and the R_v/C_v ratio for each alloy studied.

Ferrous Alloy Castings

The data contained in Fig. 1 are based upon the fact that the feeding range of risers used with ferrous alloys has been found to be $4\frac{1}{2}T$ for plates, and $6\sqrt{T}$ for bars, where T is the thickness in in. This means that when making ferrous alloy castings of simple design the feeding distance increases as the section thickness increases. However, research on the risering and feeding ranges of nonferrous alloys has been quite limited. Therefore, when this research was started, it was not known whether this relationship applied to nonferrous alloys.

The straight-line relationship between the shape factor and the R_v/C_v ratio could be assumed to be acceptable if the feeding range of the Al-7Mg alloy were to increase as the section thickness increased. Since no data were available to specify the size of riser that is required to feed the Al-7Mg alloy, the volume of the risers used in this investigation was determined from the data in Fig. 1 for the 0.30 per cent carbon steel. The size of the riser was selected as if it were intended to feed steel plates ½, 1 or 2 in. thick having a length and width of 5T. In the course of the investigations, the length of the Al-7Mg alloy plates was varied to determine the actual feeding range of the riser.

In previous research,[8,9] improved horizontal gating systems were developed that would deliver clean metal to the mold cavity with a minimum amount of turbulence. The gating system used in the present research was a simplified version of the ideal system. The simplified version was used because 1) it was easily constructed, and 2) the objective of the investigation was to study the principles affecting soundness rather than cleanliness.

TABLE 1 — PLATE, CUBE, BAR AND RISER DIMENSIONS FOR Al-7Mg ALLOY CASTINGS

Plate Dimensions, in.					Riser Dimensions, in.	
Thickness, T	Width, 5T	Length 3T	Length 5T	Length 8T	Diameter 2½T	Height Above Plate, H
½	2½	1½	2½	4	1¼	5
1	5	3	5	8	2½	5
2	10	6	10	16	5	5

Cube and Bar Dimensions, in.				Riser Dimensions, in.	
Thickness, T	Width, T	Length T	Length 4T	Diameter 1½T	Height Above Bar, H
2	2	2	8	3	5
3	3	3	12	4½	5
5	5	5	—	7½	5

EXPERIMENTAL PROCEDURE

The dimension of the plates, cubes, bars and risers that were used in the experiments are shown in Table 1. Three thicknesses of plates and cubes were used. For each thickness of plate, three different lengths were studied. Two thicknesses of bars were used with a length of 4T. The diameter of the riser used with the plates was 2½T. For the cubes and bars, the diameter of the riser was 1½T. The height of the riser when placed on the top of the casting was 5 in. When used as a side riser, the riser height was equal to the thickness of the casting plus 5 in. All of the plates, cubes and bars were cast horizontally. Additional information concerning the location of the gates, risers and chills is presented in later portions of the paper devoted to the analysis of the thermal history of the castings.

Heats of Al-7Mg alloy were prepared in a clay-graphite crucible in a high-frequency induction furnace. The melts were heated to 1400 F and degassed with chlorine so that the shrinkage voids in the cast plates would not be influenced by variable gas content in the melts. About ¼-lb of chlorine was bubbled through the 50-lb melts in the 5 min degassing period. Immediately following the degassing operation, a sample was dipped out of the furnace and allowed to solidify under a vacuum at a pressure of about 75 mm.

If no evolution of gas occurred during its solidification, 0.005 per cent beryllium was added to the melt as a 2.5 per cent beryllium-aluminum master alloy to protect the melt against oxidation during pouring. Grain refining elements were intentionally not added to the melt. The melt was then transferred to a preheated ladle and poured at 1350 F. Commercial grade aluminum (99.85 per cent purity) and magnesium (99.8 per cent purity) were used to prepare the melts. The average iron and silicon content of the melts was 0.07 and 0.06 per cent, respectively.

The molds were made by the cold-setting binder process. The sand mix used was as follows:

Silica sand, AFS fineness no. 65 — 100 lb.
Cold-set binder — 3 lb (2 per cent).
Cold-set activator — 0.3 lb (0.2 per cent).

The sand and binder were mulled together for 6 min. After the activator had been added, the mix was mulled for an additional 2 min. The permeability of the sand after mulling was between 52 and 56. After baking for 1 hr at 400 F, the molds had a hardness between 85 and 90.

Thermocouple and Radiographic Techniques

The thermocouple technique was chosen as the most suitable for studying the thermal dynamics of casting. To obtain the thermal history during solidification of the castings, thermocouples were embedded midway between the top and bottom faces of the plates and bars. As many as 16 thermocouples were placed at selected locations in a single casting.

Thermal studies, although informative, are quite time consuming to conduct. Some experiments, therefore, were conducted in which no thermal data were obtained during the solidification of the castings. The effect of changes in the location and dimension of the chills, risers and gates on soundness was then determined radiographically.

Radiographs were made of the whole casting to determine the presence of shrinkage voids and their location. Vertical sections were removed that extended from immediately in front of the riser to the end of the plate or bar. In addition, sections were made at other locations in the castings in which the presence of shrinkage voids was suspected, but not clearly indicated by radiographs of the whole casting. These sections, in turn, were radiographed so that the cross-section of the plate or bar could be examined more accurately for shrinkage voids.

Thermocouple Data

From the data obtained from the embedded thermocouples, curves were constructed of temperature versus distance from the riser at specific time periods after the castings were poured. The drawings in the middle horizontal row in Fig. 2 illustrate the type of plot. Similar plots were also made of the temperature distribution across the middle of the castings. From these graphs, the longitudinal and lateral thermal gradients were determined for the castings at their center line.

A thermal gradient is the difference in temperature that exists between two points divided by the distance between them. In this investigation, it is the difference in temperature between two points 1 in. apart at the moment that solidification starts at the point farther from the riser. Examples of plots of thermal gradients are shown in the upper horizontal row of Fig. 2. The thermal gradients were determined at the solidus temperature.

RISER POSITION EFFECT

Plates With Side Riser

A considerable amount of information can be obtained from the end-of-freeze waves on the effects that changes in the location of riser, chills and gates have on the solidification of castings. End-of-freeze isotherms were drawn at specific time intervals after pouring was completed. Positions of the end-of-freeze isotherms for the 1-in. thick plates of Al-7Mg are illustrated by the sketches in the bottom row in Fig. 2.

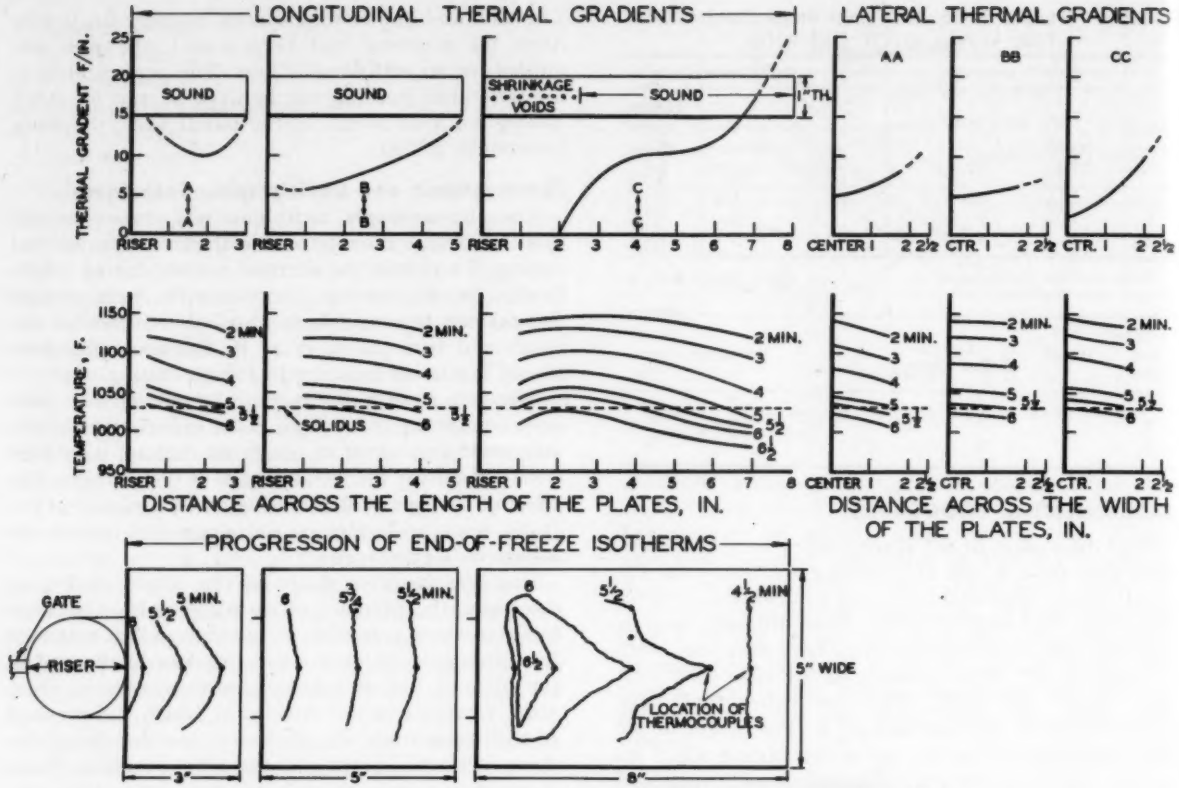


Fig. 2 — Longitudinal and lateral thermal gradients present at the solidus temperature (1030 F) during solidification of 1 in. thick Al-7Mg alloy plates.

The first group of experiments consisted of making the nine plate castings listed in Table 1. The data obtained from these nine plates were intended to establish the minimum thermal gradient required to produce castings free from shrinkage voids. The plates were made with a side riser and cast horizontally, as illustrated in Fig. 3.

The graphs in Fig. 2 illustrate the type of thermal history that was obtained. In this instance, the data are for plates 1 in. thick, 5 in. wide, and 3, 5 or 8 in. long, with a 6-in. riser that is 2½-in. in diameter.

The curves, in the upper horizontal row of Fig. 2, show that both the longitudinal and lateral thermal gradients decrease near the riser as the length of the plate increases. Shrinkage voids were not detected in the two shortest plates (3 and 5 in.). The lowest thermal gradients in these two plates were 5 F per in. Shrinkage voids were present in the longest plate (8 in.) in the region in which the analysis of the thermal history of the plate disclosed that no thermal gradient existed.

End-of-Freeze Isotherms

A study of the end-of-freeze isotherms reveals the movement of the freezing front. Solidification in the shortest plate was highly progressive toward the riser, and became less so as the length of the plate increased. In the longest plate, directional solidification stopped about 6 min after the casting was poured. By this time, solidification started to develop near the riser while liquid metal still existed at a point further distant from the riser. Consequently, when the trapped molten metal solidified within this portion of the plate, shrinkage voids developed. These data show that the limit of the feeding range of the riser for a 1-in. thick plate is about 5 in., or 5T.

The longitudinal and lateral thermal gradients for the ½- and 2-in. thick plates, as well as two special 1-in. thick plates, are shown in Fig. 4. No shrinkage voids occurred in the ½-in. thick plates because the thermal gradients were quite large. The longitudinal thermal gradients in the longest plate (4 in.) approached the minimum of 5 F/in. required for complete soundness, and established 4 in. (8T) as the limit of the feeding range of the riser for this thickness of plate.

Controlled directional solidification was not achieved in the two longest 2-in. thick plates (10 and 16 in.), and shrinkage voids were present through-

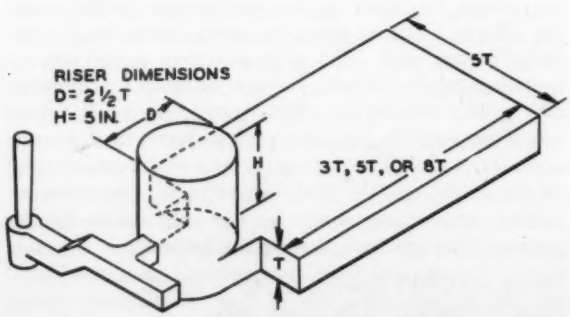


Fig. 3 — Location of gate and side riser for cast plates.

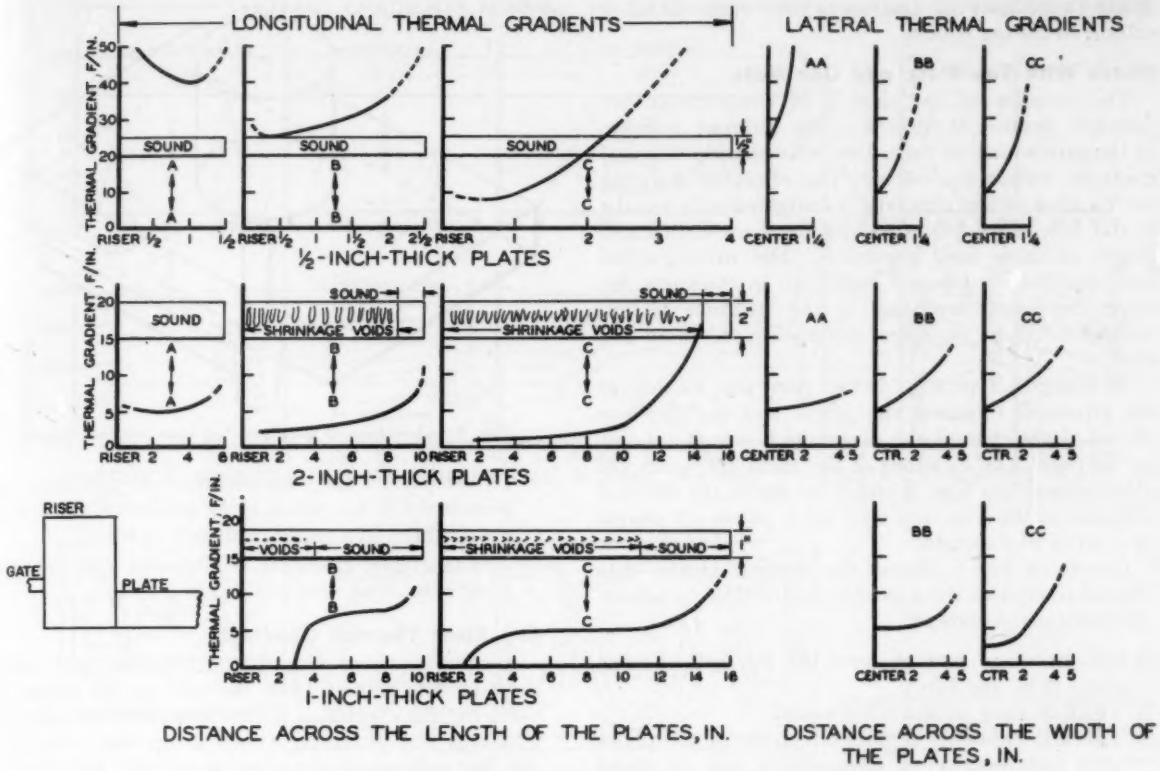


Fig. 4 — Longitudinal and lateral thermal gradients present at the solidus temperature (1030 F) during solidification of ½- and 2-in. thick Al-7Mg alloy plates.

out most of these two castings. Large thermal gradients existed near the farthest edge of the two plates because of the chill effect that the mold always creates on the edge of a casting. Farther from the edge of the plate, heat was lost uniformly, and low thermal gradients occurred across most of the plate. Therefore, directional solidification was not achieved, and shrinkage occurred.

Controlled directional solidification was barely sufficient to achieve soundness in the shortest plate (6 in.). These data establish that the limit of the feeding range of the riser for a 2-in. thick plate is 6 in., or 3T.

Thermal Gradients Effect

The correlation of the normal gradients at the center line with the presence of shrinkage in the nine plates shows that shrinkage voids will be present in plates of Al-7Mg when the thermal gradients fall below 5 to 8 F/in. In addition, it was concluded that the feeding range of a riser decreases as the section thickness of a plate increases. It is apparent that feeding of the aluminum-alloy plate castings as a function of thickness does not follow the general rule for steel castings of the same shape and having the same design of gates and risers.

The thermal gradients in the 2-in. thick plates were small. Furthermore, it was shown that the feeding range of a riser, in terms of *T*, was greater in thinner plates. Therefore, experiments were conducted to determine whether directional solidification could be increased in plates of the same dimensions as the two longer 2-in. thick plates, but with a reduction in thickness to 1 in. This change in thickness doubled the ratio of riser volume to casting volume.

Thermal Gradients vs. Shrinkage

The relationship between the thermal gradients and the presence of shrinkage in these particular 1-in. thick plates is shown in the lower row of Fig. 4. Although the thermal gradients in the center of these 1-in. plates were larger than in the companion 2-in. plates, the thermal gradients fell to zero in front of the riser, which was the portion of the plate that solidified last. A region in which no thermal gradients exist extended over a relatively small area in the center of the 10-in. long plate, and extended for a considerably greater distance in the 16-in. long plate. Radiographs revealed that shrinkage voids were present along the center line of both 1-in. plates. However, only a small amount of shrinkage was present in the 1-in. plate that was 10 in. long.

The presence of shrinkage at the center line shows the importance of the magnitude as well as the direction of the thermal gradients during solidification. When the thermal gradients are large, the beginning and end-of-freeze waves are close together, interdendritic channels are short and open and the flow of feed metal can take place without difficulty. When the thermal gradients are nonexistent or small, as in the 2-in. thick plates, the interdendritic channels may extend the length and width of the plate. Feeding under these conditions is difficult. Furthermore, the data show that controlled directional solidification cannot be achieved in the Al-7Mg alloy by the

simple expedient of increasing the ratio of riser volume to casting volume.

Plates With Top Riser and One Gate

The location of the riser is of considerable importance because it influences the ultimate position of the gate which, in turn, also influences the thermal gradients within the casting. The effect of changing the location of the riser was investigated only for the 1- and 2-in. thick plates having the same width and length as those used previously. The investigations were confined to plates 1 and 2 in. in thickness, because the work described in the previous section showed the ½-in. plate was sound when fed by an end riser.

As illustrated in Fig. 5, the riser was located at the geometric center of the plates, and the gate was placed at the edge of the plate which was the closest to the riser. The locations of the risers and gates are also indicated in Figs. 6 and 7, in which the thermal histories of the 1-in. and 2-in. thick plates are shown in a series of diagrams.

Graphs in Fig. 6 present the thermal history that was obtained with the 1-in. thick plate. The data from the three plates showed:

1) No shrinkage was observed in the two shortest plates (3 in. and 5 in.).
2) The 8-in. long plate was not sound.
3) Thermal gradients at the extremities of the plates were high because of edge effects, but decreased beneath the riser.

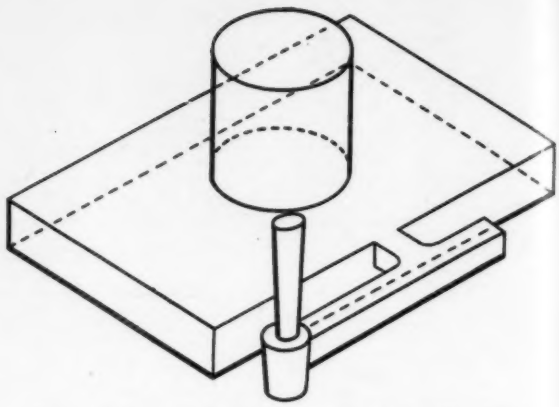


Fig. 5 — Location of gate and top riser for cast plates.

4) In the plate 5 in. long, the thermal gradients approached 5 F/in., which is the minimum value that would result in a sound casting.
5) The thermal gradients under the riser of the plate 8 in. long dropped below 5 F/in.

Top Riser Thermal Gradients

A comparison of Figs. 2 and 6 shows that, when the riser was moved from the end to the center of the 1-in. thick plate, a different pattern of thermal gradients was produced. When a top riser was used, the last portion of the plate to solidify was directly beneath the riser. The thermal gradients in this region

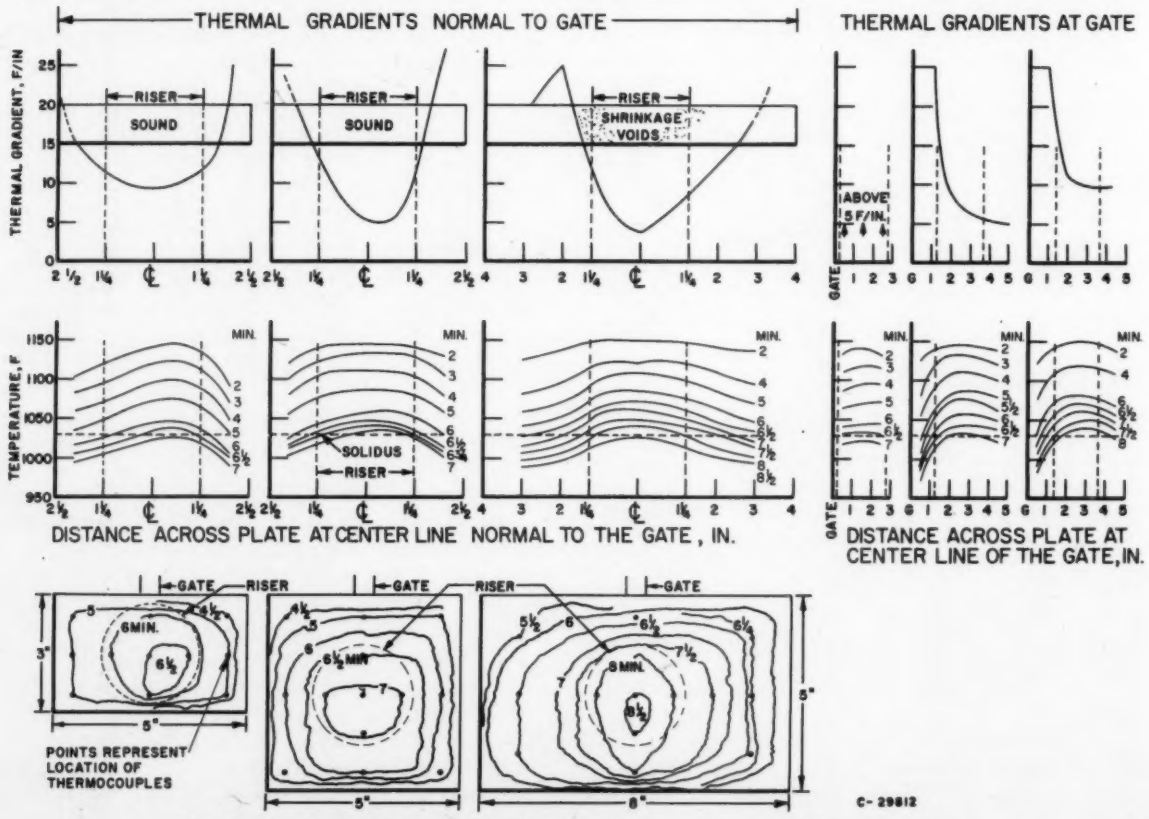


Fig. 6 — Longitudinal and lateral thermal gradients present at solidus temperature (1030 F) during solidification of 1 in. thick Al-7Mg plates fed by a center riser.

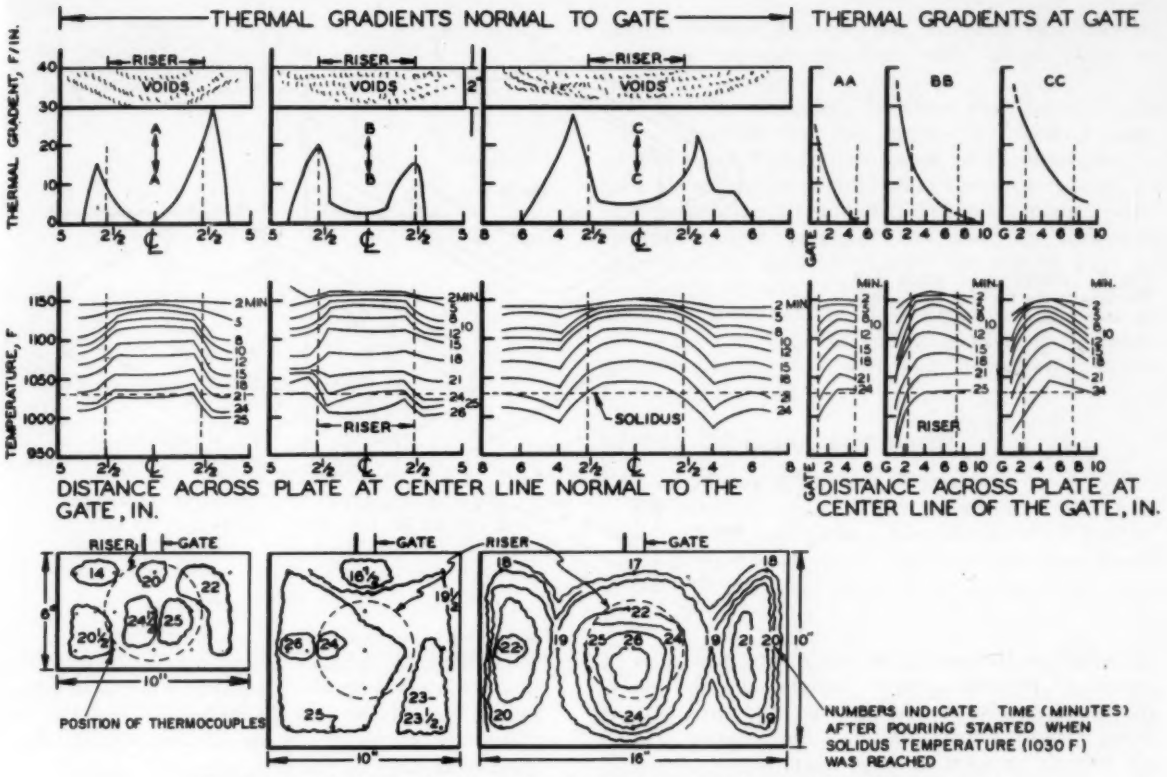


Fig. 7 — Longitudinal and lateral thermal gradients present at solidus temperature (1030 F) during solidification of 2 in. thick Al-7Mg plates fed by a center riser.

were at a minimum. Large thermal gradients, however, existed at the edges of the plate.

In the plate gated and fed at one edge, the thermal gradients were high at the edge opposite the riser, but were at a minimum immediately in front of the riser. With either the side riser or the center riser, the 3-in. long and 5-in. long plates were radiographically sound, and the 8-in. plates were not sound. Regardless of riser position, the thermal gradients in the 8-in. plates fell below the 5 F/in. required for soundness.

The thermal history of the 2-in. thick plates is given in Fig. 7. The date obtained from the three plates show the following:

1) Small thermal gradients exist in the center line of the plates directly beneath the riser and at the extremities at either side of the gate.
2) The largest thermal gradients occur in the region immediately adjacent to the riser.
3) Shrinkage voids are present throughout the plate, but are particularly severe under the riser.

In the absence of both large thermal gradients and controlled directional solidification, shrinkage voids developed throughout the plates. Thermal gradients were not so large during the late stages of solidification as during the earlier stages. If castings are to be sound, molten metal must be supplied by the riser to feed the solidification shrinkage of the last as well as the first portion to solidify. The end-of-freeze isotherms in Fig. 7 show that this was not attained with the 2-in. plates. In the longest 2-in. thick plate (16 in.), for example, the region immediately adjacent to the riser solidified before solidification was completed in all regions farther from the riser.

A comparison of Figs. 4 and 7 shows that, for the 2-in. thick plates, more favorable gradients were obtained with an end riser than with a center riser. The shortest 2-in. thick plate (6 in.) cast with a center riser was not sound. The same plate cast with an end riser, however, was sound. It is apparent that unless large thermal gradients can be produced and extended toward the riser, the riser actually may make the feeding of solidification shrinkage more difficult.

Jackson[10] found in a leaded gun-metal block casting that, as the top riser was increased from a size too small to yield directional solidification to a size larger than necessary for this purpose, porosity first decreased to a minimum and then increased. This increase in porosity with the larger top riser was attributed to thermal gradients that were reduced to values below those required to obtain maximum soundness.

The experiments with the 1- and 2-in. thick plates show that the effectiveness of a top riser decreases as the section thickness increases. Also, in 2-in. thick plates, a top riser has a shorter feeding range than a side riser.

Plates With a Top Riser and Multiple Gates

A single gate attached to the edge of a plate having a top riser creates a localized hot spot by prolonged heating of the sand in the immediate vicinity of the gate. Directional solidification is decreased and shrink-

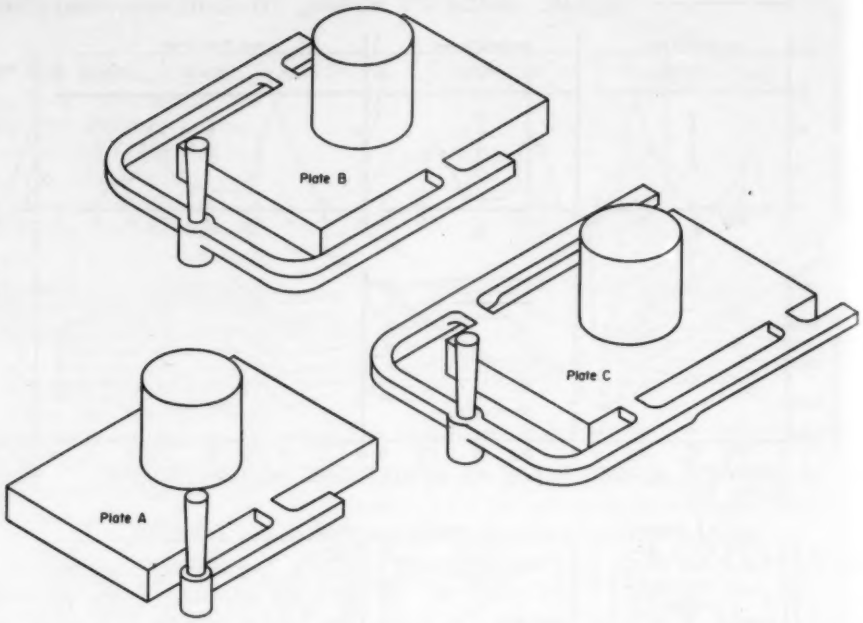


Fig. 8 — Location of single and multiple gates for cast plates fed by a top riser.

age occurs in the casting. Multiple gates would be expected to produce a more uniform distribution of the metal within the mold cavity and thus help to bring about uniform feeding.[1] Such gating systems are directly opposed to those employed to obtain directional solidification. However, if the foundryman must use a top riser and must avoid severe shrinkage voids, multiple gating may achieve this objective provided a certain amount of microporosity can be tolerated.

The effect of increasing the number of gates to 2 or 4 per plate was investigated for plates measuring 1 x 5 x 8 in. and 2 x 6 x 10 in. These sizes were chosen because with a top riser they were not radiographically sound when a single gate was attached to one edge. The multigating systems are illustrated in Fig. 8.

No thermal history of the plates was obtained, but the soundness of the plates as revealed by radiographs is depicted in Fig. 9. For comparison, the data for similar plates fed by a top riser and a single gate are included in the figure.

The soundness of the 1-in. thick plates was improved markedly by using either 2 or 4 gates per plate. Only fine microporosity could be detected directly beneath the riser. Soundness in the 2-in. thick plates was also improved, but coarse microporosity occurred beneath the riser. In both the 1- and 2-in. thick plates, the microporosity was less noticeable when 4 gates were used. It was concluded that multiple gates increase the soundness because the metal is distributed more uniformly throughout the mold cavity and localized hot spots are avoided. As a result, microporosity is more uniformly dispersed throughout the casting.

Two-In. Thick Plate With an Insulated Top Riser

The earlier experiments with single-gated, 2-in. thick plates showed that the shortest plate, which measured 6 in. long by 10 in. wide, was free of shrinkage when it was cast with a side riser, but shrinkage voids were present throughout the plate when a top riser was used. A similar plate was cast to determine whether an insulated top riser would develop sufficiently high thermal gradients in its immediate vicinity to overcome the high thermal gradient that exists in the vicinity of the gate. In the experiment, the riser insulation consisted of a thoroughly dried dental plaster cylinder that was 5 in. high, 5 in. inside diameter, with a 1-in. thick wall.

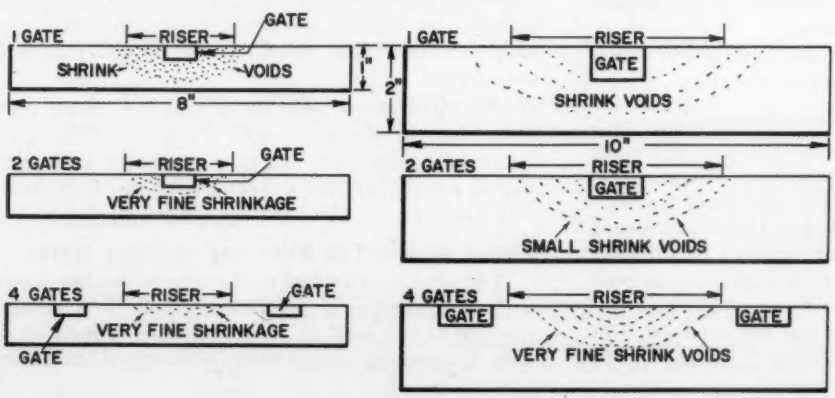


Fig. 9 — Distribution of shrinkage voids in 1- and 2 in. thick plates when gated with single or multiple gates and fed by a top riser.

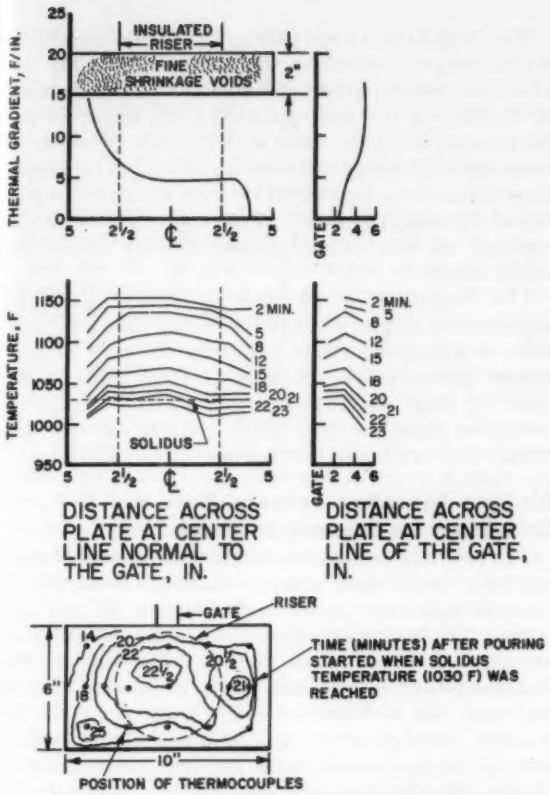


Fig. 10 — Longitudinal and lateral thermal gradients present at solidus temperature (1030 F) during solidification of a 2-in. thick plate measuring 6 x 10 in. fed by an insulated center riser.

Figure 10 shows that, on the average, larger thermal gradients were present in the center line of the plate cast with the insulated riser than with an uninsulated riser (Fig. 7). However, most of the thermal gradients were less than 5 F/in. As a result, small shrinkage voids were again present throughout the plate. No increase in feeding range was detected.

The feeding range of the riser could be increased only if the temperature of the metal in the riser could be raised appreciably to increase the thermal gradients between the riser and the plate. However, the insulating sleeve merely increased the time of solidification of the metal in the riser. As a result, the thermal gradients in the vicinity of the riser remained below the minimum required to obtain maximum soundness.

Riser Diameter Reduction

Decreasing the diameter of the insulated riser will reduce the time for solidification of the metal in the riser but, because of the short feeding range in the 2-in. thick plates, shrinkage will still occur under the riser. To demonstrate this in principle, the insulated 5-in. diameter riser 5 in. high was replaced by an insulated 2½-in. diameter riser also 5 in. high. Soundness was improved near the extremities of the plate, but shrinkage voids were again present directly beneath the riser. The smaller insulated riser was beneficial because more of the cope surface of the plate was chilled by the mold wall. Consequently, the high thermal gradients at the edge of the plate were extended toward the top riser. However, the thermal gradients beneath the riser were below those required for soundness.

It must be concluded, therefore, that the insulating sleeve by itself does not increase the feeding range of the riser. The feeding range of the riser can be said to be increased only when shrinkage under the riser is completely eliminated. When the plate is within the feeding range of the riser, an insulating sleeve can be used to increase the casting yield by decreasing the size of riser that is needed. It should be kept in mind that insulating sleeves must be dry at the time that the mold is poured. If the sleeve contains moisture, this will result in somewhat earlier freezing of the riser.

EFFECT OF CHILL LOCATION

Chilled Plates With Side Risers

The effect of chills on the soundness of steel castings has been studied to a great extent.[11-13] The investigations showed that a chill of T thickness placed at the end of a steel plate or bar opposite the riser, increases the feeding range by 2 in. for plates and by T in. for bars.[11] Tapered chills made gun-metal castings pressure tight even when the melt from which the castings were produced had a high gas content.[12] Information on the effect of conventional chills on the thermal gradients in nonferrous castings is, however, scarce. Such data are necessary if improved techniques for production of sound, light metal castings are to be established.

The effect of steel chills placed along the edge of the A1-7Mg alloy plate opposite the riser was selected for the initial investigations. Chills in this location could be used advantageously in commercial foundries to produce sound castings. The investigations were confined to plates 1 and 2 in. in thickness because the work described in an earlier section of this paper showed the ½-in. thick plate was sound without chills. The dimensions of the chills were as follows:

Chill Dimension, in.	Plate Dimensions, in.		
	Thickness	Width	Length
1 x 1 x 5	1	5	3, 5, 8
2 x 2 x 10	2	10	6, 10, 16

The thermal gradients established in the plate by using end chills are shown in the two upper rows in Fig. 11. The minimum longitudinal thermal gradient in the two shortest 1-in. thick plates (3 and 5 in.) and in the shortest 2-in. thick plate (6 in.) was about 15 F/in. No shrinkage voids were detected in the radiographs of these plates even though the lateral thermal gradient in the center of the 2-in. thick plate was down to about 5 F/in.

Shrinkage voids were still present in the 1-in. thick plate 8 in. in length, and in the two 2-in. thick plates 10 in. and 16 in. in length. Although the thermal gradients near the chilled end of these plates were as large as 100 F/in., the gradients decreased either to less than 5 F/in. or they no longer existed at the location where shrinkage voids developed. Comparison of the unchilled and chilled plates (Figs. 4 and 11, respectively) shows that, in the vicinity of the

chills, higher thermal gradients were present in the 1-in. thick plate than in the 2-in. thick plate.

Nevertheless, the rate of cooling near the riser was similar in both thicknesses of plates. As a result, feeding from the riser in the last stage of solidification was difficult and shrinkage voids developed.

End Chill Influence

These studies show that end chills exert a potent but short-range influence on the rate of solidification. After a solid skin forms next to the chill, heat must be extracted through the solidified metal. Thus, the thermal conductivity of the solidified metal becomes the controlling factor for dissipation of heat into the chill. As a result, the effect of the chill on controlling directional solidification does not extend to a great depth into the casting.

Two additional 2-in. thick plates were cast to determine whether directional solidification of the desired degree could be obtained readily with steel chills in locations other than at the far edge of the plate. Because the shortest 2-in. thick plate (6 in.) was sound radiographically, only the two longest plates (10 and 16 in.) were cast. Two 2x2x10-in. steel chills were used with each plate. One chill was again placed along the edge opposite the riser, and the other chill was in contact with the drag surface across the width of the plate at the center.

The chills were placed at the center because the thermal gradients at this location in the two longest plates with an end chill equaled the minimum thermal gradient (15 F/in.) present in the shortest 2-in. thick plate.

The magnitude of the thermal gradients present in the two plates is shown in the bottom row of Fig. 11. Although thermal gradients increased in the immediate vicinity of the chill located in the center of the plates, sections of the plate at a distance of about *T* from the chill were relatively unaffected. The immediate demand for feed metal by both the end and bottom chills caused gross shrinkage voids to form in the top half of the castings almost directly above the center chill.

The reduction in thermal gradients to values approaching zero near the risers showed that solidification was completed about 2 in. from the riser in both of the plates before sections of the plate farther from the riser were completely solidified. It must be concluded, therefore, that center chills of the type described here are ineffective in promoting soundness.

Air Gap Formation Between Plate and End Chill Effect on Thermal Gradients

The inability of the end chill to maintain a continuous high rate of heat removal has been attributed to the formation of an air gap between it and the casting. To observe the effect that the formation of an air gap may have upon the establishment of the thermal gradients, 2-in. thick plates measuring 6 by 10 in.* were cast with an end chill placed along the 10-in. edge. The opposite edge had a side riser to which the gate was attached. With the first plate, plate *A* of Fig. 12, the chill was used in the conventional

*A plate of this size was radiographically sound when fed by an end riser and not chilled.

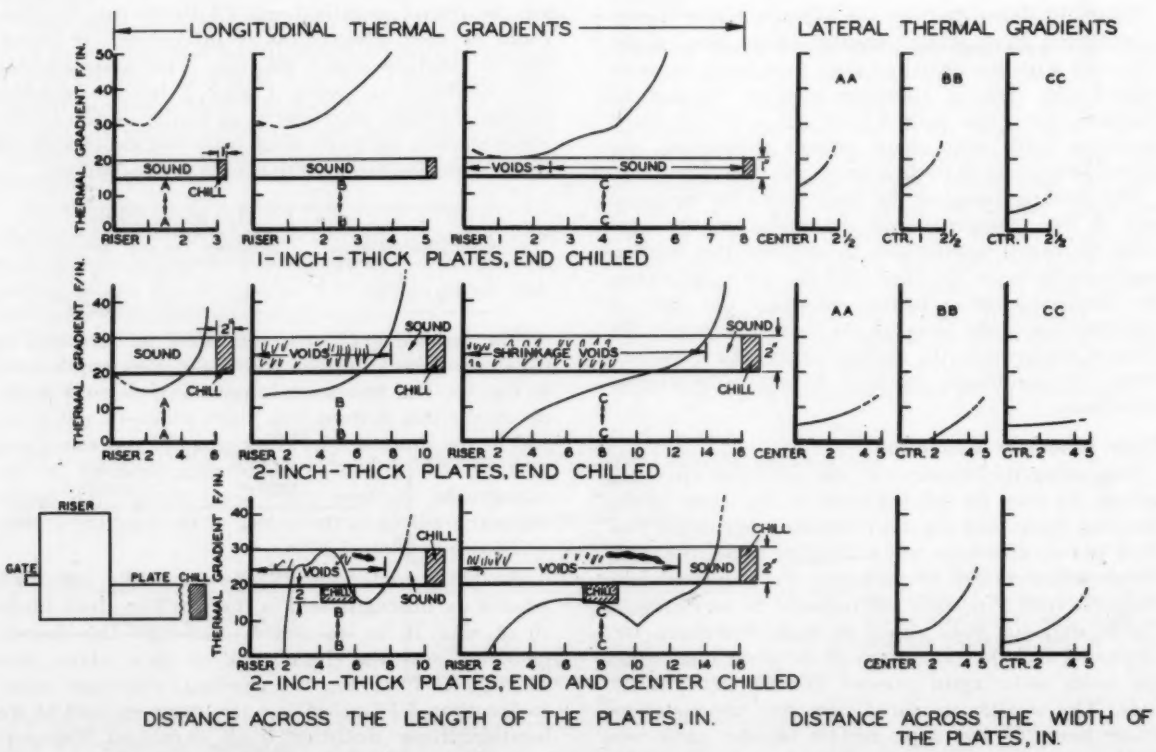


Fig. 11 — Longitudinal and lateral thermal gradients present at solidus temperature (1030 F) during the solidification of Al-7Mg alloy chilled plate castings.

manner and, therefore, the plate was free to contract away from the chill as solidification progressed.

The chill used with the second plate, plate *B*, contained two bolts which projected from the chill face into the mold cavity. Thus, the chill was, in effect, bolted into permanent contact with the cast plate during solidification of the casting.

The thermal history in Fig. 12, obtained from thermocouples embedded in the chill and the casting, showed that an air gap did develop when the chill was not deliberately secured to the casting. The air gap developed approximately 8 min after the pouring of the plate was started. During this period, the thermal gradients present at the solidus temperature in plate *A* were essentially the same as those in plate *B*, which had no gap between the casting and the chill. However, a greater rate of cooling below the solidus temperature was obtained in plate *B* than in plate *A*.

Although the increase in the rate of cooling was attributed to the continued contact between the plate and the chill, the increased rate occurred after solidification was completed. Thus, these experiments demonstrated that more continuous extensive directional solidification cannot be expected from end chills that are made to retain contact with the casting throughout the solidification period of the casting.

The maximum temperature of the chill was higher when a gap was present between the plate and the chill. Its higher maximum temperature was attributed to the fact that the melt for casting plate *A* was about 25 F hotter. The probability also existed that the thermocouple in the chill was located slightly closer toward the casting that had the loose chill than was the case for the casting that had the fixed chill. Under more precise experimental conditions, it is believed that both types of chills would reach essentially the same maximum temperature. In any event, the apparent difference in maximum temperature of the chills had no effect on the extension of directional solidification in the casting.

Wedge-Shaped Chills and Their Effect on the Thermal Gradients and Soundness of 1- and 2-In. Thick Plates

Work by Johnson, et al.,[10] had shown that the proper combination of thermal gradients and directional solidification necessary for soundness can be obtained more readily by using wedge-shaped chills. To demonstrate this, several 1- and 2-in. thick plates were made with wedge-shaped chills.

The 1-in. thick plates were cast with two chills, and the 2-in. thick plates were cast with four chills. Both thicknesses of plate were fed by an end riser to which the gate was attached. The two chills used with the first 1-in. thick plate measuring 5 by 8 in., plate *C* of Fig. 13, were 7½-in. long and 1½-in. thick, and tapered in width from 2½-in. to ⅛-in. These chills were located in the drag and were positioned parallel with each other. Their wide end was set at the edge of the plate opposite the end riser. Thus, the chills extended the length of the plate to within ½-in. of the risered edge.

Radiographs of the plate with tapered chills (plate *C*, Fig. 13) showed that shrinkage was present for a distance of only about 1 in. from the risered edge, as compared with a distance of 2½-in. in either an unchilled plate (plate *A*) or in a plate with an end

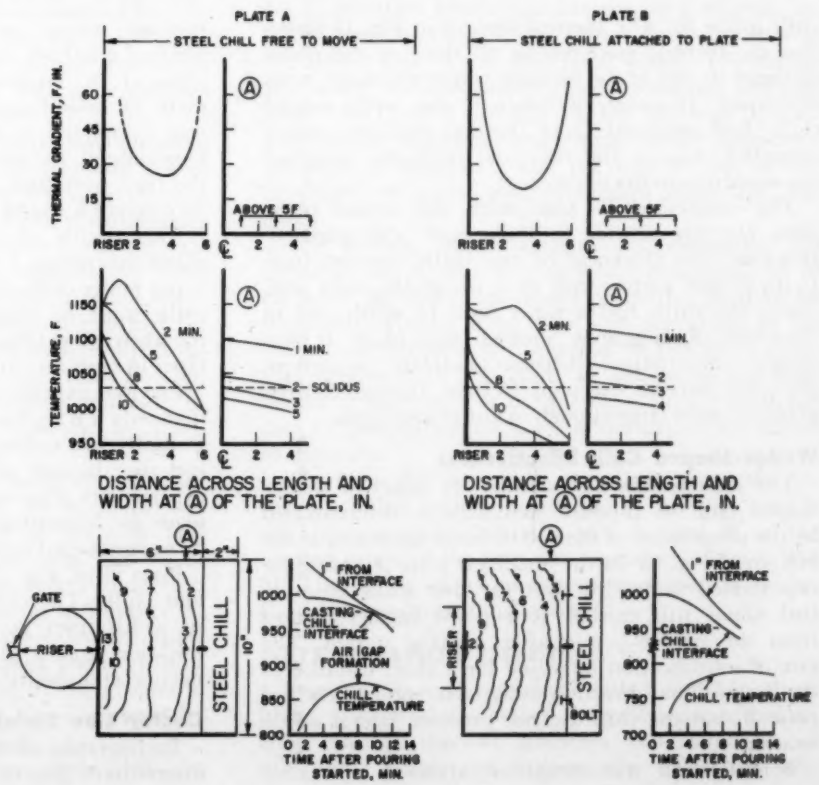


Fig. 12 — Effect on the thermal gradients of the formation of an air gap between the plate and the chill placed at the end opposite the riser of 2-in. thick Al-7Mg plates.

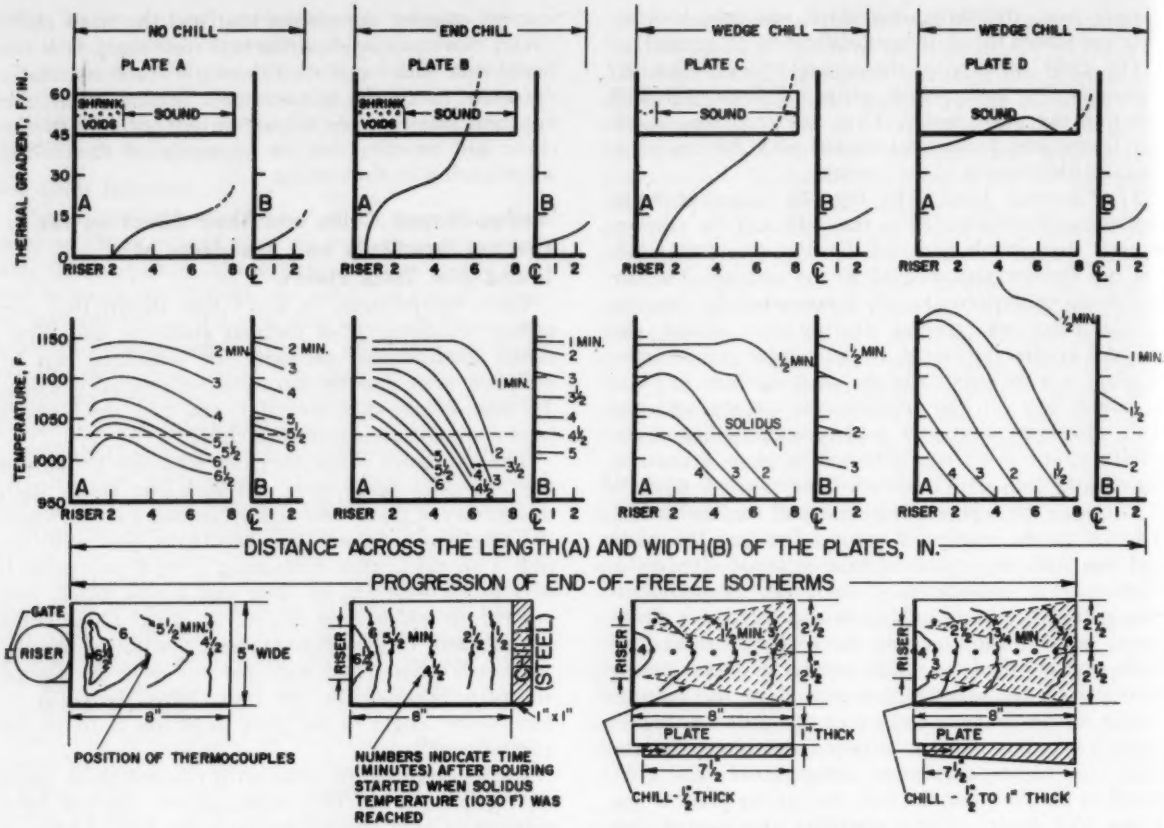


Fig. 13 — Comparison of the thermal gradients and soundness obtained in 1-in. thick plates measuring 5 x 8 in. when chills are plated at the end opposite the riser and when wedge chills are placed along the length of the plate.

chill (plate *B*). The thermal history in Fig. 13 shows that the thermal gradients in all three of the plates dropped to zero at the location where shrinkage voids developed. However, in plate *C* the wedge-shaped chills had produced large thermal gradients which extended close to the riser and markedly extended the soundness in the plate.

The tapered chills used with the second plate, plate *D*, were similar to those used with plate *C*. However, the thickness of the chills tapered from ½-in. at the narrow end to 1 in. at the wide end. Thus, the chills had a taper both in width and in thickness. Radiographs showed that plate *D* was sound. Longitudinal thermal gradients throughout the plate were in excess of 35 F/in. Lateral thermal gradients were large enough to assure soundness.

Wedge-Shaped Chill Effectiveness

The effectiveness of the correctly designed wedge-shaped chill to produce soundness is demonstrated by the progression of the end-of-freeze isotherms at the bottom of Fig. 13. In the unchilled plate *A*, solidification started to develop near the riser while unsolidified metal still existed at a point further distant from the riser 6½ min after pouring started. The rate of solidification increased for a short distance in the longitudinal direction away from the end chill of plate *B*, but the chill did not promote lateral solidification.

Solidification was completed about 4 min after pouring started in plate *C*, but the longitudinal thermal gradients were nonexistent for a distance of about 1 in. immediately in front of the riser. In plate *D*, solidification at the riser-casting interface was completed 4 min after pouring started. The longitudinal thermal gradients at the end opposite the riser were high and carried through to the riser to produce a sound plate.

The results of these experiments showed that plates measuring 1 x 5 x 8 in. can be cast sound by using wedge chills of the correct design. Wedge-shaped chills must be designed to provide an impetus to the thermal gradients. This is achieved by using a taper in both the thickness and width of the chill.

Previous experiments with 2-in. thick plates showed that only a 6 x 10-in. plate of A1-7Mg alloy cast horizontally was radiographically sound when fed by an end riser located on the 10-in edge. For this reason, experiments were conducted with a 2 x 10 x 10-in. plate to demonstrate how its soundness is affected by wedge-shaped chills.

Plate *B* of Fig. 14 shows the arrangement of the four chills. The chills were 10 in. long, and tapered in width from 2½ in. to ⅛ in. They tapered in thickness from 2 in. at the wide end to 1 in. at the narrow end, positioned parallel in the drag.

Center Line Shrinkage

Radiographs of the plate showed that only a fine intermittent line of shrinkage existed in the casting.

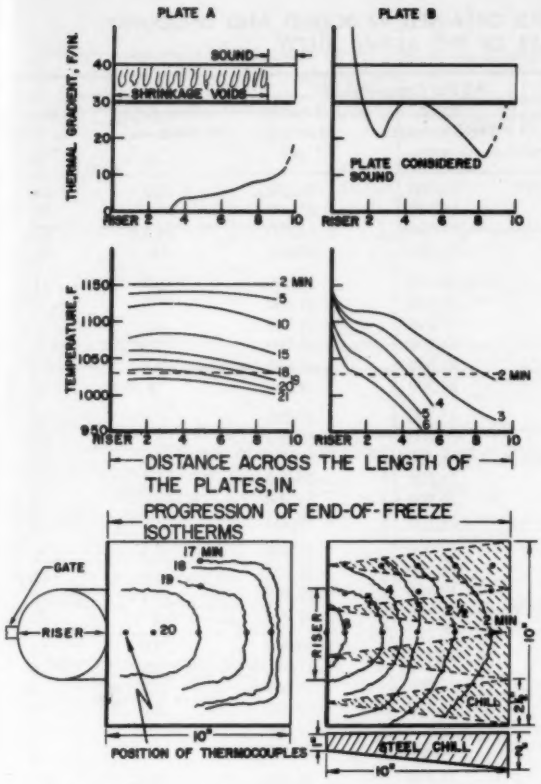


Fig. 14 — Comparison of thermal gradients and soundness obtained in 2-in. thick plates measuring 10 x 10 in. when wedge chills are placed along the length of the plate.

This center-line type of shrinkage extended from immediately in front of the riser to the opposite edge of the plate. The shrinkage occurred primarily between and above the two center chills and at a distance of about ¼-in. from the cope surface.

The thermal history of the plate was obtained from thermocouples embedded at the geometric center line of the plate. As shown in Fig. 14, large thermal gradients were present at the end of the plate opposite the riser. In front of the riser, however, the thermal gradients were from 5 to 8 F/in., close to the minimum required for soundness. Apparently the thermal gradients were below the minimum required for soundness in the vicinity where center-line shrinkage actually occurred which was ¾-in. above the geometric center line of the plate. Thermal gradients in this region could be increased, however, by chills with a still greater taper.

Again the effectiveness of the wedge-shaped chill in producing soundness is shown by a comparison of the thermal gradients present in an unchilled and chilled plate, plates *A* and *B*, respectively, Fig. 14. In the unchilled plate *A*, solidification started to develop near the riser while unsolidified metal still existed at a point farther distant from the riser. Solidification in this plate was not completed until about 20 min after pouring started. Solidification at the geometric center line in the chilled plate *B* was completed in about 6½ min after pouring started.

The results of these experiments show that the large thermal gradients and progressive directional solidification required to produce sound castings up to 2 in. thick can be obtained by wedge-shaped chills with a double taper. In everyday foundry practice, the general shape of the tapered chill can be approximated by using a series of button or spot chills which decrease in size as they approach the riser.

Comparison of Room Temperature Tensile Properties Obtained in Sound and Unsound 1- and 2-In. Thick Plates

When casting the A1-7Mg alloy, the feeding range of the riser decreases as the section thickness increases, which is contrary to what happens in casting steel. For example, a 1-in. thick plate of the aluminum alloy cast horizontally can be fed by an end riser for a distance of 5T, while a 2-in. thick plate can be fed for a distance of only 3T. Nevertheless, the experiments in the preceding sections of this paper show that the feeding distance in 1-in. thick plates can be extended from 5T to at least 8T by the use of wedge-shaped chills. Similarly, the feeding distance can be extended from 3T to 5T in 2-in. thick plates. The increase in feeding distance is the result of the large thermal gradients produced by the wedge chills.

The soundness of the plates cast from the A1-7Mg alloy has been based on radiographic examination. The acceptability of the plates and the methods used to achieve soundness should be reflected by the tensile properties of the plates. Therefore, plates were cast to determine whether an improvement in tensile properties does occur when wedge chills are used to extend the feeding range of a riser.

One- and 2-in. thick plates were cast with an end riser. The dimensions of the plates are listed in Table 2. Standard tensile specimens with a 0.505-in. reduced section were machined from coupons cut from the center line of the individual plates. The tensile properties obtained from these specimens are presented in Table 2 and Fig. 15. For purposes of comparison, the minimum and typical tensile properties obtained on separately cast test bars by one of the producers of commercial aluminum ingot are also given.

The average tensile strength and elongation obtained in the plates cast with wedge chills equaled or exceeded the minimum tensile properties obtained in separately cast test bars. The high properties of the chilled castings were attributed to the fine grain structure produced by the chills as well as to the high degree of soundness of the plates. It should be noted that neither manganese nor a grain refiner was added to the melts.

The data also show that the tensile properties for the unchilled plates decreased 1) as the section thickness increased (plates 1 and 4), and 2) as the feeding range of the riser was exceeded (plates 2 and 5). Thus, it is shown that tensile properties do reflect the soundness of the castings.

GATE LOCATION EFFECT

Experiments were conducted with 1- and 2-in. thick plates to determine whether the gate location has an effect on the thermal gradients and soundness of side-risered plates cast horizontally. Previously, the

TABLE 2 — ROOM-TEMPERATURE TENSILE PROPERTIES OBTAINED IN SOUND AND UNSOUND 1- AND 2-IN. THICK PLATES CAST OF THE Al-7Mg ALLOY

Plate	Plate Dimensions, in. Thickness	Length	Width	Sound	Specimen[a]	0.2 Per Cent Offset Yield Strength, psi	Tensile Strength, psi	Elongation in 2 in., %	Reduction in Area, %
1	1	5	5	Yes	1	15,400	32,200	12.3	14
					2	14,200	27,400	11.6	16
					3	13,500	28,700	14.3	15
					Ave.	14,400	29,400	13	15
2	1	8	5	No	1	13,100	24,000	9.7	11
					2	13,300	21,400	7.7	8
					3	14,200	24,100	8.5	8
					4	14,100	25,100	9.0	9
					5	15,000	25,500	9.5	14
					Ave.	13,900	24,000	9	10
3[b]	1	8	5	Yes	1	13,700	33,200	16.9[c]	20
					2	14,700	36,400	25.0	40
					3	14,400	35,800	22.0	26
					4	14,800	36,200	23.7	31
					5	15,900	37,800	19.1	25
					Ave.	14,700	35,900	21	28
4	2	6	10	Yes	1	14,200	24,500	8.3	11
					2	14,000	23,100	8.0	11
					3	15,800	27,300	10.4	15
					Ave.	14,700	25,000	9	12
5	2	10	10	No	1	13,400	23,600	9.0	11
					2	12,400	17,600	5.2	5
					3	12,500	18,500	6.3	7
					4	13,400	20,900	8.0	7
					5	—	24,900	10.3	17
					Ave.	12,900	21,100	8	9
6[b]	2	10	10	Yes	1	15,600	35,400	15.3	27
					2	14,700	33,200	18.8	28
					3	14,600	34,200	22.5	38
					4	14,400	35,400	26.2	32
					5	13,900	34,400	24.7	37
					Ave.	14,600	34,500	22	32
Minimum[d]							35,000	9	
Typical[d]							40,000	13	

a) Specimen 1 is in front of riser; Specimens 3 and 5 are at end opposite riser (Fig. 15).
b) Plates cast with wedge chills.
c) Broke outside gage marks.
d) Data obtained from "ALMAG 35" published by William F. Jobbins, Inc., Aurora, Ill.

gate had been connected directly to the side riser. In the present experiments the gate was attached to the edge of the plate opposite the side riser. The thermal history of the 1- and 2-in. thick plates (plates *B* and *C*, respectively) is presented in Fig. 16. Similar data obtained previously for a 1-in. plate gated directly into the side riser are included as plate *A*, Fig. 16, for comparison purposes.

Plates *A* and *B* measured 1x5x5 in.* From the thermal history of plate *B*, it was apparent that gating opposite the side riser increased the time of solidification to 10½ min as compared with about 6 min for plate *A* which was gated directly into the side riser. Although longitudinal thermal gradients were large in front of the riser of plate *B*, the lateral thermal gradients in front of the gate dropped below the 5 F/in. required for soundness. A small amount of center-line shrinkage along the length of the plate was revealed by radiography.

The second experimental casting (plate *C*) measured 2x6x10 in. Thermal gradients in this plate fell below 5 F/in. in front of the gate. Shrinkage was present throughout the plate as compared with only a small amount of center-line shrinkage in a similarly gated 1-in.-thick casting (plate *B*).

It was apparent that gating at the edge of a plate opposite the risered edge decreases the soundness of the plate. The decrease in soundness results from the localized heating of the sand in the vicinity of the gate. The heating of the sand, in turn, causes the thermal gradients in the plate to drop below the 5 F/in. required for soundness.

CUBES AND BARS OF Al-7Mg ALLOY CAST HORIZONTALLY

Investigations were conducted on cube and bar castings to 1) determine the feeding range of the riser, and 2) observe the effect that chills have on the thermal gradients and soundness of the castings.

The bars were cast horizontally. For all castings, the riser was located at one end, and the gate entered this side riser. The dimensions of the cube and bar castings and the riser that was used with each casting (giving length of cube and bar castings, and diameter and

*A plate of this size was radiographically sound when fed by a side riser to which the gate was attached.

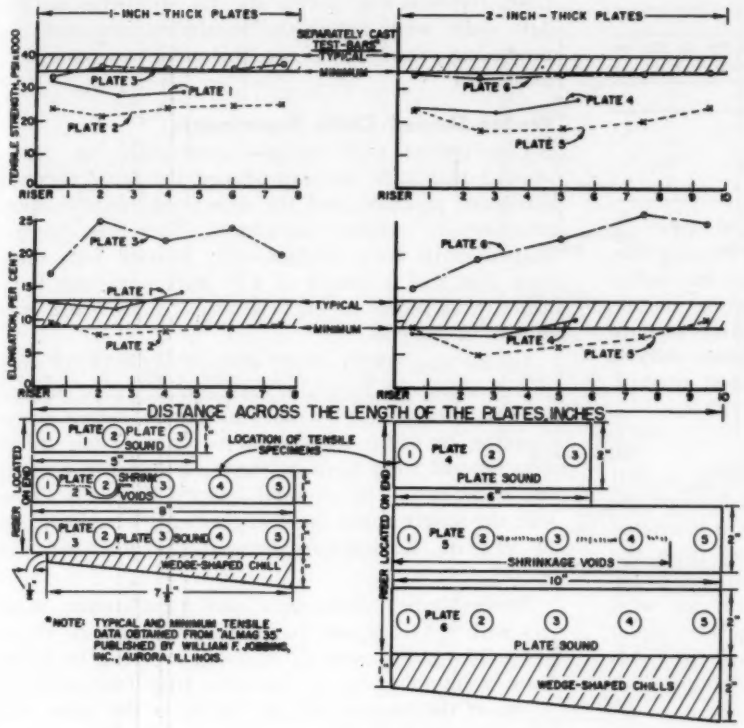


Fig. 15 — Comparison of room-temperature tensile properties obtained in sound and unsound 1- and 2-in. thick Al-7Mg plates.

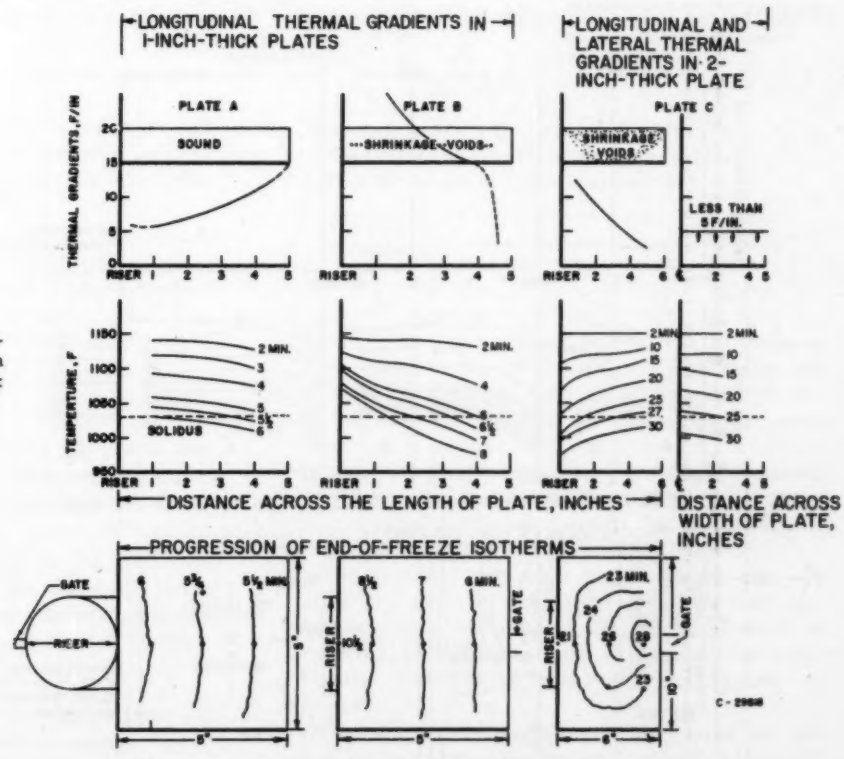


Fig. 16 — Effect on thermal gradients and soundness of changes in gate location in 1- and 2-in. thick plates fed by an end riser.

height of riser) are listed in the tabulation given below.

	Dimensions for Cube or Bar of Indicated Section Size, in.		
	2	3	5
Length of cube, in.	2	3	5
Length of bar, in.	8	12	—
Diameter of riser, in.	3	4½	7½
Height of riser, in.	5	5	5

The thermal history of the unchilled cubes is shown in Fig. 17. In the 2-in. cubes, the thermal gradients were never less than 5 F/in. The cube was radiographically sound. In the 3- and 5-in. cubes, thermal gradients were less than 5 F/in. and, as a result, they were unsound. These experiments showed that the feeding range of the riser is less than *T* for cubes with a cross-section greater than 2x2 in.

Cubes Tested

In previous experiments with plate castings, a chill placed at the end opposite the side riser increased the thermal gradients. The beneficial effect was restricted to a distance of about 1 T to 2 T away from the chill. To determine if a similar relationship exists for cubes, only the 2- and 3-in. cubes were cast with cube-shaped steel chills whose dimensions were equal to *T*. The chills were placed at the end of the cubes opposite the risers, as illustrated by the isotherm plots in Fig. 18.

The 2-in. cube was again sound, but small shrinkage voids occurred at the risered end of the 3-in. cube. The thermal gradients in both cubes were increased by the steel chill, and solidification was progressive from the chilled to the risered end of the cube. Nevertheless, lateral thermal gradients in the 3-in. cube were below the minimum required for producing soundness; hence, this cube was somewhat unsound.

Wedge-Shaped Chills Experiments

Experiments with wedge-shaped chills on plates showed that such chills produced the large thermal gradients, and provided the directional solidification necessary to achieve soundness. Therefore, wedge-shaped chills were studied with 2- and 3-in. thick bars that had a length of 4T. Bars were cast horizontally, as shown in Fig. 19, with the riser and gate located at one end.

In the first experiment, wedge-shaped chills were placed along the length at each side of a 2-in. square by 8-in. long bar (4T). The chills were 7½ in. long, tapering in width from 2 in. to ⅛-in. Their thickness tapered from ½-in. at the narrow end to 1 in. at the wide end. The thermal history (Fig. 19) shows that the longitudinal thermal gradients were in excess of 40 F/in. Radiographs showed that the bar was sound.

Wedge-shaped chills were used in a similar manner with a 3-in. square bar having a length of 12 in. (4T). The chills were 12 in. long, tapering in width from 3 in. to ⅛-in. In thickness they tapered from ¾-in. at the narrow end to 1½-in. at the wide end. The thermal history is also presented in Fig. 19, and shows that the longitudinal thermal gradients were lower than those in the 2-in. square bar. Large thermal gradients were present at the end opposite the riser,

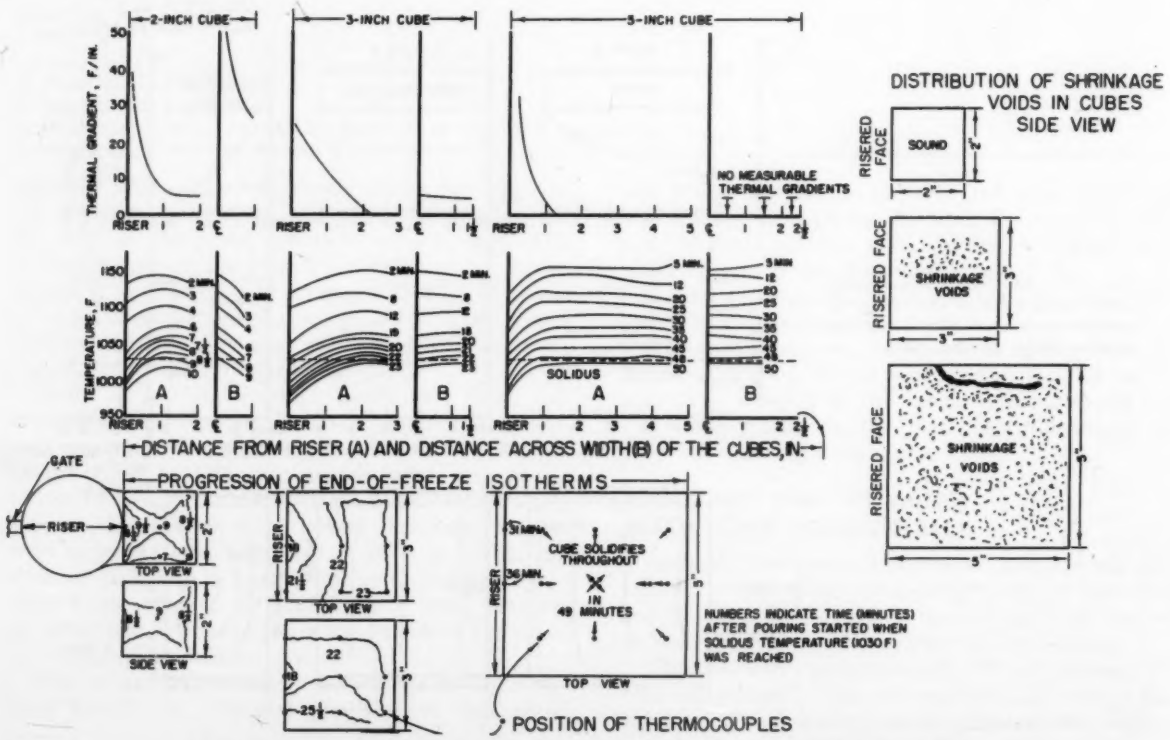


Fig. 17 — Longitudinal and lateral thermal gradients present at solidus temperature (1030 F) during the solidification of 2-, 3- and 5-in. Al-7Mg cubes fed by an end riser.

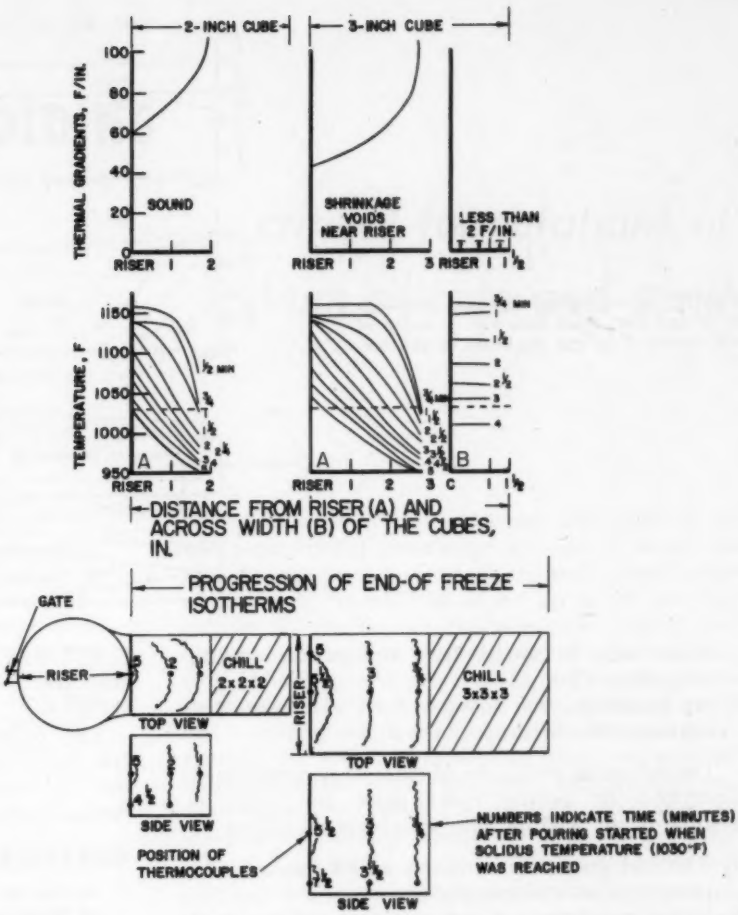


Fig. 18 — Longitudinal and lateral thermal gradients at solidus temperature (1030 F) during the solidification of 2- and 3-in. square chilled cubes of Al-7Mg alloy fed by an end riser.

but they decreased to about 30 F/in. in the middle of the bar. The thermal gradient immediately in front of the riser was only 15 F/in. Radiographs showed that this bar was sound.

These experiments reveal that 2- and 3-in. square bars with a length of 4T can be made radiographically sound with wedge-shaped chills. Thus, a method is available for producing soundness in 2- and 3-in. square bars beyond the normal feeding range of the riser.

CONCLUSIONS

The experimental work performed on plates and bars cast from the Al-7Mg alloy emphasizes that the development and maintenance of large thermal gradients throughout the plates and bars are extremely difficult. The gate as well as the riser and chills must be located correctly to create the large thermal gradients, and coordinated rate of directional solidification required for producing sound castings.

Horizontal Castings

The following observations were made from the experiments conducted with horizontal castings:

1) The feeding range of the riser in terms of *T* (the section thickness) decreases as the section thickness is decreased.
2) With a side riser, the length of plate found to be radiographically sound when expressed as a function of the plate thickness is as follows:

Thickness, in.	Length, in.
½	8T
1	5T
2	3T

3) Because the feeding range of the riser in terms of *T* decreases as the section thickness increases, additional risers rather than an increase in the size of the riser are required to produce sound castings.
4) When a top riser must be used on plates, increased soundness can be obtained by using multiple gates.
5) The feeding range in a square bar with a cross-section greater than 2 by 2 in. is less than 1 T.
6) A conventional steel chill placed at the end opposite the riser of cast plates or bars increases the thermal gradients, but the beneficial effect is restricted to the immediate vicinity of the chill. The end chill does not increase the feeding range of the riser.
7) Wedge-shaped chills with a double taper on the drag surface extend the feeding range in 1-in. thick plates from 5T to 8T. In plates 2 in. thick, the feeding range is extended from 3T to 5T.
8) Wedge-shaped chills with a double taper on op-

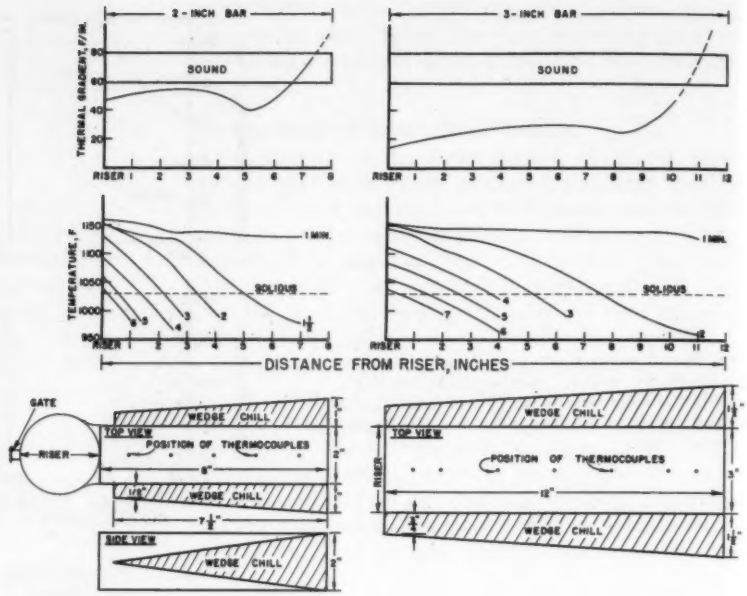


Fig. 19 — Longitudinal thermal gradients in 2- and 3-in. thick bars with a length of 4T, where *T* is the thickness of the bar.

posite sides of square bars increase the feeding range from 1T to 4T.

9) An insulated riser does not have a greater feeding range than an uninsulated riser.

The following principles of thermal dynamics were developed for locating risers, gates, and chills on castings of the A1-7Mg alloy to achieve soundness:

1) Thermal gradients in excess of 5 F/in. must be present to achieve soundness.
2) When the design of the casting permits, attach the riser to the side instead of the top of the casting.
3) When a side riser is used, attach the gate to it.
4) When a top riser must be used, employ multiple gates to obtain more uniform feeding.
5) To extend the feeding range of a riser, use wedge-shaped chills having a double taper. Place the chills so that directional solidification occurs toward the riser.
6) For a plate casting, the minimum thickness of the wedge chills at the wide end should be at least equal to the section thickness of the casting.
7) When the design of the casting permits wedge-shaped chills to be placed on both sides of a section, such as a bar, the thick end of the chill should equal at least one-half the thickness of that section.

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REFERENCES

1. Ruddle, R. W., *The Solidification of Castings*, The Institute of Metals, London, England (1957).
2. Bishop, H. F. and Johnson, W. H., "Risering of Steel Castings," *Foundry*, 70-74 (Feb., 1956), and 136-141 (March, 1956).
3. Bishop, H. F., Myskowski, E. T. and Pellini, W. S., "A Simplified Method for Determining Riser Dimensions," AFS Transactions, *63*, 271-281 (1955).
4. Bishop, H. F. and Ackerlind, C. G., "Dimensioning of Risers for Nodular Iron Castings," *Foundry*, 115-119 (Dec., 1956).
5. Varga, J., "A New Method for Studying Riser Requirements," *Foundry*, 106-109 (July, 1957).
6. Chvorinov, N., "Control of the Solidification of Castings by Calculation," *Foundry Trade Journal*, 95-98 (Aug. 10, 1939).
7. Caine, J. B., "A Theoretical Approach to the Problem of Dimensioning Risers," AFS Transactions, *57*, 492-501 (1949).
8. Grube, K. and Eastwood, L. W., "A Study of the Principles of Gating," AFS Transactions, *58*, 76-107 (1950).
9. Grube, K., Kura, J. G. and Jackson, J. H., "A Study of the Principles of Gating as Applied to Sprue-Base Design," AFS Transactions, *60*, 125-136 (1952).
10. Jackson, R. S., "Application of Insulated Feeders to Sand Castings in Long-Freezing-Range Copper Alloys," *Foundry Trade Journal, 100*, 487-493 (1956).
11. Myskowski, E. T., Bishop, H. F. and Pellini, W. S., "Application of Chills to Increasing the Feeding Range of Risers," AFS Transactions, *60*, 389-400 (1952).
12. Johnson, W. H., Bishop, H. F. and Pellini, W. S., "Application of Chills to Improve Pressure Tightness of Gun Metal (88-8-4)," AFS Transactions, *62*, 243-251 (1954).
13. Walther, W. D., Adams, C. M. and Taylor, H. F., "Techniques for Improving Strength and Ductility of Aluminum Alloy Castings," AFS Transactions, *62*, 219-230 (1954).

OPTIMUM CO_2 MOLDING

critical formulations of sodium silicate and sand

By Roderick J. Cowles

ABSTRACT

The difficulty of poor collapsibility, predominate as a limitation in the sodium silicate-CO_2 method of sand bonding, is obviated by establishing simple laboratory test methods to determine an optimum quantity of sodium silicate binder. This is designated as the minimum critical percentage for each sand considered. Additives such as sugar, coarse grain silica flour and iron oxide, properly used, contribute to the flowability, ramming qualities and green strength and enhance collapsibility as well as provide hot strength and cushioning during thermal changes.

The binder requirements should be developed on the hypothesis of adhesive bonding rather than on the basis of a mortar composition, which fills voids between the grains. Adhesive binders are most effective in thin films and with the greatest number of sand grain to grain adhesive contacts uniformly distributed in the sand structure. The viscosity and penetrating qualities of the binder composition determine the degree of correlation that binder requirements have to B.E.T. specific area measurements.

The B.E.T. data can be used to designate those sands which will need more or less binder than their AFS fineness number indicates. A straight line correlation between minimum critical percentage and a dimensional parameter calculated from AFS fineness data, shape factors, ratio multipliers and specific volume ratios raised to the fourth power is given.

The limitations due to high relative humidity and the use of clay additives are discussed. Experimental data, as well as numerous references, are used to support these findings.

INTRODUCTION

The problems of the sodium silicate-carbon dioxide method of bonding sand cores and molds have been recognized and enumerated in many references. A principle difficulty, collapsibility, is obviated in this work by establishing simple laboratory test methods to determine an optimum quantity of sodium silicate binder. This "minimum critical percentage" varies according to the foundry sand and the type of sodium silicate selected.

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The addition of 1.5 to 2.0 per cent sugar to the minimum critical percentage amount of binder establishes adequate green strength, and allows collapsibility with the addition of 0.4 up to 0.8 per cent sodium silicate beyond the minimum critical percentage amount. The proper choice of sodium silicate with enhancing additives such as sugar, silica flour and iron oxide can produce a composition of excellent ramming and flowing qualities. The many advantages for a foundry operation using properly formulated sodium silicate-CO_2 plus additive compositions more than offset the particular limitation of sensitivity to relative humidities above 76 per cent.

High ratios of silica to soda tend to retard this degradation, but for storage purposes the relative humidity should be controlled below approximately 70 per cent. Since the effect of moisture is reversible, any molds or cores air-dried before handling will regain their strength and surface hardness for normal foundry use.

Some sacrifice of other properties such as strength and surface hardness must be accepted when selecting high ratio silica to soda binders to assist in this problem of high humidity where the cores and molds can be used within a short time after gassing with CO_2. The preferred silicates used in relative humidities below 70 per cent have a viscosity range between 300 and 2000 centipoise, a ratio of silica to soda (between 2 to 2.4) to 1 and total solids of 43 to 47 per cent.

Sand Grain Contacts Adhesive

The sodium silicate, silica gel, sugar combination as the basic bonding material should be considered as an adhesive between sand grain contacts rather than a mortar composition filling voids between the grains. Such an adhesive is effective in thin films with the greatest number of small areas uniformly distributed as adhesive contact points in the sand structure. This accounts for a noticeable increase in gassed strength of cores and molds formulated with an addition of approximately 4 per cent of a coarse 90 to 140 mesh silica flour together with the recommended quantities of binder. Improved distribution of the binder

with a considerable increase in sand to sand contacts provides optimum use of the sodium silicate-sugar binder formulation.

Iron oxide aids collapsibility, provides hot strength during metal solidification and, due to a soda-iron oxide-silica eutectic composition at high temperatures, results in a smooth surface at the mold metal interface which prevents molten metal penetration.

This paper presents specific surface area data for various foundry sands. The B.E.T.* method used is a sensitive indication of differences between round smooth grain sands and those of angular rough surfaces. Highly adsorbent impurities will indicate greater surface areas than the proportional need for binder indicates; but this binder increase depends upon the viscosity and penetrating qualities of the binder composition. Although the binder requirements are not directly related to B.E.T. area differences, the specific area measurements can be used to designate those sands which will need more or less binder than their AFS number indicates.

Dimensional parameters of these sands are calculated from their AFS fineness data, together with their comparative specific volumes raised to the fourth power, and correlated in a straight line relationship with the minimum critical percentage requirement of a preferred sodium silicate binder for each sand.

Formulations for foundry molding of cores or molds by the sodium silicate binder-CO_2 setting method are recommended for the five sands which have been studied in this particular development approach. The formulations depend considerably upon the parameters of the sand which have been determined in their relation to the binder requirements.

THEORETICAL CONCEPTS FOR SILICATE BONDING

Sand size distribution for optimum structural properties in foundry molds has progressed measurably in the past 40 years.[1-11] The mold formation in most cases consists of a mortar-like bonding built up between the sand grains—a surface coating over the sand to retain permeability if possible, but adequate to capsulate and establish conforming structural units between the sand particles.[12] These bonding materials consist of inorganic-organic combinations such as water-clay-cereal flour and sea coal which contribute green strength as well as hot strength and form a cushioning effect between sand particles. Diet[11] mentions that cushioning additives will eliminate or reduce mold wall fractures in such a bonded system.

Foundry technology depends upon our knowledge of properly bonding sands or other similar refractory particles for handling prior to casting, and for retaining the metal as it solidifies without surface cracking or reacting with the molten metal. Dimensional stability of the mold mass is also important.[13]

There is then a difficult transition to accept new bonding concepts for foundry molds, and perhaps some of the requirements for optimum molding qualities are lacking. The nature of the sodium silicates requires a different approach; one which in many ways parallels that of phenolic resins in shell molding.[14] In either case, we should not expect the silicate or phenolic to act as a structural filling mortar between the sand grains. As a hypothesis, we must accept these bonding agents as adhesives between the closest surfaces of the refractory particles (Fig. 1).

Work of Adhesion

Sprinkle and Taylor[14] present valid data to substantiate the relation between strength or work of adhesion of the phenol-formaldehyde bond and shell mold tensile strengths. The science of adhesion, and its relation to surface wetting and intermolecular forces, is complex enough to establish many scientists as full time contributors in its continued study. Symposium papers[15] by Reinhart, Czyak, Kraus, Kemball, Bikerman, Blomquist, Koehn, W. S. Macfarlane and J. F. Sewell are only an indication of the many studies both fundamental and applied related to adhesives.

Once we make the assumption that we have an adhesive bonding system, the approach has been considerably established. Some fundamental concepts of adhesive technology should not be ignored in attaining better understanding of the sodium silicate bonded sand system.

A more detailed description of the physio-chemical properties of the sodium silicates themselves are given in a later section. Adhesive bonding considers five different phases or layers. These phases are solid, to surface layer, to adhesive, to surface layer, to second solid. If the bonding strength between any of these phases is weaker than the stresses locally applied, rupture of the adhesive bond follows. Of course, the sand surfaces themselves form the interlayer between solid and adhesive. Impurities on this surface layer as microscopic powders, such as clays and salts, may weaken the bond in this phase.

This surface layer between the solid and adhesive should give the least interference to optimum forces of molecular attraction. The mechanical surface construction of the sand is also important. Sand surfaces which are broken and contain cracks are subject to fracture under mold stresses. Smooth, clean surfaces provide best adhesive bonding with the minimum binder requirements. If the surface layer bond structure is strongest, the solids themselves or the adhesive may rupture. Mechanical impact, thermal stresses by shock or expansion may cause breaking of the solid sand material behind the surface interlayer.

*Method of evaluation devised by Brunauer, Emmett and Teller referred to later in this report.

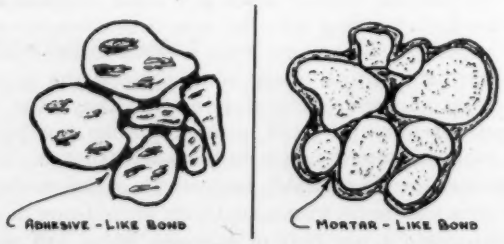


Fig. 1 — Left — Phenolic or sodium silicate bonding for sand molds and cores. Right — Normal clay bonding for sand molds and cores.

In addition, the mechanics of the adhesive phase must be considered. The cohesive strength within the adhesive is dependent upon the uniformity, the thickness of the layer and the structural composition. The thinner and more uniform adhesive composition has the least probability of structural rupture.

Breaks in the bond can occur because of tensile stresses, shear stresses or impact stresses. The composition of the bond, the manner in which it is formed, the quantity and distribution of this bonding material, as well as the number of bond contacts, can produce a wide variety of physical strength values within the mold or core. Those combinations which give high tensile and compressive strengths may have bonds of poor flexibility and toughness. A combination of these properties is most desirable.

A sand adhesive bond structure may have high compressive strength and poor resistance to surface rubbing or foundry washing. This point is substantiated by data (Table 2, experiments 251, 327, 324, 326; Table 3, experiment 118C,) where it is noted that compressive strengths do not correlate with surface hardness results.

EXPERIMENTAL STUDIES RELATED TO SILICATE FORMULATIONS

Simple procedures of evaluating sodium silicate bonded, CO_2 gassed-sand formulations by forming and testing laboratory cylinder specimens has enabled a comparatively comprehensive correlation of numerous interrelated variables. Plant trials and proven production formulations have substantiated many of these experimental findings. The first laboratory and plant run experiments provided a basis for the more comprehensive studies which followed, and are presented here.

Accepting the hypothesis of adhesive bonding, and the many published variations of sodium silicate compositions for consideration, the experimental possibilities become uncountable. A selection of conditions and the limits for their variation must be made initially to keep the quantity of experimental studies at all reasonable. Even with this screening approach thousands of specimen were tested.

It was established that any of the properties such as:

TABLE I — SOME FORMULATION STUDIES WITH CYLINDER CORE SPECIMEN USING AFS 105 SAND CODE D

Exp. No.	Silicate Binder Used, No.	Binder, %	Additives, %					Compressive Strength, (psi)		Dietert Surf. Hardness		Relative Humidity, %	Comments
			Sugar	Water	Iron Oxide	Silica Flour	Other	Before Heat	After Heat	Before Heat	After Heat		
173	1	3.7	2.0	0.5		4.0		550	110	83	66	Room	Poor collapsibility
173	1	3.7	2.0	0.5		4.0		3	154	30	71	88	Poor collapsibility
174	1	3.0	2.0	0.5				500	1	77	45	56	Excellent strength and collapsibility
174	1	3.0	2.0	0.5				2	2	0	57	88	Excellent strength and collapsibility
A2AO(10)	1	2.5	2.0					400	2	85	Low	Under 60	
A2AQ(8)	1	2.5	1.0					380	2	78	Low	Under 60	
A2AQ(9)	1	3.0	1.0					628	2-200	84	0-70	Under 60	Variable collapsibility
7/29/57	1	3.0	2.5					360	80	82	0-60	Under 60	
A2AQ(11)	1	3.0	2.0					150	2-125	88	0-65	Under 60	
236	1	3.5	2.0					400	20	88	32	34	Excellent
237	1	4.0	2.0					390	75	92	40	34	Very good, but may have problems of collapsibility
238	1	5.0	2.0	0.5				140	143	84	45	72	Poor collapsibility at higher humidity
238A	1	5.0	2.0	0.5				592	0	89	0	40	Excellent at lower humidity
250	1	3.0	2.0	0.5			*0.2	510	0	80	0	40	Stand 28 days, excellent results all around
160	8	4.0	2.0	0.5				35	0	92	0	46	Inadequate strengths due to binder difference
160	8	4.0	2.0	0.5				9	0	68	0	88	Inadequate strengths due to binder difference
160A	8	3.0	2.0	0.5	2.0			10	0	88	0	Room	Inadequate strengths due to binder difference
160A	8	3.0	2.0	0.5	2.0			3	0	48	0	88	Inadequate strengths due to binder difference
160B	8	5.0	2.0	0.5	2.0			9	0	90	0	Room	Inadequate strengths due to binder difference
160B	8	5.0	2.0	0.5	2.0			18	0	77	0	88	Inadequate strengths due to binder difference 160B dries out quickly
160C	8	7.0											No good, not workable
171	8	4.5	2.0			4.0							Too sticky
171A	8	3.5	2.0			4.0		100	0	93	0	68	Fair surface
171A	8	3.5	2.0			4.0		14	0	61	0	88	Poor, due to high humidity
176	8	3.5	2.0					158	0	87	0	50	Good surface

*Cresol-phenolic organic derivative

After Heat—2200 F for 5 min

TABLE 2 — SOME FORMULATION STUDIES WITH CYLINDER CORE SPECIMEN USING SUB-ANGULAR AFS 70 CODE B

Exp. No.	Silicate Binder, No.	Binder, %	Additives, %					Compressive Strength, psi		Dietert Surf. Hardness		Flow-ability (Dietert)	Rel. Hum. %	Comments
			Sugar	Water	Iron Oxide	Silica Flour	Other	Before Heat	After Heat	Before Heat	After Heat			
203	1	2.0						95	2	53	47		Normal	Poor rub resistance,
203	1	2.0						2	2	7	42		88	good surface
204	1	2.5						77	15	56	57		Normal	Good surface, flowable
204	1	2.5						5	20	0	63		88	mix
205	1	3.0						110	170	62	68		Normal	
205	1	3.0						29	140	0	75		88	
337	1	1.5						6	0	8	0	85	100-44	Very dry, very poor surface
338	1	4.0						500	300	74	74	84	100-44	Excellent surface, poor collapsibility
206	1	2.0	2.0	0.5		4.0		183	2	84	52		40	
206	1	2.0	2.0	0.5		4.0		2	2	0	55		88	
207	1	2.5	2.0	0.5		4.0		200	22	88	56	85	40	Good except for humidity resistance
207	1	2.5	2.0	0.5		4.0		2	18	0	59		88	
207	1	2.5	2.0	0.5		4.0		263	32	85	67		52	Nov. 21, 1958
264	1	2.5	2.0	0.5		4.0		65		79			56	After 48 hr
230	1	2.3	2.0	0.5		4.0		220	40	87	63		62-74	Excellent all-around cores
230	1	2.3	2.0	0.5		4.0		2	50	10	60		88	Collapsibility not too good, low strength
236	1	3.5	2.0					400	20	88	32		86-34	Excellent mix and surface
251	1	3.0	2.0					340	0	57	0		40	Stand 29 days
262	1	3.0	2.0						90		81		56	2600 F and 2800 F
252	1	2.5	2.0	0.5	2.0	4.0		400	0	80	0		52	Stand 18 hr (good)
263	1	2.5	2.0	0.5	2.0	4.0		100	0	82	4		56	Stand 48 hr (good)
323	2	2.4	2.0					59	2.5	48	30		40-55	
323	2	2.4	2.0					19	2.5	5	30		88	Soft and soggy
327	2	2.4	2.0	0.5				188	Not Tested	37	Not Tested		40-55	Note
327	2	2.4	2.0	0.5				6	Not Tested	1.2	Not Tested		88	Soft
324	2	3.0	2.0					288	112	52	55		40-55	
324	2	3.0	2.0					219	112	57	55		10-30	Note
324	2	3.0	2.0					2.5	112	2	55		76-88	
325	2	3.6	2.0					225	Not Tested	50	Not Tested		40-55	Note
325	2	3.6	2.0					19	Not Tested	10	Not Tested		88	Soft
326	2	3.6	2.0	0.5				344	Not Tested	52	Not Tested		40-55	Note
326	2	3.6	2.0	0.5				12	Not Tested	10	Not Tested		88	Soft
341	2	6.0	2.0					600	420	84	82	82	55	Stand 48 hr, poor collapsibility
215	8	2.5	2.0			4.0			65					Too sticky to remove
216	8	3.0	2.0				3		33				80	Too sticky
217	8	3.0	2.5			4.0	145*	0	82	17		83	63	Sticky *compressive strength (Dietert)
217	8	3.0	2.5			4.0	2	0	17	17			88	
231	8	3.5	2.5			4.0	220	2	92	26			45	Dries fast
231	8	3.5	2.5			4.0	3	2	38	38			88	
287	8	3.2											46	Not workable
313	8	2.5	2.0	0.5	2.0	4.0	62	0	64	0		80	55	(More binder necessary for strength) No. 8 is a questionable binder for this sand

NOTE—All samples allowed to stand minimum of 5 hr at room conditions, then in desiccator for a minimum of 5 hr.
*Cresol-phenolic organic derivative.

a) Compressive strength.
b) Surface hardness.
c) Collapsibility with heat.
d) Ability to flow and ram into a uniform smooth load supporting core or mold surface.
e) Part cleanly from the casting, would depend most upon the variations of:
   1) Characteristics of the sand selected (see detailed discussion).
   2) Type and quantity of sodium silicate used as the binder (see description of sodium silicate binders).
   3) Additives used. Favorable ones include, 90 to 140 mesh silica flour, syrup or sugar, iron oxide, in some cases water, and a cresol phenolic organic powder used in small quantities; of controversial value are selected clays, pitch, sea coal and cereal.
   4) Relative Humidity. Structural strength and surface hardness are reduced considerably by humidities above 76 per cent. The effect of moisture is reversible, however, and the favorable properties lost by exposure to high humidity are regained in dry conditions, accelerated by oven drying if desirable.
   5) Gassing. Equipment method; quantity, rate and

TABLE 3 — SOME FORMULATION STUDIES WITH CYLINDER CORE SPECIMEN USING ROUND GRAIN OTTAWA SAND, CODE A

Exp. No.	Binder Used, No.	Binder, %	Additives, %					Compressive Strength, psi		Dietert Surf. Hardness		Flow-ability (Dietert)	Relative Humidity, %	Comments
			Sugar	Water	Iron Oxide	Silica Flour	Other	Before Heat	After Heat	Before Heat	After Heat			
104	1	1.5						279	132	75	72		Below 60	Smooth surface Dries fast
105	1	1.0						40	0				Below 60	Poor cores
106	1	2.0						361	628	77	87		Below 60	Poor collapsibility
346	1	1.8						48	54	53	47		50	Good surface, R.H. 25% when made
111	1	1.25	1.0	0.25				138	0	86	0		Below 60	Poor cores
118	1	1.8	1.5	0.375				283	14	82	57		Below 60	Smooth surface
118	1	1.8	1.5	0.375				585	8	85	65		Below 60	
118B	1	1.8	1.5					470	3	60	50		Below 60	
118C	1	1.8	2.0					612	2	41	23		Below 60	Mix dries rapidly
119	1	1.8	1.5	0.375				251	4	80	59		Below 60	Usable formula
119A	1	1.8	1.5	0.375				203	8	70	53		Below 60	lower limit of sugar
121	1	1.8	1.5	0.375				628 +	2	78	51			Cores dried in oven
121C	1	1.8	1.5	0.375				337	50	74	77		52	Repeat of 121 with CO_2
120	1	1.8	2.0		2.0			558	0	70	0		24	Good results but surface hardness
120A	1	1.8	2.0		2.0			628 +	0	60	0		28	Too low
122	1	1.8	2.0	0.5	2.0			538	0	89	0	82	38	Selected as an excellent formula
126	1	1.8	2.0	0.5	2.0			550	0	80	42		30	
126	1	1.8	2.0	0.5	2.0		Fire Clay	10	0	33	40		76	
133	1	1.8	2.0	0.5			2.0	465	0	90	55		55	Fair cores
134	1	1.8	2.0	0.375	2.0		*0.25	327	0	80	0		29	Good results
136	1	1.8	2.0	0.375			*0.25	351	2	63	38		31	Surface hards needs
137	1	1.5	2.0	0.375			*0.25	426	6	68	0		35	improvement
138	1	1.8	1.5	0.375	2.0		*0.25	628 +	0	71	0		35	
161	1	1.8	2.0	0.33			*0.2	300	12	63	3		46	
161	1	1.8	2.0	0.33			*0.2	3	3	8	10		88	
167	1	2.5	2.0	0.5	2.0	4.0		628	60	95	72		42	Excellent, very smooth,
167	1	2.5	2.0	0.5	2.0	4.0		100	185	43	74		88	good detail, see standard and 336
Standard	1	2.4	2.0	0.5	2.0	4.0		628 +	51	90				Excellent
168	1	1.8	2.0	0.5			$Ca(OH)_2$ 6.0	212		80			Room	Poor surface, too many
168	1	1.8	2.0	0.5			6.0	108					88	lumps, heat given off
258	1	3.5	2.0		2.0	4.0		628 +	94	82	75		25-45	Questionable collapsibility
258A	1	1.8	2.0	0.5	2.0	4.0		592	6	82	30		25-45	Excellent results
258A	1	3.5	2.0	0.5	2.0	4.0		628 +	425	87	70		25-45	Poor collapsibility
253	1	1.4	2.0					80	0	69	0		10-30	Sample placed in desic-
253	1	1.8	2.0					90	70	65	55		10-30	cator immediately
253	1	2.5	2.0					125	188	83	69		10-30	Poor collapsibility
253	1	3.5	2.0					430	220	72	79		10-30	Sample stood 5 hr before desiccator, poor collapsibility
256	1	1.8	2.0			4.0		130		54			10-30	
256	1	3.5	2.0			4.0		628 +	462	88	88		10-30	Poor collapsibility
255	1	1.4	2.0					215	0	46	0		40-55	Low surface hardness
255	1	1.8	2.0					262	25	50	40		40-55	
255	1	2.5	2.0					331	100	59	71		40-55	
255	1	3.5	2.0					462	230	65	72		40-55	Poor collapsibility
340	1	6.0	2.0					628 +	628 +	95	100		40-55	
257	1	1.8	2.0			4.0		91	0	60	0		40-55	
257	1	3.5	2.0			4.0		628 +	417	88	81		40-55	Poor collapsibility
339	1	2.7	2.0			4.0		550	67	78	56		40-55	Acceptable results
136A	1	1.8	2.0	0.5				467	78	81	71		52	
122D	1	1.8	2.0	0.5	2.0			487	0	84	0		52	Good results
156	8	3.5	2.0	0.5				80	0	82	32		45	Poor surface
156	8	3.5	2.0	0.5				30	2	53	27		88	
158	8	2.5	2.0	0.5	2.0			180		93			45	Good all-around core
158	8	2.5	2.0	0.5	2.0			10		50			88	
158A	8	2.5	2.0	0.5	2.0			180	0	93	0		45	
158A	8	2.5	2.0	0.5	2.0			11	0	61	0		76	
159	8	2.5	2.0	0.5		4.0		275	3	93	45		50	
159	8	2.5	2.0	0.5		4.0		**30	3	69	45		88	
162	8	2.5	2.0	0.33			*0.2							Impossible to remove from core box
163	8	2.5	2.0		2.0			450	0	95	0		45	Good surface
163	8	2.5	2.0		2.0			23	0	40	0		88	
164	8	2.5			2.0									Dries too rapidly, no good

(continued on next page)

TABLE 3 — CONTINUED

Exp. No.	Binder Used, No.	Binder, %	Additives, % Sugar	Water	Iron Oxide	Silica Flour	Other	Compressive Strength, psi Before Heat	After Heat	Dietert Surf. Hardness Before Heat	After Heat	Flow-ability (Dietert)	Relative Humidity, %	Comments
163	8	2.5	2.0		2.0	4.0		628+	0	90	0		47	Excellent formula, see 159, 172, and 218
165	8	2.5	2.0		2.0	4.0		*73	0	58	0		88	Excellent cores
166	8	2.5	2.0	0.5	2.0	4.0		518	0	94	10		38	Fair surface, good formula
166	8	2.5	2.0	0.5	2.0	4.0		**25	0	63	0		88	
157	6	2.5	2.0	0.5				500	0	80	40		45	Good
157	6	2.5	2.0	0.5									88	No good, no strength

*Cresol-phenolic organic derivative.

**Show some resistance to high humidity.

time of distribution into the piece gassed. Extent of gelling versus drying by CO_2 or other gases.

and to a lesser degree the following require control:

1) Mixing, methods and time.
2) Storage time of mix.
3) Storage time of gassed piece.
4) Method of forming core or mold.
5) Temperature at the time of mixing, gassing and storage.

The standardization of most conditions which would cause variance allowed a more specific study of those variables which are most important. These are considered:

a) Variations of sand.
b) A determination of optimum choice of sodium silicate binder.
c) An attempt to reduce the deleterious effect of high relative humidity by the use of additives or sprays.
d) The establishment of additives which can assist and enhance the binder properties of the sodium silicate.
e) The effect of controlled water addition to the original mix related to the sand silicate system used.

Minimum Critical Percentage of Sodium Silicate

Early in the experimental work, together with information from outside technologists in this field, it seemed apparent that additives such as clay, pitch, sea coal and foundry syrup contributed to better collapsibility, but simultaneously reduced the bonding values of the sodium silicate used. The laboratory evaluation method uses cylinders of various formulation heated to a high temperature such as 2200 F for a determined time in a muffle furnace. After cooling, the compressive strength of these cylinders seems a good indication of core collapsibility in plant operations.

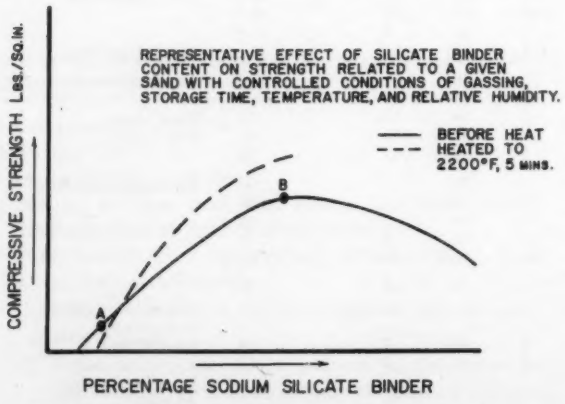


Fig. 2 — Typical curve for compressive strength related to weight percentage of sodium silicate used.

Weight Per Cent of Sodium Silicate

Figure 2 shows a typical curve, some of which have appeared in the literature, for compressive strength related to weight percentage of sodium silicate used.[16,17] The solid line represents the strength of the core or mold before metal is poured. Point *B* would appear to be the optimum amount of sodium silicate to give best surface hardness and handling strength. We have to drop back to point *A* on the curve, however, to obtain low strength after heating, in other words, attain collapsibility. Increased amounts of sodium silicate beyond this point only increase the glass bond at high temperatures resulting in a hard ceramic formation, shown as the dotted line.

In agreement with the discussion concerned with the theory of adhesives, it was realized that there is an optimum quantity and adhesive film thickness for each sand under consideration. The number of contacts between sand grains contributing structural strength is also important.

Work with the optimum quantity *B* of sodium silicate requires moving the dotted line a considerable distance to the right to assure good collapsibility. This requires additional quantities of modifying binder materials such as clay and aluminum oxide resulting in only reduced strength for handling the initial cores or molds. The studies have shown that the best approach was to consider the quantity *A* of sodium silicate binder, which we will call the minimum critical percentage of sodium silicate, which gives some strength and yet allows collapsibility.

Figure 3 shows such curves with no additives used for five different sands. The differences in silicate binder requirements and the resulting collapsibility among various sands is quite marked. The physical characteristics of the sand in measurable parameters is most important in considering sodium silicate

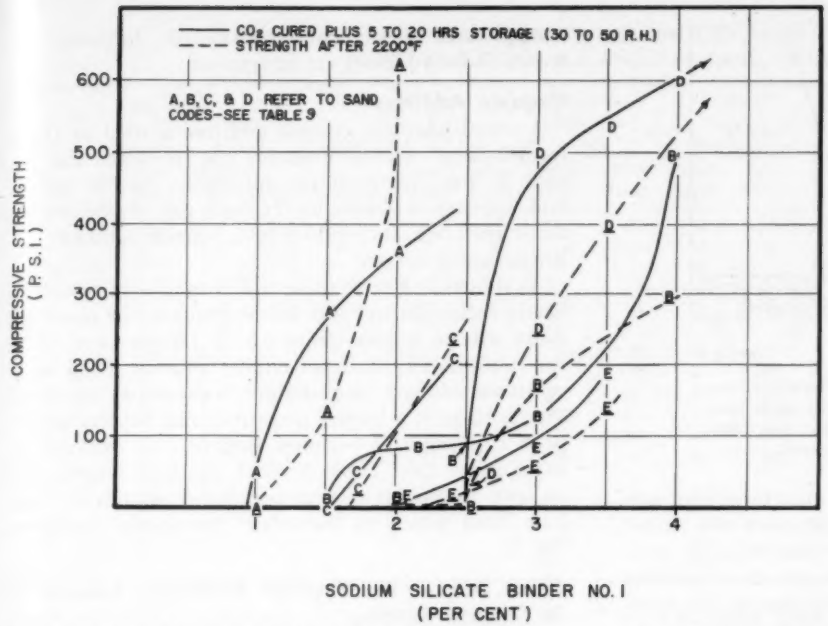


Fig. 3 — Relationship between compressive strength and per cent silicate binder (no additives) for five different sands, before and after high temperature.

bonding formulations (Table 9). The number of bond contacts, the area of these contacts and the condition of the sand surface are apparently most important, and substantiate the concepts of bonding presented here.

Additives

The favorable additives listed previously used in reasonable percentages will add considerably to the strength, ramability and surface hardness if used with the lower *A* percentage of a sodium silicate such as no. 1 which is a 2 to 1 ratio silica to soda, 45 per cent total solids, 51 degree Baumé, or the *A* amount plus approximately 0.5 to 1.8 per cent more silicate depending on the sand silicate system. Both sugar and iron oxide assist the collapsibility allowing the additional amount of binder above Point *A*. Consider Figure No. 4 using the round grain sand Code No. A and the no. 1 silicate binder.

The critical percentage of sodium silicate in this system is about 1.1 per cent. The addition of 2 per cent sugar, as well as the addition of 2 per cent sugar with 4 per cent of a coarse 90 to 140 mesh silica flour, moves the collapsibility curve almost 1 per cent silicate to the right. Figure 5 presents convincing evidence of the additional value of adding 2 per cent iron oxide and ½ per cent water. Water addition should be made with care, since the dilution of sodium silicate can result in apparently good initial gas strength but reduced strength with storage. In this case, however, the water seems to enhance the value of the sugar silicate adhesive system.

The value of adding silica flour depends upon the

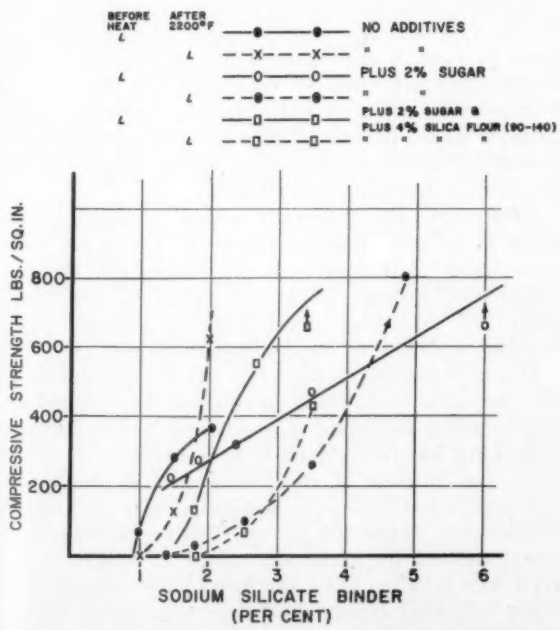


Fig. 4 — Effect of adding sugar and silica flour on strength and collapsibility of a single sand-silicate binder combination (sand A and Binder No. 1).

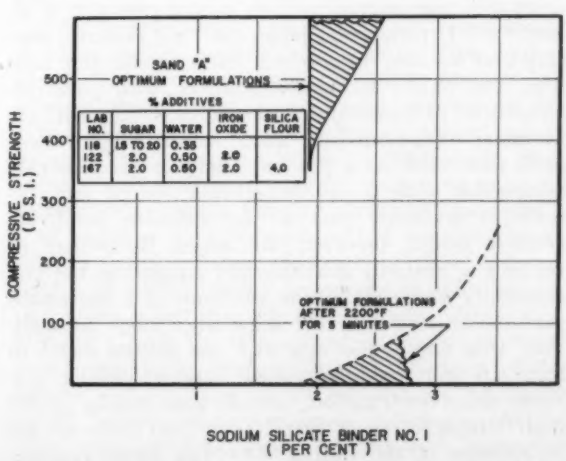


Fig. 5 — Evidence of the additional value of adding 2 per cent iron oxide and ½ per cent water.

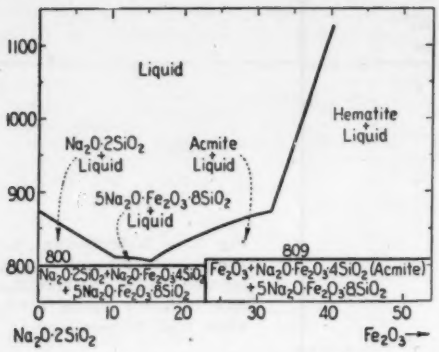


Fig. 6 — Phase diagram showing the possible compositions depending upon percentages of ingredients included and temperatures (deg. Cent.) considered.

particle distribution system of the sand used. Fine silicate flour such as 325 mesh in most cases will reduce the properties of the bond rather than enhance them. The coarser flours on the other hand act as building blocks between the sand grains increasing the number of adhesive bonded contacts.

The silicate without sugar is either too weak in the quantities we recommend for proper collapsibility after metal pour, or becomes brittle with poor properties of toughness. The tensile or compressive strength, however, in an all silicate bond can be quite high especially if the quantities are excessive compared with the recommended minimum critical percentage. Such a bonding adhesive, however, develops many stresses as the silicate dries out and shrinks pulling away from the sand contacts.

This more rigidly bonded structure is in direct opposition to the argument for cushioning additives such as sugar, which eliminate mold and core surface fractures. The sugar contributes strength allowing a lower amount of sodium silicate, and initially retains moisture which contributes to the flexibility and toughness of bond structure.

Clay

Additives such as clay interfere with the chemistry of the silicate bond, and must be used with care because they agglomerate in such a high pH (highly alkaline).[30] Experimental plant trials are evidence that clay may be used to produce fairly satisfactory castings. The results are spotty, however, with poor collapsibility predominate, and indicate the difficult problems of control in a questionable system. However, iron oxide as a powdered additive is evidently compatible with a sodium silicate system and becomes a favorable part of the adhesive bond. Its greatest values, however, are added flowability in forming a uniform mixture, and enhancing the collapsibility which allows the additions of a few tenths per cent silicate with the sugar-silica flour formulation. Iron Oxide also acts with the molten metal to form an iron oxide-soda-silica complex which prevents metal penetration, retards core cracks at the high temperatures of metal pour and adds to the smoothness of the casting.[31,32] The phase diagram of Fig. 6 shows the possible compositions depending upon percentages of ingredients included and the temperatures (deg. cent.) considered. Liquids, eutectics and solid mixes are designated.

Organic Additive

A cresol phenolic organic additive is used in these formulations in small amounts not to exceed 0.2 per cent. It has enhanced the ramability of the mixes, and appears to form just enough gas at the metal-mold interface and results in a smooth, almost polished, casting surface.

As shown in Fig. 5, these additives all contribute to better collapsibility, and allow additions of more sodium silicate binder (from 0.5 to 1.5 per cent) than the minimum critical percentage allowed. Even with optimum additive formulations, however, as shown in Fig. 5, there is a binder quantity limit beyond which poor collapsibility becomes evident. An example is experiment 256, Table 3. Each sand, of course, has its own minimum critical percentage of silicate as a base from which to formulate, previously shown in Fig. 3.

Silica Sand — The Physical Variations Related to Silicate Bonding

The first laboratory and plant investigations of the sodium silicate-CO_2 cured process repeatedly emphasized the inability to formulate with a variety of different sands to attain acceptable molds and cores. It seemed important at this time to study the properties of the sand related to bond requirements considering the sodium silicate as an adhesive bonding system. The work was carried out in two approaches simultaneously, correlating the results wherever possible. The first is the laboratory evaluation method which was devised using cylindrical cores with all variables controlled including one selected sodium silicate bonding material, but varying the quantity used with the type of sand being evaluated.

Figure 3 is an example of plotted data determined by this approach. A code description of these sands is given in Table 9 and includes data established by the second approach, that of measuring all the known parameters of the sand and obtaining numerical comparisons where possible. There has been a fairly successful correlation of the silicate binder needs and the measured characteristics of the sand.

Arbitrarily, the minimum critical percentage was taken as the percentage of binder no. 1 with no additives which will contribute approximately 100 psi compressive strength to the test cylinders, simultaneously not exceeding a collapsibility strength after heating at 2200 F of approximately 60 psi.

It was impossible to meet this latter requirement in the case of sand *C*, since its collapsibility follows almost parallel and equal to its strength before heating. This sand was recommended as satisfactory by other users, but because of these data sand *B* similar in description but with a collapsibility after heat curve considerably below the strength before heat curve is recommended by preference. Sand *E*, which has a high AFS number, has a problem of the collapsibility curve following closely the before heat curve. Figure 5 shows how additives are used to separate these two curves for any sand binder combination.

This is accomplished, however, much more easily

TABLE 4 — SOME FORMULATION STUDIES WITH CYLINDER CORE SPECIMEN USING FINE GRAIN SUB-ANGULAR SAND, CODE E

Exp. No.	Silicate Binder Used, No.	Binder, %	Additives, %				Compressive Strength, psi		Dietert Surf. Hardness		Flowability (Dietert)	Relative Humidity, %	Comments
			Sugar	Water	Iron Oxide	Silica Flour	Before Heat	After Heat	Before Heat	After Heat			
169	1	3.0	2.0	0.5			212	68	92	80		Room	Good surface, poor collapsibility potential
169	1	3.0	2.0	0.5			4	125	51	78		88	
A1AQ(13)	1	2.5	2.0						45	27-52			Low strength
275	1	1.5											Too dry
349	1	1.8	2.0	0.5	2.0	4.0	271	0	70	0		50	Good surface
271	1	2.0											Too dry
272	1	2.5					50	0	66	0		41	
348	1	2.5		0.5			12	3	77	68	82	50	Fair surface
306	1	2.5	2.0	0.5	2.0	4.0	250	56	70	0		55	Usable for operation
273	1	3.0					92	67	79	70		41	Sticks slightly
274	1	3.5					196	150	85	70		41	Hard to work with
350	1	3.5	2.0	0.5	2.0	4.0	258	144	82	70		50	Excellent surface
276	8	3.0											Not workable
277	8	3.5											Not workable
278	8	4.0											Very fragile
170	8	4.5	2.0				163		94			Room	Difficult to remove from mold
170	8	4.5	2.0				80		80			88	

with some sands than with others. The strength before heat curve is raised considerably at any given percentage binder, and the collapsibility curve is reduced moving to the right and requiring higher percentages to show the undesirable strength after heat. Sand *A* is a well distributed round grain sand that has a low specific surface and appears ideal for the sodium silicate-CO_2 cured system.

Noncontrolled Addition Effect

Promiscuous addition of sodium silicate to such a sand can cause difficulty in collapsibility and hot casting cracks, because the mold cannot crush or move to accommodate shrinkage of the metal.[11,13] Using a well controlled percentage of sodium silicate together with additives, such as sugar and iron oxide, hot rigidity of the mold is reduced, obviating problems of hot casting cracks, rattail and buckle.

Sand *E*, on the other hand, with a high AFS fineness number of 158, a five screen distribution and a measured low-packed density, evidently would tend to reduce the contacts for adhesive bonding and indicate a desire to add greater amounts of silicate for bonding strength.[9,10] However, as the curve in Fig. 3 shows, the strength is raised only slowly with the addition of more silicate.

Experimental studies, Table 4 experiments 349, 306, and 350 substantiate the belief that coarse flour 90 to 140 mesh addition to such a sand as sand *E*, filling in the voids and greatly increasing the number of bond contacts, enhances the use of this sand for the CO_2 molding process. Due to limited experimental time the optimum addition of silica flour for sand *E* has not been determined. The following formula is proposed—3½ per cent no. 1 silicate binder, 2 per cent sugar, ½ per cent water, 2 per cent iron oxide and 8 per cent silica flour 90 to 140 mesh.

Although sand *D* is a controlled three screen sand, it contains considerable impurities mostly of a micaceous nature, which because of its laminate form has an excessive specific surface, this being unit area per unit weight presented as sq cm/gram.[6,7,8] The addition of silica flour to sand *D* does not improve the bond contacts. The micaceous inclusions evidently interfere with this possibility and otherwise contribute in this respect.

The method of calculating specific surface by the use of sieve analysis is well presented.[2] For clean washed sands, and using proper ratio multipliers dependent on the type sands—round grain, sub-angular and angular, these methods seem quite comparable and indicative of the actual specific surface.

Laboratory Methods

There are other highly developed laboratory methods, however, which measure the amount of gas in molecular layers, such as argon, adsorbed on all parts of the surface of any powdered material. This method devised by Brunauer, Emmett and Teller called the B.E.T. method was used as a further evaluation on these five sands for comparative specific surface area.[8] Sand *D*, Table 9, has a considerably higher B.E.T. specific surface compared with its sieve analysis evaluation for specific surface, 810 sq cm compared with 315.

This comparison is reversed in the case of sand *E* where specific surface calculated by sieve analysis shows 432 sq cm/gram, and the B.E.T. Method 152. Sand *A* shows such a low specific surface by the B.E.T. method that it approaches the area of perfect spheres. These anomalies show the difficulty of using any one measured parameter for comparing one sand to another to predict foundry binder requirements.

In addition the minimum critical requirements of various sodium silicates for each of these different sands add to the further knowledge of their properties. Using a low viscosity 60 centipoise silicate such as no. 8 (Table 10), it is almost impossible to attain good strengths by using additional quantities of this silicate with *D* sand which has a high adsorptive power because of its micaceous inclusions. On the other hand, sand *D* can be bonded with reasonable quantities of a viscous binder of 380 to 2000 centipoise.

With some water even the high viscosity no. 12 binder of 70,000 centipoise can be used in low percentages of about 2 per cent for adequate bonding of this sand. The viscosity range of 380 to 2000 appears an overall best for a variety of sands such as the five considered in Table 9. Other properties of the silicate must be taken into consideration, however, and are described in more detail in a later section.

Figures 7a and 7b are an attempt to correlate graphically the minimum critical percentage of a sodium silicate, such as the no. 1 silicate, required by

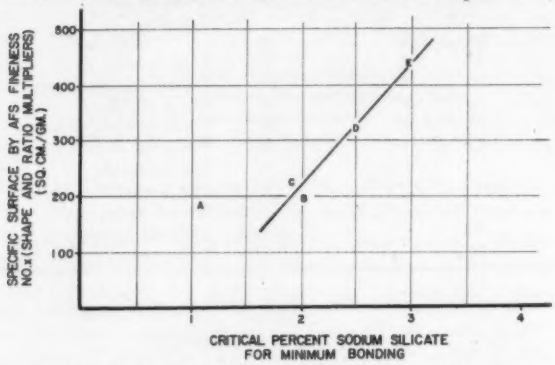


Fig. 7a — Sands *B, C, D* and *E* show a general relationship between the (AFS number) × (shape and ratio multipliers) and the critical percentage requirement of sodium silicate no. 1 for minimum bonding. Sand *A* does not satisfy this correlation.

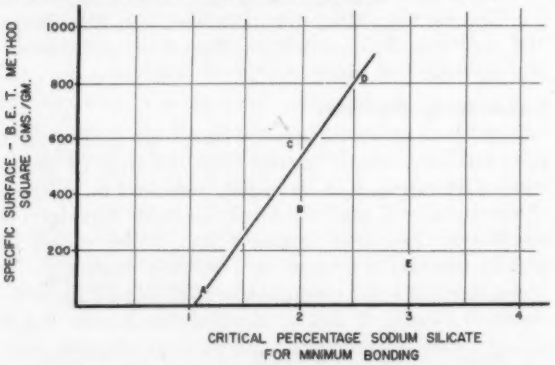


Fig. 7b — Sands *A, B, C* and *D* show a general relationship between specific surface by the B.E.T. method and the critical percentage requirement of sodium silicate no. 1 for minimum bonding. Sand *E* does not satisfy this correlation.

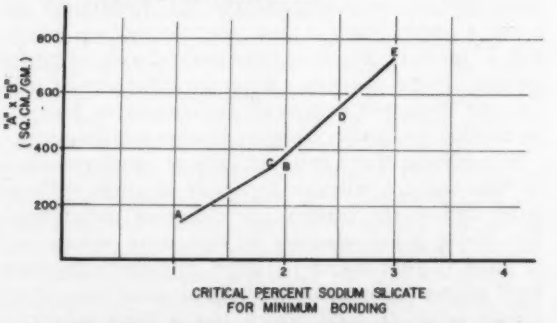


Fig. 7c — Plot showing a correlation for the factors *A* × *B* with the critical percentage requirement of sodium silicate no. 1 which satisfies all sands. *A* is (AFS no.) (est. surface shape factor) (ratio multiplier). *B* is (ratio representing specific volume differences).[4]

the five sands with their specific surfaces calculated by the AFS number times its shape and ratio multipliers, as well as by the B.E.T. method. As such, if correlation could be devised, initial silicate formulations could be predicted within a close range with a minimum of experimental work.

There is a fairly good general relationship for four of the five sands in each case—sand *A* being out when using the AFS fineness times the multiplier method, and sand *E* being particularly out of correlation with the B.E.T. method. By bringing another parameter into this, that of specific volume, we have developed a correlation shown in Fig. 7c by which all five sands correlate with this minimum percentage of sodium silicate for critical bonding. Variations in specific volume, a figure representing volume per unit weight, were determined with the test cylinders. The same weight of 185 grams gave variations in height of cylinder due to differences in sand as follows:

Sand Sample	Height Variations, in.
A	1.89
B	2.06
C	2.00
D	2.08
E	2.15

Specific Volume Comparison

The lowest common denominator of these height differences is a ratio comparison of their specific volume differences. This number, taken to the fourth power called *B*, is included with the other data in Table 9. Figure 7c shows the relationship of this number times the AFS number with its multipliers designated as *A*, plotted against the critical per cent sodium silicate for minimum bonding. There is a surprisingly good correlation which may be used as a guide for predicting this minimum bond requirement for any sand selected. There is no mathematical explanation for taking this ratio of specific volumes to the fourth power.

Hypothetically, we can consider those sands with high specific volume (indicating greater void content) as having a reduced number of contacts for structural bonding within a given volume. Thus, greater percentages of sodium silicate per unit weight would be required to give structural strength. As pointed out previously, however, it is better to fill these voids with fine grain sand such as the coarse silica flour which are large enough to act as structural building blocks contributing a considerable increase in number of bonding points.

Any one of these numerical parameters evaluates some characteristics of the sand considered, each contributing to the approach which may be used in formulating optimum use of sodium silicate and additives related to the molten metal poured against the mold surface. Practical results of these studies are presented in a later section. It is noted that each sand requires differences in formulation to obtain optimum properties in the cores and molds produced. Promiscuous use of proprietary formulas, expected to give equal performance on any sand chosen, seems unreasonable considering the results of these studies.

Proper cooperation between the formulator and the

foundry technologist related to the sand used, metal poured, size of castings and end use requirements should give optimum performance provided the formulator is aware of the changes necessary between one sand and another. The formulator has equipment for proper blending of all ingredients, saving the foundry the problem of obtaining a homogeneous mix at the time of make-up with the sand.

The attempt to numerically describe the characteristic difference in sands has been valuable from a practical use interpretation, but does not provide the establishment of theoretical concepts which can be used as a guide to predict the optimum sand for the CO_2 process. Reference papers[9,10] provide an excellent background for this type of thinking, but point out the limitations of the mathematical approach related to actual sand performance in foundry usage. Perhaps the data included here will contribute to later studies of a more mathematical effort.

THE SODIUM SILICATE BINDERS

Table 10 is a compilation of published properties of the 12 sodium silicate binders under in these studies. The no. 1 binder has been used predominantly in the study of other variables related to this CO_2 process. Comparatively, binders of this description show the best overall results and have been recommended with all the sands under consideration. Numerous references have presented technical, theoretical descriptions of the various silicates produced commercially.[16,18,19,21,29] Total solids content, viscosity and density difference given in Baumé are all easily understood properties which relate to flowability and actual dried binder qualities. The less understood variables are the ratio of soda to silica and the effect of water quantity in the chemical reaction with carbon dioxide to form carbonic acid, sodium hydrogen carbonate and the binder ingredient silica gel. These affect the rate of reaction with the carbon dioxide, the type of gel formation, the subsequent resistance to moisture pickup and final binder strength.

Silica to Soda Ratio

The silica to soda ratio, as well as the water content determined by the total solids, is of particular importance in determining the character of the final silicate adhesive bond. A strong, dry, brittle adhesive bond results where the silica content is the greatest and the soda is in lower percentages. As the soda (alkaline) percentage increases, there is a tendency to retain additional moisture in the system resulting in a more flexible film adhesive.

Gassing

The extent of CO_2 gassing can also vary the final properties of the silicate bond. The gassing mechanism consists in the formation of carbonate ions in the water system which form hydrated sodium bicarbonate in various complexes depending upon the amount of CO_2 injected, the water content and the temperature of the system. In addition to the water held chemically by the sodium carbonate, the hydrogen ions lower the pH causing the formation of silica gel which takes up water.

The greatest contribution of strength is through the residual sodium silicate which has been dried out by the removal of water in hydrating the sodium carbonate and in forming the silica gel. As the silica gel and hyrated sodium carbonate lose their water in storage, they lose strength, but they do contribute to the initial strength established by the gassing of CO_2.

Water Content Control

High water contents, together with continued gassing, finally deplete the sodium silicate as it is hydrated to silica gel and the hydrated sodium carbonate. It is important, then, to control the water content carefully and/or control the gassing time. If cores or molds are used immediately, this reduction in strength due to drying out of the silica gel and hydrated sodium carbonate is of less importance since this occurs generally in storage.

High relative humidity contributing continued moisture prevents the drying out of the sodium silicate to form the adhesive bond, and usually slows down the whole reaction as the highly viscous silica gel prevents permeation of the CO_2 gas through the sand sodium silicate system. Since there is strength contributed by the silica gel and the hydrated sodium carbonate, if the system is not dried out but retains some moisture there can be a final strength contribution by all three components. This results in a flexible system which is not completely dehydrated.

A moisture tempering of the bond, then, provides flexibility and toughness. Additives such as sugar assure the presence of moisture, and also provide a favorable dilution and flowability of the silicates during mixing. This property of low viscosity flow provides a thin adhesive layer, uniformly distributed over the sand grains and the coarse silica flour. The greater the thickness of this bond, the more chance for breakdown in the adhesive layer.

Sugar Addition

The control of water content for retaining such a flexible system is more difficult where ambient conditions vary from low humidities to high relative humidities. The rate of change and character of the hydrated components (silica gel and sodium bicarbonate) can be greatly improved by the addition of sugar, either in granular form or as a liquid syrup. This hydrophilic additive stabilizes the gel structure contributing strength and preventing rapid CO_2 gassing effects as well as rapid removal of moisture in dry conditions. There are many hypotheses concerning gel structures.

Some believe the structure predominately amorphous, and others believe them composed of myriads of fine needle-like crystals which vary in flexibility and elasticity. In any case, the sugar contributes a flexible toughness to the binder system which allows the use of lower percentages of the less flexible inorganic sodium silicate ingredient.

Experimental tests have shown comparatively poor results for invert sugar and foundry *A* syrup. Table sugar, uniformly distributed into the mix together with the sodium silicate and other additives, gave good results.

Any gel system can be deleteriously affected by

TABLE 5 — SOME FORMULATION STUDIES WITH CYLINDER CORE SPECIMEN USING SUB-ANGULAR HIGH DENSITY AFS 81, CODE C

Exp. No.	Silicate Binder Used, No.	Binder, %	Additives, %				Compressive Strength, psi		Dietert Surf. Hardness		Flowability (Dietert)	Relative Humidity, %	Comments
			Sugar	Water	Iron Oxide	Silica Flour	Before Heat	After Heat	Before Heat	After Heat			
239	1	2.0					125	108	56	58	80	58	Poor collapsibility
240	1	2.4					206	234	64	68	84	58	Poor collapsibility
243	1	2.7				4.0	100	35	51	58	81	58	120 mesh silica flour
244	1	2.7				4.0	95	80	55	45	82	58	90 mesh silica flour
245	1	2.7				4.0	100	100	59	46	82	58	120 mesh silica flour
246	1	2.7				4.0	135	100	64	62	85	58	200 mesh silica flour
247	1	2.7	2.0	0.5		4.0	156	137	78	72	82	58	Stand 5 hours
248	1	2.4	2.0	0.5		4.0	473	20	81	59	79	40	Stand 29 days, acceptable formulation
249	1	2.4	2.0	0.5	2.0	4.0	250	0	72	0	83	40	Stand 29 days, acceptable formulation
254	1	2.0	2.0	0.5		4.0	104	0	72	0	80	40	Stand 23 days, acceptable formulation
259	1	1.5										77	Not workable
260	1	1.7					48	23	53	43		56	
261	1	2.0				4.0	20	45	59	37		56	(Low strengths)
242	8	2.7									78	58	Too soft, no strength

changes in pH or the introduction of incompatible electrolytes. The use of additives with such an adhesive system is quite critical and accounts for the varied results with clays,[30] pelleted pitch, aluminum oxide, etc. With certain controlled formulations these may prove useful, but experience indicates that it would require considerable experimental effort to determine such optimum compositions.

Relative Humidity and Moisture Content

Minimum quantites of properly distributed bonding materials are considered an advantage. We are still faced with the problem, however, of almost complete breakdown of the sodium silicate bond under high relative humidity conditions. The absorption and adsorption of water into the system reduces the cohesive strength of the bond to almost nothing. We are faced with a favorable inclusion of some moisture and a considerably unfavorable result from excess moisture as absorbed by the humectant sugar-silica system.

Tables 1 to 7 show a convincing number of formulations containing from 1.8 to 2.5 per cent sodium silicate providing all the qualities of best CO_2 molding, provided the relative humidity is kept below approximately 70 per cent. A real effort has been made to obviate this problem of relative humidity. Favorable resistance to high humidity has been obtained by removing the sugar, but with the immediate return to the problem of poor collapsibility, reduced resistance to surface rubbing breakdown and poor properties of flow ramability.

Some systems have been found (Table 8) which are partially resistant to high humidities, the best being a surface spray with a silicone solvent system. Other data in this table show the possibilities for air-drying and the questionable use of calcium hydroxide. This material causes a rapid reaction which would be difficult to control. Many other materials were experimentally tried as additives with no benefit. Suppliers of sodium silicate suggest no possible method for alleviating this degradation with high humidity.

Continuous storage in high humidity can do nothing but degrade the surface hardness of a core or mold and eventually cause crumbling of the structure itself. It has been found, however, that the system is reversible to moisture, that is, the strength can be regained under dry conditions accelerated with a convection of dry air.

Compressive Strength and Surface Hardness Changes With Relative Humidity

Figure 8 is a bar chart showing changes in compressive strength and surface hardness with relative humidity. Two forms of sodium silicate have been considered in this evaluation of sensitivity to relative humidity. References[18,27,29] point out that silicates having high silica to soda ratios are inherently more resistant to high moisture conditions. Number 8 binder has a high ratio of 3.22 to 1 as compared to the 2-1 ratio of no. 1 binder selected for its overall good properties. The no. 8 binder also has a much lower solids content and a viscosity of 70 centipoise compared with 380 for the no. 1 binder of 2 to 1 ratio.

The experimental results plotted in the bar chart (Fig. 8) show a considerable difference between compressive strength and surface hardness for the binder. Relative humidities above 60 per cent reduced the compressive strength in every case. Although the no. 8 binder at the lower humidities gives less than half the strength of the 2 to 1 ratio silica to soda when used with the round grain *A* sand, the strength is reduced almost to an unusable condition when used with *D* sand. In surface hardness, however, the results are reversed at the higher 88 per cent relative humidity.

Compressive strength and surface hardness data evidently do not correlate with each other, but must be related to quantity of water present and the properties of the sand used. There are many other data in the experimental studies which further establish that the surface hardness can be enhanced with the addition of some moisture. Continued exposure to

TABLE 6 — COMPARATIVE EVALUATION OF VARIOUS SODIUM SILICATE TYPES WITH CYLINDER CORE SPECIMEN

Exp. No.	Sand Used	Silicate Binder, No.	Binder, %	Additives, %				Compressive Strength, psi		Dietert Surf. Hardness		Flow-ability Dietert	Rel. Hum., %	Comments
				Sugar	Water	Iron Oxide	Silica Flour	Before Heat	After Heat	Before Heat	After Heat			
104	A	1	1.5					279	132	75	72		Below 60	Smooth surface
122D	A	1	1.8	2.0	0.5	2.0		487	0	84	0		52	
167	A	1	2.5	2.0	0.5	2.0	4.0	628	60	95	72		42	Excellent
167	A	1	2.5	2.0	0.5	2.0	4.0	100	185	43	74		88	
239	C	1	2.0					125	108	56	58	80	58	
249	C	1	2.4	2.0	0.5	2.0	4.0	250	0	72	0	83	40	Stand 29 days
259	C	1	1.5										77	Not workable
272	E	1	2.5					50	0	66	0		41/55	
306	E	1	2.5	2.0	0.5	2.0	4.0	250	56	70	0		54	Usable for operation
274	E	1	3.5					196	150	85	70		41/55	Hard to handle
A2A	D	1	3.0					500	250	80	68		Room	Poor collapsibility
174	D	1	3.0	2.0	0.5			500	1	77	45		56	
174	D	1	3.0	2.0	0.5			2	2	0	57		88	
204	B	1	2.5					77	15	56	57		Normal	Good surface
204	B	1	2.5					5	20	0	63		88	
252	B	1	2.5	2.0	0.5	2.0	4.0	400	0	80	0		52	
345	D	3	3.0	1.5	0.3	2.0	*0.2	93	0	50	0		55	Gassed 10 S at 35 psi
335	B	3	3.0	2.0	0.5	2.0	4.0	180	0	56	0		50	Gassed 10 S at 35 psi
285	A	3	1.5					55	0	0	0	88	51-34	Very poor surface
311	A	3	1.8	2.0	0.5	2.0	4.0	245	0	60	0	82	48-55	Gassed 10 S at 35 psi Gassed 10 S at 35 psi
344	A	3	1.8					52	0	47	0	87	58	Fair surface
336	A	3	2.4	2.0	0.5	2.0	4.0	563	0	75	0		55	Gassed 10 S at 35 psi
296	C	4	2.8					66	0	39	0		45-39	Gassed 10 S at 35 psi
318	C	4	3.0	2.0	0.5	2.0	4.0	120	0	60	0	73	55	Gassed 10 S at 35 psi
286	A	4	1.5					6	0	0	0	87	51-34	Gassed 10 S at 35 psi Very poor surface
312	A	4	1.8	2.0	0.5	2.0	4.0	520	0	75	0	78	55	Gassed 10 S at 35 psi
345	A	4	1.8					12	0	23	0	85	50	Poor surface, 10 S gassing
291	B	4	2.4					55	11	49	20	78	43	Gassed 10 S at 35 psi
315	B	4	2.4	2.0	0.5	2.0	4.0	55	0	52	0	83	55	Gassed 10 S at 35 psi
300	D	4	2.5										55	Too dry
320	D	4	3.2	2.0	0.5	2.0	4.0	250	0	71	0		55	Gassed 10 S at 35 psi
305	E	4	3.0					12	0	32	0		55	Gassed 10 S at 35 psi
279	A	9	2.0										55	Too dry
307	A	9	2.0	2.0	0.5	2.0	4.0	362	0	71	0	80	53	
233	A	9	2.5	2.0		2.0	4.0	2	0	25	0		80	Sticky, lumpy
233	A	9	2.5	2.0		2.0	4.0	2	0	0	0		88	
235	B	9	2.3											Poor results
301	E	9	3.5											Too dry
292	C	9	3.2					2.5	0	0	0	80	43-34	Very poor surface
316	C	9	3.7	2.0	0.5	2.0	4.0	55	0	65	0		55-68	
281	A	10	2.5					130	220	80	80	92	46	
308	A	10	2.0	2.0	0.5	2.0	4.0	480	35	81	23	81	56	
293	C	10	3.2					71	160	84	75	83	43	
317	C	10	3.2	2.0	0.5	2.0	4.0	440	130	82	58	73	68-55	
284	A	11	1.5										54	Not workable (lumps formed)
284A	A	11	1.5		0.3			2.5	0	27	0	83	43	
310	A	11	1.5	2.0	0.9	2.0	4.0	350	4	68	0	82	55	
283	A	12	1.5											Not workable, silicate did not go into mix
303	E	12	2.5		0.5			50	0	33	0		55	No comparison
280	A	5	1.8					38	14	27	5	76	44	Poor surface
343	A	5	1.8					38	132	60	65	87	55	10 S gassing, good surface, 48 hr, stand
295	C	5	2.8					48	50	71	63	84	44	10 S gassing, fair surface, 72 hr, stand
304	E	5	3.0					19	0	50	0	84	44	10 S gassing
299	D	5	3.2					19	0	68	0	81	44	10 S gassing
288	B	5	2.6					18	0	32	0	75	44	10 S gassing
157	A	6	2.5	2.0	0.5			500	0	80	40		45	
157	A	6	2.5	2.0	0.5								88	No good
282	A	7	1.5					10	0	8	0	88	46	Surface poor, dries rapidly
309	A	7	1.8	2.0	0.5	2.0	4.0	160	0	81	23	81	55	
302	E	7	3.0					6.3	0	30	0		55	
297	D	7	2.5					6	0	25	0	82	42	
319	D	7	2.8	2.0	0.5	2.0	4.0	56	0	58	0	78	55	
289	B	7	2.4					20	0	35	0	89	35	
314	B	7	2.7	2.0	0.5	2.0	4.0	120	0	59	0	78	55	
164	A	8	2.5			2.0							45	Dries too rapidly

(continued on next page)

TABLE 6 — CONTINUED

Exp. No.	Sand Used	Silicate Binder, No.	%	Additives, % Sugar	Water	Iron Oxide	Silica Flour	Compressive Strength, psi Before Heat	After Heat	Dietert Surf. Hardness Before Heat	After Heat	Flow-ability Dietert	Rel. Hum., %	Comments
165	A	8	2.5	2.0		2.0	4.0	628+	0	90	0		47	
165	A	8	2.5	2.0		2.0	4.0	73	0	58	0		88	
158	A	8	2.5	2.0	0.5	2.0		180		93			45	
158	A	8	2.5	2.0	0.5	2.0		10		50			88	
159	A	8	2.5	2.0	0.5		5.0	275	3	93	45		50	
159	A	8	2.5	2.0	0.5		5.0	30	3	69	45		88	
242	C	8	2.7									78	58	Too soft, no strength
170	E	8	4.5	2.0				163		94			Room	Difficult to remove
170	E	8	4.5	2.0				80		80			88	from mold
276	E	8	3.0											Not workable
231	B	8	3.5	2.5			4.0	220	2	92	26		45	Dries out, but good surface
231	B	8	3.5	2.5			4.0	3	2	38	38		88	
287	B	8	3.2										46	Not workable
313	B	8	2.5	2.0	0.5	2.0	4.0	62	0	64	0	80	55	
160	D	8	4.0	2.0	0.5			22	0	93	0		50	
160	D	8	4.0	2.0	0.5			2	0	61	0		88	
176	D	8	3.5	2.0				158	0	87	0		50	Good surface
290	B	12	1.8											Not workable, silicate does not mix in
294	C	12	2.2											Silicate does not mix in
177	D	12	2.0	2.0	0.5			350	0	79	0		60	Excellent surface
177	D	12	2.0	2.0	0.5			2	0	7	0		88	
178	D	12	2.0	2.0										Too dry
298	D	12	2.0										57	Not workable

*Cresol-phenolic organic derivative

high humidity, however, will finally result in the breakdown of the surface as well as the internal structure.

A 10-in. 90 deg. elbow core formed of formulation 165 (Table 3) was placed in an open structure in the rain for more than 24 hr, with reductions of surface hardness from 65 to 60 resulting in a final core reasonably strong enough for a metal pour. This formulation using the no. 8 silicate under normal conditions of humidity, however, is not equivalent in overall properties with other selected formulations using the 2 to 1 ratio silicate, and should be used only under high humidity conditions to give fair casting results. A solvent carried silicone spray, applied over the surface on the CO_2 molded core, further reduced the penetration of moisture (Table 7, experiments 135, 145).

A convincing argument is given in literature[18] with supporting data, to keep at a minimum the initial strength contributed by the CO_2 gassing. The strength attributed to the CO_2 silica reaction is in the form of a silica gel. Subsequent drying removes the gel formation resulting in a complete breakdown of the strength initially contributed by the gel. On

TABLE 7 — COMPILATION OF SPECIAL STUDIES

Exp. No.	Sand Used	Silicate Binder, No.	Binder, %	Additives, % Sugar	Water	Iron Oxide	Silica Flour	Other	Compressive Strength, psi Before Heat	After Heat	Dietert Surf. Hardness Before Heat	After Heat	Rel. Hum., %	Comments
129	A	1	1.8	2.0	0.5	2.0			600	0	90	0	30	Gassed with CO_2 20 S at 35 psi then 30 S at 35 psi with N_2
129	A	1	1.8	2.0	0.5	2.0			0	0	0	0	88	
118F	A	1	1.8	1.5					585	209	62	40	28	Gassed with compressed
118F	A	1	1.8	1.5					189	186	70	52	76	air 20 S at 35 psi
118G	A	1	1.8	1.5					566	5	52	47	29	Gassed CO_2 and com-
118G	A	1	1.8	1.5					200	0	70	60	76	pressed air 20 S at 35 psi
121A	A	1	1.8	1.5	0.375				322	10	68	33	31	Gassed with compressed
121A	A	1	1.8	1.5	0.375				5	110	10	58	76	air 20 S at 35 psi
121B	A	1	1.8	1.5	0.375				478	6	73	42	31	Gassed CO_2 and com-
121B	A	1	1.8	1.5	0.375				110	8	40	48	76	pressed air 20 S at 35 psi
135	A	1	1.8	2.0		2.0			504	0	75	0	29	Sprayed with a solvent
135	A	1	1.8	2.0		2.0			38	0	55	0	88	carried silicone spray
123	A	1	1.5	2.0		2.0			0	0	0	0	76	Sprayed with a water
123	A	1	1.5	2.0		2.0			0	0	0	0	88	containing silicone spray
145	A	1	1.8	2.0	0.5				325	16	80	60	60	Cores dipped in 12%
145	A	1	1.8	2.0	0.5				200	37	48	60	88	hydrous magnesium silicate methanol solution
146	Noncode round grain	1	1.8	2.0	0.5				465	6	98	82	54	Dipped in aluminum
146	Noncode round grain	1	1.8	2.0	0.5				0	5	2	90	88	hydrogen phosphate

(continued on next page)

TABLE 7 — CONTINUED

Exp. No.	Sand Used	Silicate Binder, No.	Binder, %	Additives, %					Compressive Strength, psi		Dietert Surf. Hardness		Rel. Hum., %	Comments
				Sugar	Water	Iron Oxide	Silica Flour	Other	Before Heat	After Heat	Before Heat	After Heat		
155	Noncode round grain	8 and Noncode binders	2.0	2.0	0.5				563	0	79	22	45	
155	Noncode round grain	8 and Noncode binders	2.0	2.0	0.5				2	0	7	40	88	
201	A	Noncode binders	3.0	2.0		2.0	4.0		154		85		Normal	Very dry, fair surface
201	A	Noncode binders	3.0	2.0	0.5	2.0	4.0		2		0		88	Very dry, fair surface
202	D	Noncode binders	3.5	2.0	0.5	2.0			179		85		Normal	Dries fast
202	D	Noncode binders	3.5	2.0	0.5	2.0			2		8			
126	A	1	1.8	2.0	0.5	2.0		0.25[a]	550	0	80	42	Room	
126	A	1	1.8	2.0	0.5	2.0		0.25[a]	8	0	30	40	76	
127	A	1	1.8	2.0	0.5	2.0		1.0[b]						No strength
128	A	1	1.8			2.0		1.0[b]						No strength
141	A	1	1.8	2.0	0.5	2.0		2.0[c]						Too soft
168	A	1	1.8	2.0	0.5			6.0[d]	212		80		Room	Heat given off
168	A	1	1.8	2.0	0.5				108				88	
133	A	1	1.8	2.0	0.5		Fire Clay 2.0		465	0			55	
140	A	1	1.8		0.5	2.0		2.0[e]						Too weak
153	Noncode round grain	1	1.8	2.0	0.5			4.0[f]						Initial gas form an impervious coating stopping effectiveness of gas on center core
154	Noncode round grain	1	1.8	2.0	0.5			2.0[f]						No strength
219	Noncode round grain	8	2.5	2.0				1.0[g]						Too dry
202	Noncode round grain	8	2.5	2.0				0.25[g]						Sticky. Not workable
272	Noncode round grain	1	1.5	2.0				0.125[g]	2	0	0	0	45	All ingredients except zinc stearate mixed 4 min. Zinc stearate added and mixed with others for 1 min. Total 5 min.
272	Noncode round grain	1	1.5	2.0				0.125[g]	1	0	0	0	88	All ingredients except zinc stearate mixed 4 min. Zinc stearate added and mixed with others for 1 min. Total 5 min.
212	Noncode round grain	8	2.5	2.0			4.0	6.0[h]	250	0	90	0	Normal	Dries fast, flowability 78
212	Noncode round grain	8	2.5	2.0			4.0	6.0[h]	4	0	30	0	88	
213	Noncode round grain	8	2.5	2.0			4.0	3.0[h]	170	0	92	0	45	Flowability 81
213	Noncode round grain	8	2.5	2.0			4.0	3.0[h]	27	0	38	0	88	
214	Noncode round grain	8	3.0	2.0			4.0	4.0[h]	2.5	0	30		88	No humidity resistance

[a] A water containing silicone spray
[b] Aluminum hydrogen phosphate
[c] Liquid urea resin
[d] $Ca(OH)_2$
[e] Hydrous magnesium silicate
[f] Portland Cement ($CaCl_2$—10%)
[g] Zinc Stearate
[h] ZnO

the other hand, the remaining silicate binder which has not been reacted by the gas will contribute proportionally considerably more strength as the water is removed leaving the silicate as an adhesive binding material.

Excessive water in the initial system enhances the rate of the CO_2 reaction giving initial high silica gel strengths but resulting in later decreased strengths as the system is dried out.

Water Quantities to Control

There are then, three quantities of water to consider and control carefully in the CO_2 molding process. This is assuming that the sand itself is dry:

1) The water is a part of the sodium silicate binder.
2) The water which is added with the sugar, syrup or as a controlled additive quantity in the initial mix.
3) The pickup of moisture in storage due to high relative humidities.

Proper choice of formula will obviate all these problems of water content provided the relative humidity during storage before usage is below 70 per cent. The gassing reaction to form the silicate gel becomes less sensitive to the water content with the favorable additives such as sugar and iron oxide. A desirable uniform coating of the sand grains during the mixing of formulation ingredients, or with a prepared composition, depends considerably upon viscosity of the binder material as well as the flowability of the sand during such a mixing action.

Controlled Gassing

A thin film of the silicate adhesives is most desirable. Excessive quantities of the binder only reduce the cohesive strength of the overall film. The strength of any mass is proportionately greater the smaller the quantity considered. Probability of structural breakdown gives credence to this hypothesis.

The previous discussion concerning water content

TABLE 8 — VARIATIONS IN GASSING AND ADDITIVES TO OBVIATE HUMIDITY PROBLEM

Exp. No.	Formula Change	Result	Effect on Collapsibility
168	Calcium hydroxide	Good	
118E, 121	Pre-drying in oven	Good	No effect
118F, 121A	Gassing with compressed dry air	Fair	No effect
118G, 121B	Gassing with combination of CO_2 and compressed air	Fair	No effect
135	A solvent carried silicone spray	Fair	No effect
145	Dipping in a hydrous magnesium silicate methanol mix	Fair	Some decrease
212, 213	Zinc oxide	Fair	
165, 166, 229	High silica to soda ratio silicates	Fair	Not favorable
122, 163	Iron oxide	Poor	Very good
129	Nitrogen replacing CO_2	Poor	No strength
129	Gassing with Nitrogen and CO_2	Poor	No effect
123	A water containing silicone spray	Poor	Not determined
124, 126	A water containing silicone compound mixed into formula	Poor	Not determined
127	Aluminum hydrogen phosphate addition	Poor	Incompatible
146	Dipping in an aluminum hydrogen phosphate	Poor	Decrease greatly
140	Addition of a hydrous magnesium silicate	Poor	No strength
201, 202	Liquid and solid silicate mixture	Poor	No effect
155	Mixed silicates	Poor	Not determined
161	Cresol-phenolic plastic powder	Poor	Decrease slightly
133	Fireclay	Poor	No effect
153	Calcium chloride	Poor	No strength
219, 220, 227	Zinc stearate	Poor	No effect
141	Liquid urea resin	Poor	No strength

of the silicate formulation indicated a need for controlled gassing. Laboratory studies have considered the variables of the gas pressure, gas composition and quantity and rate of gas usage, as controlled by venting and size of feed apertures. The laboratory gassing of test cylinder specimen was determined as 20 sec through a $\frac{5}{16}$-in. diameter exit at 35 lb gage pressure. A fine mesh corrugation support at the bottom of the cylinder during the gassing provided adequate venting.

Plant trials with muffle boards clamped as manifolds over wooden or metal core boxes gave best gassing performance when the boxes were well vented to allow a gas passage which will not exceed 10-in. laterally and 4 in. in diameter of flow (Fig. 9).

The structural strength initially developed with the CO_2 gas allows immediate handling of the mold or core. This is not the best bond which sodium silicate can contribute, however, since drying in storage will add considerably to the strength. Minimum CO_2 gassing to give adequate structural strength for normal core and mold handling is preferred. If the sand-silicate structure can be retained in place either by green strengths or adequate moderately high temperature resistant supports, convection drying or dielectric heating methods can be used to remove the moisture resulting in a strongly bonded silicate adhering structure requiring no CO_2 gas application.

In this case no gelling by CO_2 gas occurs, and there is no irreversible loss by drying of silicate bonding materials. This is one approach to semi-precision molding using heat methods similar to those used in phenolic shell molding.[26]

Figure 10 is a laboratory demonstration of one of

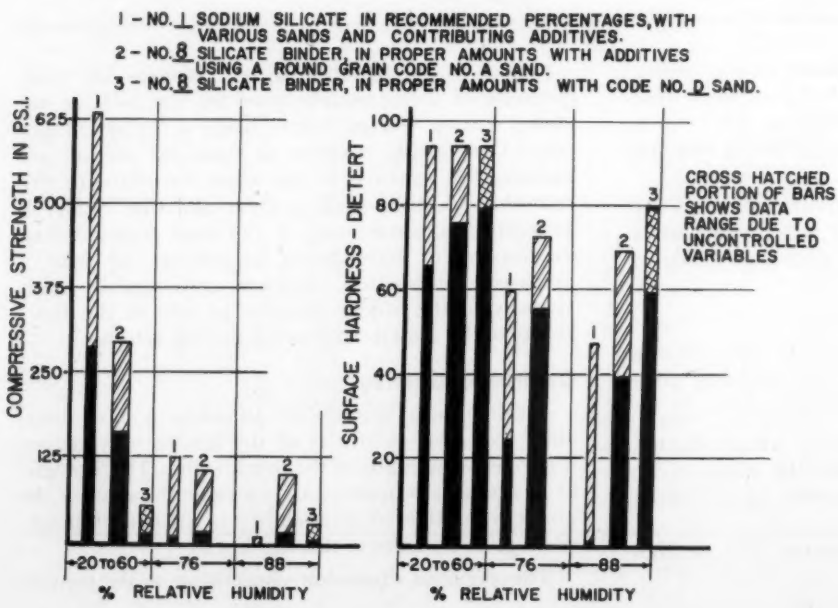


Fig. 8 — Relative humidity effect on strength and surface hardness.

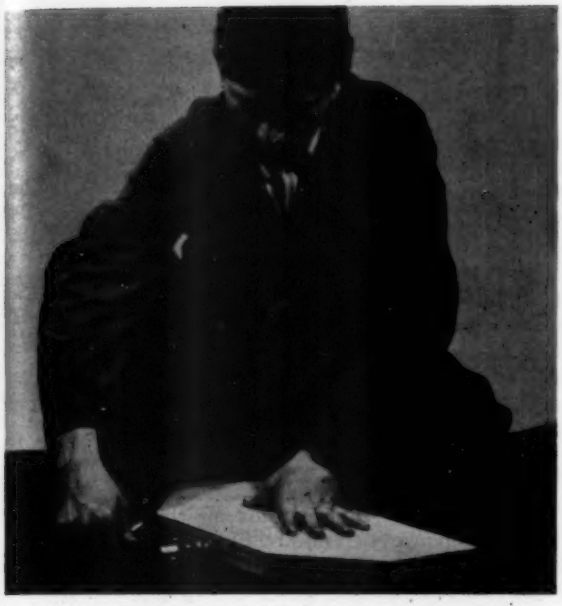


Fig. 9 — Gassing production half cores using a manifold and vented box.

the many methods of gassing[20,34] in which a glass reinforced epoxy pattern of shell construction, vented adequately through the surface, can be gassed from the inside cavity out through a mold of shell thickness providing enough initial strength for pattern removal and subsequent curing in a dielectric oven. There are advantages for gassing with a vacuum chamber where short runs on varied sizes of cores and molds are produced. There are reductions of CO_2 gas requirements also purported. Experiments 129, 118F, 118G, 121A, 121B (Tables 7 and 8) show the favorable possibilities of using dry air, hot air, nitrogen gas or the dilution of CO_2 gas with dry air.

The controlled gassing of experimental cylinder specimens supported literature knowledge, and supplied information that each sodium silicate binder

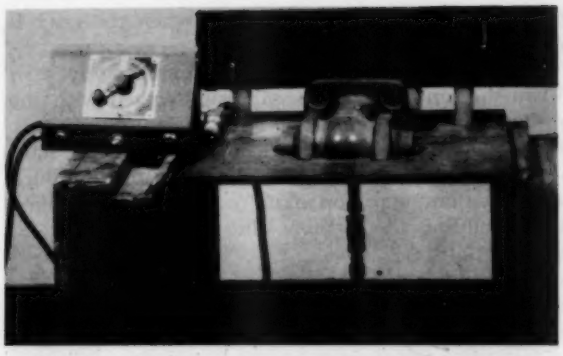


Fig. 10 — Laboratory demonstration of one of the methods of gassing in which a glass reinforced epoxy pattern of shell construction can be gassed from the inside cavity out through a mold of shell thickness.

composition related to difference of silica to soda ratio, water content and viscosity has its own optimum quantity of CO_2 for adequate initial strength as a gel form adhesive. The 51 degree Baumé binder no. 1, for instance, was gassed 20 sec at a 35 lb pressure, compared with 5 to 10 sec for a 52 degree Baumé silicate containing more silica due to a 2.4 to 1 ratio rather than a 2 to 1 ratio silica to soda content.

OTHER VARIABLES FOR CONSIDERATION

This designation should not be construed as minor. They are no less important than those previously considered. With the proper attention, however, they can easily be controlled within certain operating limitations. Let us consider:

Mixing Methods and Time

The order in which the ingredients of the formulations are added to the mixing unit, the action of the mixer and the time of the mixing can either be carefully controlled or given little attention. We know that CO_2 from the air and the action of drying will pre-set the silicate adhesive bond. If this occurs before the core or mold is formed, this amount of strength-giving sodium silicate is lost, and will even

TABLE 9 — NUMERICAL PARAMETERS COMPARING FIVE REPRESENTATIVE SANDS WITH THE MINIMUM CRITICAL PER CENT BINDER 1 REQUIREMENT

		Rounded Grain Ottawa Wedron	Sub Angular AFS 70 New Jersey	Sub Angular High Density AFS 81 New Jersey	Cape Sand AFS 105	Fine Grain Sub Angular Dividing Creek
	Code Number	A	B	C	D	E
Retained on screen	40	0.2	0.7	1.8	0.6	0.2
	50	4.4	7.4	17.3	1.4	1.8
	70	31.2	30.6	30.4	6.0	4.6
	100	41.2	42.4	22.6	30.6	11.2
	140	15.4	14.6	11.3	34.4	22.4
	200	5.6	3.0	8.0	19.2	25.6
	270	1.4	0.8	4.4	4.6	14.0
	325	0.6				
	PAN		0.5	3.7	3.2	20.3
AFS Fineness No.		73.92	70 ± 5	81 ± 5	105	158
Est. Surface shape factor		1.35	1.50	1.50	1.65	1.50
Specific surface by using ratio multiplier (1.82), called *A*, sq cm/gm		182	191	221.1	315	432
Specific surface B.E.T. Method, sq cm/gm		50	358	584	810	152
4th power of specific vol. ratios, called *B*		1.0	1.43	1.25	1.47	1.69
Multiple of *A* x *B*		182	273	276	464	730
Minimum Critical % Binder 1 Requirement		1.1	2.0	1.9	2.5	3.0

interfere with the proper bonding when the sand is rammed into position.

The mixing should be stopped immediately then when the sand grains are uniformly coated with the proper distribution of the formulation ingredients. Proprietary mixes have been pre-blended and have uniform ingredient distribution. Other additives such as silica flour and iron oxide can be pre-dry mixed before adding the sodium silicate, sugar and water compositions. A cresol-phenolic organic powder added in the small amount of 1 gram/lb of sand improves the flowability of those mixes which ram with difficulty.

Mixers operated from 1½ to 5 min depending on the difficulties of the mix, give adequate results in the laboratory. Mixers which throw the sand causing aerated drying should be avoided. The literature,[27] as well as experimental results, favor a simultaneous addition of sodium silicate, sugar and water rather than mixing them each separately into the sand.

Storage of Mix and Cores

The storage life of a sodium silicate sand mix depends considerably upon how it is kept uniform in water content. Tight metal containers covered with wet burlap bags sealed off from the CO_2 of the atmosphere can be kept for at least a day.

The storage life of a formed and gassed core or mold, however is almost indefinite, provided the relative humidity is not excessive for a time which would cause penetration and collapse of the bond structure. Even if they become weakened by high relative humidity, if they hold their shape, cores and molds will still be usable when adequately dried out before handling.

The Flowability of the Mix

The flowability of the sand binder mixture depends considerably upon the composition such as sand, type of silicate and additives used. The choice of recommended formulations given later has taken into account the valuable aspect of flowability. Structural strength and surface uniformity of the formed part depends upon optimum packing and uniform density of all parts of the structure. Flowability, as it concerns ramability and adaptability to best blowing or vibrating, makes these characteristics possible.

There is no reason why any of the recommended formulations should not be used for blowing, ramming or vibrating equal with green sand and urea bonded formulations. Plant runs have proven these formulations quite adaptable. Variations from these general types of formulas such as using additional binder, varying the type of silicate used or changing the sand without changing the sodium silicate content, may result in problems in ramming or blowing to form optimum core structure.

Adding water, or changing the silicate to improve the flowability of the mix without proper consideration of other properties, may result in poorly bonded cores or cores with poor collapsibility. Choice of sand to give adequate ramability is also important, and should not be determined without careful study. A laboratory attempt to compose a sand of perfect distribution for packing resulted in one with poor flowability. Sand distribution must consider flow and permeability as well as structural characteristics.

Temperature

Practically all chemical reactions are sensitive in some degree to temperature. The bonding with sodium silicate is no exception. However, ordinary ambient conditions are usually considered in the operationable range. Since the viscosity of the silicate decreases with temperature and the gelling action is accelerated with temperature, these would be markedly reduced if the sand or silicate were used immediately from a low temperature storage. Re-used sand, if allowed to remain hot in the mix, would cause accelerated drying resulting in a poorly bonded structure due to pre-dried silicate on the sand previous to mold forming.

RECOMMENDED FORMULATIONS

These selected formulations are based upon optimum properties determined by experimental studies taken from the various tables of results. Each sand is presented separately. Figures 11, 12 and 13 are examples of favorable results in the actual use of these formulations compared with standard core results.

Code D Sand

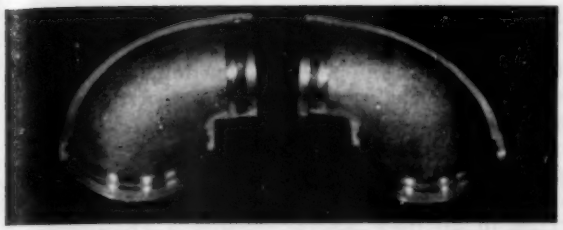
This sand is easily available in New England but not recommended as a selected sand for CO_2 molded cores. The high surface area with variable quantities of contaminants, such as mica, require higher percentages of bonding mixtures. The addition of small amounts of selected organic additives to this sand, are found to contribute:

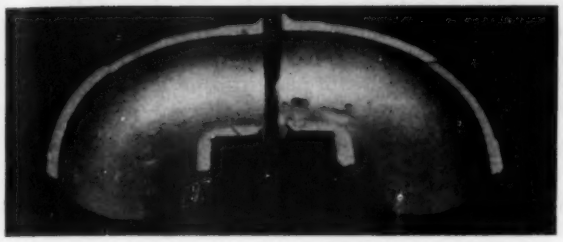
a) Flowability (ramming qualities) providing uniform density of the core or mold structure.
b) Production of just enough gas at the metal mold interface to prevent penetration and veining which seem prevalent when pouring the copper metals against this sand.

The formulas given in Table 11 have shown quite acceptable properties in the laboratory studies, and those designated have shown good performance in the plant operations.

Numbers 1 and 2 binders are 2 to 1 silica to soda ratio sodium silicates, 51 degree Baumé of about 380 centipoise viscosity. Number 8 binder is a 3.22 to 1 silica to soda ratio sodium silicate, 38 degree Baumé of about 60 centipoise viscosity. Binders nos. 1 and 2 have been found superior in their effectiveness in the CO_2 bonding process, but no. 8 binder gives some resistance and better performance in conditions of high humidity. Numbers 3, 4 and 5 binders are the type sodium silicates predominately used in the foundry industry. The silica to soda ratio is 2.4 to 1, 52 degree Baumé of about 1700 centipoise viscosity and 47 per cent total solids.

The laboratory evaluation comparisons show acceptable results, but in many ways the resulting cores and molds are inferior to the preferred 1 and 2 binders. The important advantage of these 3, 4 and 5 silicate binders is the shorter time of gassing required, which allows increased production with shorter gas



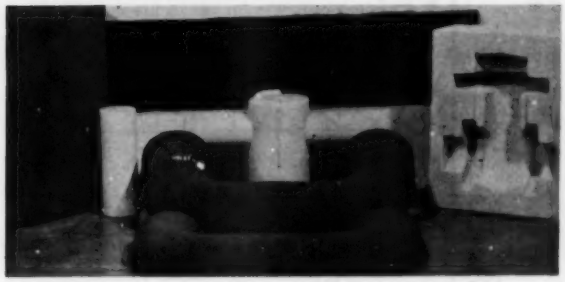


Figs. 11-13 — Examples of favorable results in the actual use of the recommended formulations compared with standard core results.

Fig. 11 — (left) — Sectioned casting, urea cored.

Fig. 12 — (lower left) — Sectioned casting, CO_2 cored.

Fig. 13 — (below) — Finished half cores and molds, including bonded half cores.



application cycles. Formulation 320 is included where this increased production rate seems important. Numbers 1 and 2 binders could be used in this same formulation with best results.

These formulations give adequate collapsibility over the higher temperature range required for brass (Figure 14 shows the results for formula 250 immersed in various furnace temperatures for 1 min and 5 min).

Code B Sand

Table 12 is a compilation of selected formulations using code *B* sand. The test data indicate formulas 230 and 252 as superior in all around requirements. Figure 14 is a collapsibility study of formulas 251, 207 and 252 using this sand, in which the cylinder specimen were immersed at various furnace temperatures for 1 min and 5 min. Code *C* sand is one which was recommended as possibly superior for the sodium silicate-CO_2 process. Data do not substantiate this possibility, and these recommended formulations do not include this sand competitively.

Code A Sand

This round grain sand is of highest silica composition and seems particularly suited for the sodium silicate-CO_2 molding process. Low percentages of binder are adequate, and a wide variety of formulations give excellent results (Table 13). Where the simple formulas such as 118 may present problems of hot expansion deformations (such as scabbing, and hot casting cracks), formulas such as 167 will correct this deficiency and reduce problems of penetration.

Code E Sand

Code *E* sand is included (Table 14) because it provides smooth surface cores and molds with lowest density packing. This combination reduces the possibility of providing adequate strength without sacrificing collapsibility. However, this sand with the silicate adhesive-type binder probably is superior to others where hot expansion problems are serious, causing mold wall fractures. The subangular, multiple screen properties of this sand allow movement with metal shrinkage where a coarser closely packed

TABLE 10 — PHYSICAL PROPERTIES OF SODIUM SILICATE BINDERS CONSIDERED IN THESE STUDIES

Silicate Binder Code Number	1	2	3	4	5	6	7	8	9	10	11	12
Ratio Na_2O:SiO_2	1:2	1:2	1:2.4	1:2.38	1:2.4	1:2.87	1:2.54	1:3.22	1:3.75	1:1.8	1:1.6	1:2.0
Na_2O, %	15	14.7	13.7	13.9	13.8	11.14	12.85	8.2	6.75	13.4	19.5	18.0
SiO_2, %	30	29.4	32.9	33.1	33.1	32.04	32.6	26.4	25.3	24.1	31.2	36.0
Total solids, %	45	44.1	46.6	47.0	46.9	43.18	45.45	34.6	32.05	37.5	50.7	54.0
Baumé (degrees)	51	50.5	52	52.25	52	47.59	50.5	38	35	44.6	58.5	59.3
Viscosity (centipoise)	380, est.	350	1600, est.	1700, est.	1700, est.		1120	60	220	60	7000	70000

Minimum Critical Percentage Sodium Silicate Binder							Best With No Water Addition					Additional Maximum % Can Add Using Proper Additives	
Sand A	1.1	1.6	1.6	1.7	1.5	1.4	1.7	2.1	*1.5	1.6	N.W.	*N.W.	1.4
B	2.0	2.2	2.4	2.4	2.6		2.6	3.2	*N.G.			*N.W.	1.2
C	1.9			2.8	2.5				*N.G.	2.5		*N.W.	0.6
D	2.5		3.0	2.8	3.4		2.8	3.0					0.8
E	3.0			3.3	3.3		3.2	3.8	*N.G.			*2.8	0.6

Critical percentage figures for binder 1 are based on considerable data. The critical percentages for the other binders dependent upon the sand used are based only on minimum experimental results—taken from Table 6.
*Requires water addition, N.W.—not workable

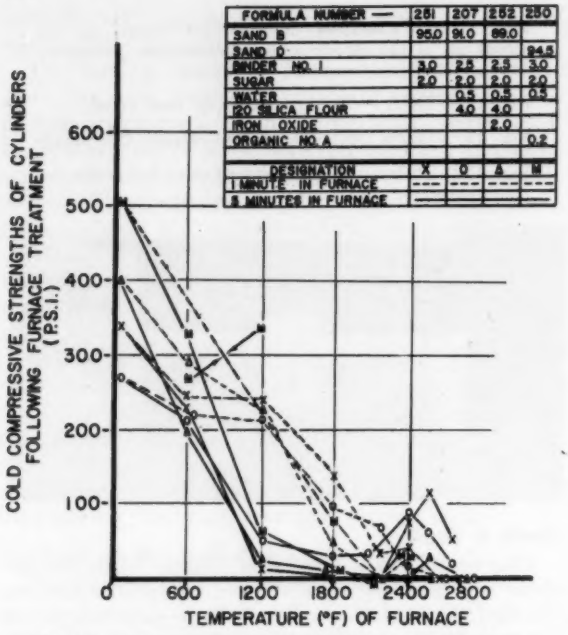


Fig. 14 — Furnace temperature effect on representative solium silicate CO_2 formulations.

round grain sand will require additions of organic cushioning materials such as wood flour[11] to extend the hot deformation properties of the mold structure.

The silicate bonded cores of this sand can be assisted in structural strength considerably with the addition of the coarser silica flours. Since the work with this sand was limited to a few trials, the optimum quantity addition is only estimated (Table 14).

LABORATORY EVALUATION METHOD

The initial series of cores made in the laboratory for metal casting in works production indicated the variable characteristics which seem important for controlling plant operations. The operating costs, quality of casting surfaces and dimensional precision are dependent upon the formulation selection and the conditions of mixing and gassing. To attain the optimum in all these aspects requires a knowledge of formulation and process variables. This knowledge was accumulated by laboratory and plant studies. The laboratory methods evolved by initial trial and error methods correlated with plant runs is described in the following paragraphs.

Procedure

Sand, sodium silicate, sugar and other additives are weighed out to the nearest 0.1 gram on laboratory balances. Mixes of 12½ lb in one laboratory mixer or 25 lb in a second mixer are prepared to a homogeneous consistency (Fig. 15). The dry ingredients were mixed first, followed quickly by a simultaneous addition of all liquid materials (sugar being initially dispersed in water if a water addition is included). Otherwise, the sugar is added with the sodium silicate.

The test cylinders are formed by placing 185 grams of the fresh mixtures into a polished metal cylinder of 2¼ in. inside diameter. The sand silicate mix is smoothed to a uniform thickness in the cylinder and placed in a press, as shown in Fig. 16. A solid metal cylinder piston with a self-aligning fixture attachment in the laboratory press is pressed into the cylinder against the sand to a 500 lb total pressure.

TABLE 11 — RECOMMENDED FORMULATIONS WITH CODE D SAND AND ADDITIVES

Lab. Exp. No.	See Table No.	Date of Plant Run	Silicate Binder Used, No.	Binder, %	Additives, %				
					Sugar	Water	Iron Oxide	Silica Flour (90-140M)	Other
		12/27/57	1	3.0	2.0	0.33			
174	1	10/10/58 and	1 or 2	3.0	2.0	0.5			*0.2
250	1	10/14/58 also 6/24/58 and 6/4/58							
236	1		1 and 2	3.5	2.0				
320	6		3, 4 and 5	3.2	2.0	0.5	2.0	4.0	
176	1		8	3.5	2.0				
177			12	2.0	2.0	0.5			

*Cresol phenolic organic derivative.

TABLE 12 — RECOMMENDED FORMULATIONS WITH CODE B SAND AND ADDITIVES

Lab. Exp. No.	See Table No.	Date of Plant Run	Silicate Binder Used, No.	Binder, %	Additives, %			
					Sugar	Water	Iron Oxide	Silica Flour (90-140m)
*251 and 236	2		1	3.0 to 3.5	2.0			
324	2	Plant Production	2	3.0	2.0			
230 and 207	2		1	2.3 to 2.5	2.0	0.5		4.0
252	2		1	2.5	2.0	0.5	2.0	4.0

*Compare with Exp. 324 with 3% no. 2 binder (Table 2); No. 236 gave better collapsibility.

TABLE 13 — RECOMMENDED FORMULATIONS WITH CODE A SAND AND ADDITIVES

Lab. Exp. No.	See Table No.	Date of Plant Run	Silicate Binder Used, No.	Binder, %	Additives, %				
					Sugar	Water	Iron Oxide	Silica Flour	Cresol Phenolic Organic Derivative
118	3	12/31/57	1	1.8	1.5 to 2.0	0.35			
122	3		1	1.8	2.0	0.5	2.0		
167	3		1	*2.2 to 2.5	2.0	0.5	2.0	4.0	
136	3	12/31/57	1	1.8	2.0	0.35			0.2
165	3		8	2.5	2.0		2.0	4.0	
157	3		6	2.5	2.0	0.5			
308	6		10	2.0	2.0	0.5	2.0	4.0	
336 and 312	6		5, 4, 3	*2.0 to 2.4	2.0	0.5	2.0	4.0	

*Percentage depends upon requirements of collapsibility.

TABLE 14 — RECOMMENDED FORMULATIONS WITH CODE E SAND AND ADDITIVES

Lab. Exp. No.	See Table No.	Date of Plant Run	Silicate Binder Used, No.	Binder, %	Additives, %			
					Sugar	Water	Iron Oxide	Silica Flour (90-140m)
349 and 306	4		1	1.8 and 2.5	2.0	0.5	2.0	4.0
*350	4		1	3.5	2.0	0.5	2.0	4.0
		Proposed Formula.....	1	2.5 to 3.5	2.0	0.5	2.0	6 to 12

*Shows poor collapsibility; 2.4% is the limit for retaining acceptable collapsibility even with the additives.

This appears to equal uniform plant ramming, giving a compaction of sand similar to blown cores with density and permeability varying only according to the sand and silicate sugar formulation used. All cylinders of a given formulation, however, came out surprisingly of the same height indicating a uniformity of density by this controlled pressure method.

The cylinder assembly is removed from the press, tapped lightly with a rubber hammer and gassed with a standard plunger gassing unit for 20 sec at 35 lb pressure (Fig. 17). In some cases, the gassing time was changed depending upon the type of sodium silicate used as the bonding ingredient. Figure 18 is an example of such cores ready for test. Scratch hardness was made on each test series. Two to four readings were made before the muffle furnace heat immersion and after heat immersion (Figs. 19, 20). Figure 21 shows a cylinder being evaluated from compressive strength before heat application.

At least three cores from each series were place in the muffle furnace at a controlled high temperature. In most cases, this was at a temperature of 2200 F. The time of heat immersion was eventually standardized at 5 min (Table 16 gives a correlation comparing two, five, and 15 min in the furnace at 2200 F). Figure 22 is an example of such a furnace run.

The cylinders after heat (Fig. 23) will vary in strength and surface hardness according to the potential collapsibility of the formulation tested (Figs. 20 and 24).

Variations in storage time and conditions of relative humidity were recorded at the time of mixing and in storage. Controlled comparisons were made on cylinder cores kept in desiccators at different relative humidities (Fig. 25). Tables 15 and 16 are examples of data recorded during two experimental studies. The data recorded in Tables 1 through 8 are compiled from such sheets.

The flowability of some of the experimental mixes as recorded in the tables of results were determined with a flowability indicator which was used as an



Fig. 15 — Mixes of 12½ lb in one type laboratory mixer or 25 lb in a second type mixer are prepared to a homogeneous consistency.

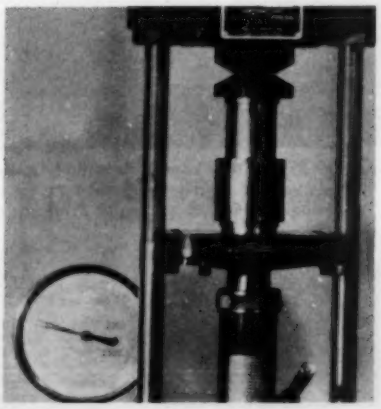


Fig. 16 — A solid metal cylinder piston with a self-aligning fixture attachment in a laboratory press is pressed into the cylinder against the sand to a 500 lb total pressure.

TABLE 15 — REPRESENTATIVE DATA SHEET FOR AN EXPERIMENTAL LABORATORY STUDY OF ONE FORMULATION (CO_2 MOLD STUDIES)

Sand	sand A in code designation Test: 122 standard formula
Sod. Sil.	binder 1, 51 degree Baumé, 1.8%
Additives	sugar, 2.0%; water, 0.5%; iron oxide, 2.0%
Mixing	5 min, 12.5 lb total
Mold Form	2¼ x 3 in. cylinder, hand rammed and pressed at 500 lb
Gassed	20 sec at 35 psi
Furnace Heat	2200 F.
Remarks	Immediately after gas S 95; shows poor resistance to high humidity Compressive strength 0 after heat, good collapsibility

Storage hr.	Designation: Min. In Furnace	Designation: Sample No.	Compressive Strength: Before Heat	Compressive Strength: After Heat	Compressive Strength: After gas.	S Hardness: After stand. (5 hr)	S Hardness: After heat	General Appearance
			Normal relative humidity 38%					
5	0	1	2000			90		Excellent
5	0	2	2175			90		
5	0	3	2250			87		
5	5	4		0			0	
5	5	5		0			0	
5	5	6		0			0	
			Relative humidity 76%					
5	0	7	350			35		
5	0	8	350			60		
5	0	9	500			35		
5	5	10		0			0	
5	5	11		0			0	
5	5	12		0			0	
			Relative humidity 88%					
5	0	13	50			10		
5	0	14	25			30		
5	0	15	50			50		
5	5	16		0			0	
5	5	17		0			0	
5	5	18		0			0	

attachment to a 315 sand rammer. Flowability is only one parameter, and although good flowability is required for optimum cores and molds it must be combined with other properties such as strength, surface hardness and collapsibility after heat. These are all considered in determining best formulations and operating variables. The data compare equal with the flowability of sand formulations used in other molding processes.



Fig. 17 — The cylinder assembly is removed from the press, tapped lightly and gassed with a standard plunger gassing unit for 20 sec at 35 lb/sq in. pressure.

REFERENCES

1. B. H. Booth and C. A. Sanders, "Another Look at Sand Grain Distribution," AFS TRANSACTIONS, *vol. 62*, p. 499, 1954.
2. Robert E. Morey, "Use of Sieve Analysis in Determining Surface Area of Sand," *Foundry*, pp. 100 and 101, Aug. 1957.
3. *Foundry Sand Handbook*, Sec. 5, "Methods of Determining Fineness of Foundry Sands," American Foundrymen's Society, Sixth Edition, 1952.
4. J. M. Leaman and D. C. Ekey, "Statistical Techniques for Classifying Foundry Sands," AFS TRANSACTIONS, *vol 64*, pp. 679-687, 1956.
5. AFS Mold Surface Committee 8-H, "Influence of Sand Distribution and Surface Coatings on Metal Penetration," AFS TRANSACTIONS, *vol. 64*, pp. 82-90, 1956.
6. "Symposium on New Methods for Particle Size Determination in the Sub-Sieve Range," Special Technical Publication No. 51, A.S.T.M., March 4, 1941.
7. H. E. Rose, *The Measurement of Particle Size in Very Fine Powders*, four lectures delivered at King's College, London, Chemical Publishing Company, Inc., New York, 1954.
8. R. D. Cadle, *Particle Size Determination*, Inter-Science Publishers, Inc., New York, 1955.
9. J. B. Caine and C. E. McQuiston, "The Theoretical Concepts of the Packing of Small Particles," AFS TRANSACTIONS, *vol. 66*, p. 36, 1958.
10. E. J. Grott, "Particle Packing — Principles and Limitations," AFS Sand Division, Basic Concepts Committee 8-B, AFS TRANSACTIONS, *vol. 66*, p. 553, 1958.

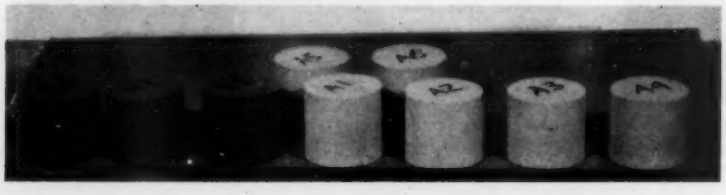


Fig. 18 — Completed cores ready for testing.

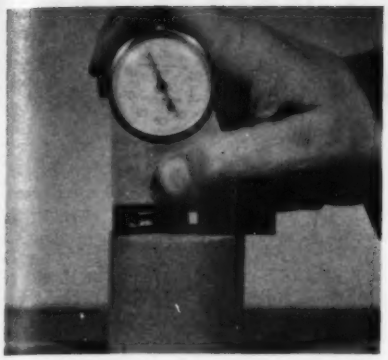


Fig. 19 — Surface hardness testing before heat.

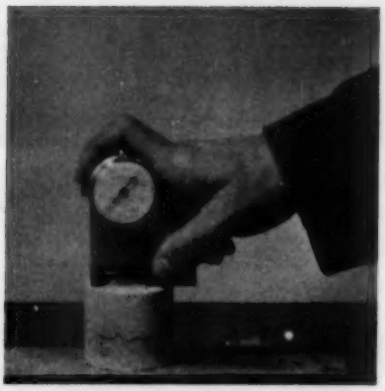


Fig. 20 — Surface hardness testing after heat.

11. H. W. Dietert, "Processing Molding Sand," AFS TRANSACTIONS, *vol. 62*, p. 1, 1954.
12. W. Davies, *Foundry Sand Control Testing, Research and Development*, The United Steel Companies, Ltd., Sheffield, England, 1950.
13. R. W. Heine, "Molding Sands, Molding Methods, and Casting Dimensions," AFS TRANSACTIONS, *vol. 64*, pp. 398-407, 1956.
14. J. K. Sprinkle and H. F. Taylor, "Adhesion of Phenol-formaldehyde to Various Refractory Oxides," AFS TRANSACTIONS, *vol. 65*, p. 300, 1957.
15. The following papers, taken from a symposium, "Adhesion and Adhesives, Fundamentals and Practice," Case Institute of Technology, Cleveland, Ohio, and in London, England. Printed by the Society of Chemical Industry, John Wiley & Sons, 1954. Reinhart, Frank W., "Survey of Adhesion and Types of Bonds Involved;" S. J. Czyzak, "on the Theory of Adhesion;" Gerard Kraus, "Energy of Adhesion, Molecular Forces and the Adhesive Rupture;" Charles Kemball, "Intermolecular Forces and the Strength of Adhesive Joints;"

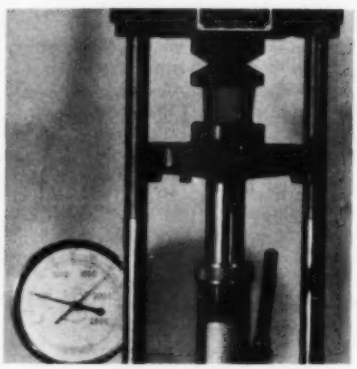


Fig. 21 — Cylinder being evaluated for compressive strength (pounds to crush) before heat application.

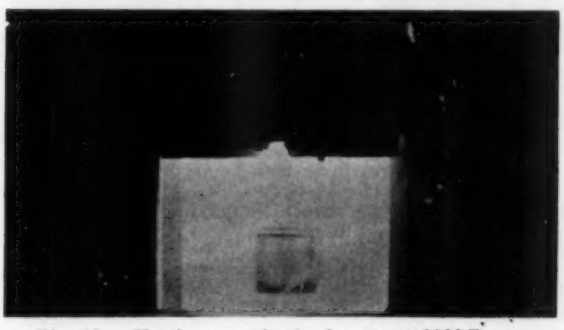


Fig. 22 — Heating cores in the furnace at 2200 F.

TABLE 16 — REPRESENTATIVE DATA SHEET FOR AN EXPERIMENTAL LABORATORY STUDY OF ONE FORMULATION (CO_2 MOLD STUDIES)

Sand ..... D in code designation
Sod. Sil ..... binder 1 51 degree Baumé, 3.5%
Additives ..... none
Mixing ..... 4 min 10 lb batch
Mold Form ..... 2.25 x 3 in. cylinder, hand rammed and pressed at 500 lb
Gassed ..... 20 sec at 35 psi
Furnace Heat ..... 2200 F., 2, 5, and 15 min
Remarks ..... Shows poor collapsibility

	Density		Compressive Strength			S Hardness		
Designation	Before Heat	After Heat	Before Heat	After Heat	After gas.	After standing	After Heat	General Appearance
				2 min				
25S	1.52		2150		72			Very good
26S	1.52	1.50		975	75	80	70	
27S	1.54	1.50		1800	70	81	68	
28S	1.49	1.45		1090	75	85	68	
				5 min				
29S	1.58		2425		81			
30S	1.53	1.50		975	68	90	65	
31S	1.51	1.50		1450	55	87	68	
32S	1.51	1.47		1950	65	83	68	
				15 min				
33S	1.54		2075		60			
34S	1.53	N/D		75	55	78	N/D	
35S	1.49	1.32		674	65	81	N/D	
36S	1.49	1.45		1915	60	72	68	

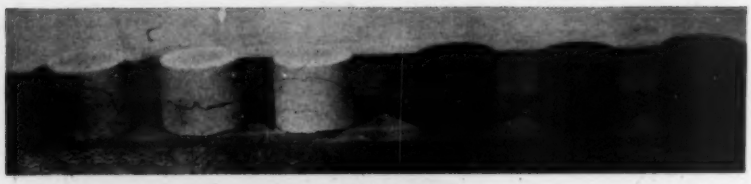


Fig. 23 — Cores after heat.

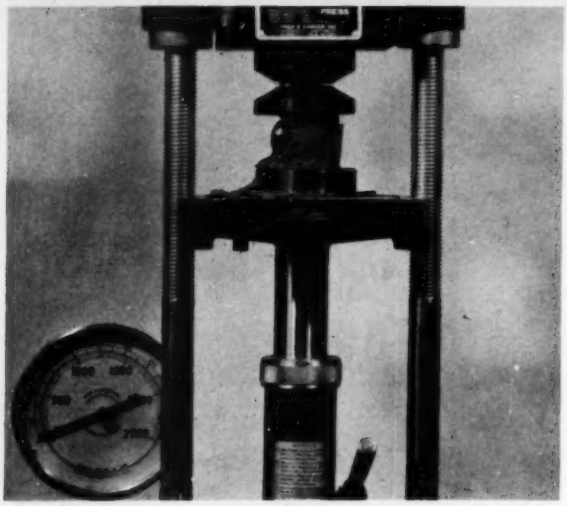


Fig. 24 — Testing compressive strength after heat.

J. J. Bikerman, "The Mechanism of Adhesion;" R. F. Blomquist, "Types and Uses of Adhesives;" G. W. Koehn, "Behaviour of Adhesives in Strength Testing;" W. S. Macfarlane and J. F. Sewell, "Sodium Silicate as an Adhesive."

16. D. V. Atterton, "The Carbon Dioxide Process," AFS TRANSACTIONS, *vol. 64*, pp. 14-40, 1956.
17. John E. Gotheridge and Frank Pursall, "Do's and Dont's in Hardening Sand with CO_2," *Foundry*, pp. 128-135, March, 1956.
18. Walter E. Gruver, Jr., "The Carbon Dioxide Process," *Foundry*, pp. 106-110, June, 1957.
19. T. E. Barlow, "Developments of High Pressure Molding with CO_2 Process Sands," AFS TRANSACTIONS, *vol. 64*, pp. 336-338, 1956.
20. The following papers concerning the CO_2 process, AFS TRANSACTIONS, *vol. 64*, 1956. J. E. Huss, "A Survey of the CO_2 Process," p. 98; D. V. Atterton, "The Carbon-Dioxide Process," p. 14; T. E. Barlow, "Developments of High Pressure Molding with CO_2 Process Sands," p. 336; F. M. Scaggs, "Core Making with CO_2 Process," p. 333.
21. Robert D. Fenity, "The Use of Sodium Silicate in Steel Casting Molds and Cores Literature Survey," *Journal of Steel Castings Research*, S.F.S.A., *no. 5*, pp. 7, 8, 9, and 14, May, 1956.
22. A. B. Steck, "Water Glass Carbon Dioxide Process," *Steel Foundry Facts No. 163*, S.F.S.A., pp. 8-11, Feb., 1956.
23. A. B. Steck, "The Economic Importance of the Carbon Dioxide Process," *Steel Foundry Facts No. 174*, S.F.S.A., pp. 11-12, Feb., 1957.
24. Herbert Gersch and Cameron G. Harman, "Sodium Silicate Carbon Dioxide Inorganic Bonds for Molds and Cores," Report on Research Project No. 43, "Inorganic Bonding of Molding Materials," Technical Research Committee, S.F.S.A., Sept., 1958.
25. Jose Navarro and H. F. Taylor, "Inorganic Binders Solve Shell Molding Problems," AFS TRANSACTIONS, *vol. 64*, pp. 625-635, 1956.
26. B. J. Ahearn and G. A. Gartner, "Sodium Silicate Bonded Shell Molds," *Foundry*, Feb., 1958.
27. E. A. Lange and R. E. Morey, "Sodium Silicates for the CO_2 Process," AFS TRANSACTIONS, *vol. 66*, p. 315, 1958.
28. James G. Vail, *Soluble Silicates, vols. 1 and 2*, American Chemical Society, Monograph No. 116.
29. E. A. Lange, Abstract from Metallurgy Division, Metal Processing Branch, N. R. L., problem no. M02-03, Principles of Casting Metals (Carbon Dioxide-Sodium Silicate Process for Bonding Foundry Sands).
30. Alan S. Michaelas, "Rheological Properties of Aqueous Clay Systems," from the book *"Ceramic Fabrication Processes,"* edited by W. D. Kingery, Technological Press of M.I.T. and John Wiley and Sons, pp. 23-31, 1958.
31. Ernest M. Levin, Howard F. McMurdie and F. P. Hall, "Phase Diagrams for Ceramists," Edited and Published by the American Ceramic Society, Inc., 1956.
32. David B. Atterton, "Surface Finish of Steel Castings," *Foundry*, pp. 92-95, Feb., 1958.
33. D. C. Ekey and E. G. Vogel, "Steel Foundry Uses New CO_2 Gassing Technique," *Foundry*, pp. 134-138, Aug., 1957.
34. Leonard Romano and Anton Dorfmueller, Jr., "Molding with CO_2," *Foundry*, pp. 104-107, Oct., 1957.



Fig. 25 — Cores in desiccators, relative humidities 35, 76 and 88 per cent.

APPLICATION OF PRESS FORGED CASTINGS

By P. Gouwens, T. Watmough and J. Berry

ABSTRACT

Recent developments have proven the feasibility of a duplex process for creating high strength metal shapes. This process is a combination of casting and press forging at selected temperatures to obtain a final part which has unique properties and advantages. The first part of the process consists of making a casting by conventional quality foundry practice. This casting will be so dimensioned that the deformation imparted to it by subsequent forging will result in a correctly shaped and sized final part.

Therefore, the casting would correspond in forging parlance to the last-stage forging blank. Proper heat treatment would then be employed to put the metal in a solution-treated or austenitized condition prior to press forging. Forging is done isothermally at temperatures below that which would result in recrystallization but which would influence other metallurgical reactions.

It is noteworthy that the forging stage of this process differs from conventional hot forging. The usual hot forging work is done at temperatures high enough to impart almost unlimited ductility to the metal and a low flow stress. This makes for ease of forming and freedom from metal rupture during forming. In addition, the metallurgically active condition of the metal results in rapid recrystallization and its accompanying relief of strain-induced hardening, so that this strengthening mechanism is largely absent in a normally finished forging.

INTRODUCTION

Technically speaking the forging work discussed in this duplex process, since no recrystallization occurs, is properly termed cold-working, even though temperatures up to 800 F are employed. These higher temperatures impart easier deformation characteristics to the part than with cold-working, but also result in a degree of relief from the strain-hardening effect. However, these factors are not the reasons for selection of specific temperatures for the press-forging operation.

Instead, the temperatures selected are based on potential metallurgical structure changes with time. The work can affect the orientation, size and distribution of hardening nuclei and, in addition, can shift the time and temperature at which structural changes occur. Strength benefits are likely to accrue as a result of these changes.

The objectives of this process are to modify the mechanical properties of castings without unduly reducing their economic advantages. Specific advantages possible are as follows:

1) Higher ultimate tensile strengths, even up to ultra-high strength levels when compared to heat treated castings of the same composition.
2) Corresponding increases in yield and proof strength without the same degree of ductility decrease usually associated with strengthening by heat treatment only.
3) Relative freedom from the marked directional properties encountered in most forgings.
4) Better resistance to impact and fatigue loads at normal temperatures.
5) Reduced lead time and costs when compared to forgings for short run or prototype parts.

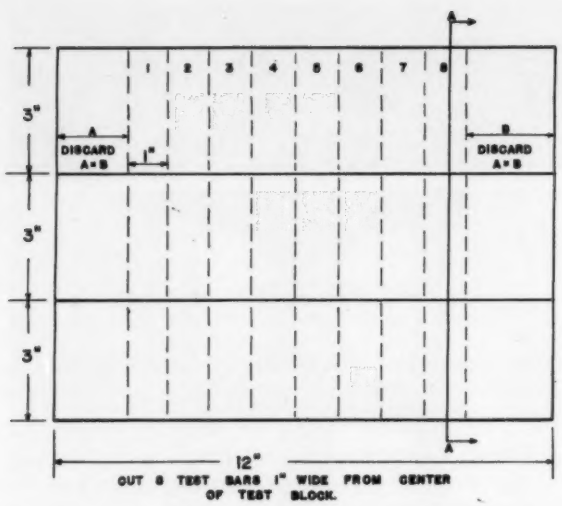
LITERATURE BACKGROUND

The technical literature contains information pertinent to the press forging of castings. Sims[1] made a comparison between the mechanical properties of SAE 4100 steel castings and forgings, one aspect of which included the use of cast forging blanks. Where unsound castings were used, worth-while improvements in mechanical properties resulted; but sound castings gave only a minor improvement after forging. Conventional hot forging techniques were used. He further reported that this practice of casting forging blanks was prevalent in captured World War II German material, both in ordnance and in aircraft. An economic appraisal of the combination process was favorable to it.

The effect of strain on the rate and temperature of metallurgical reactions in certain metallurgical structures has long been known qualitatively, but quantitative investigations of the influence of such strain-induced structures on the mechanical properties are of recent vintage. Thus the warm working of an age-hardenable austenitic steel by Harris and Bailey[2] in the precipitation temperature range of 1292 to 1472 F increased the proof stress from 31,400 psi to 72,800 psi. The maximum aged strength without the forging work was only 44,800 psi.

Lips and Van Zuilen[3] using a 4.5 per cent Ni, 1.5 per cent Cr, 0.35 per cent C steel obtained tensile strengths in excess of 400,000 psi with 42 per cent reduction in area. This compared with heat treated properties of 295,000 psi and 5 per cent reduction in area. This tremendous improvement was obtained

P. GOUWENS, T. WATMOUGH and J. BERRY are with Metals Rsch., Armour Rsch. Foundation of Illinois Institute of Technology, Chicago.



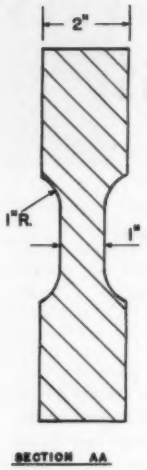


Fig. 1 — Test block casting.

by forging metastable austenite below the recrystallization temperature.

Schmatz and Zackay[4] in an extension and verification of the previous work investigated a range of lower alloy high-strength steels. The steels were deformed by compression, before any transformation had occurred, while at temperatures of 600 and 1000 F. Deformations up to 75 per cent were applied by rolling a wedge 5 in. long which tapered from ¼ to ⅛-in. thick. After rolling, martensite was produced by quenching. Tempering temperatures of 500 to 600 F gave steel with tensile strength in excess of 300,000 psi and 6 per cent elongation.

Austenite Deformation

Kula and Dhosi[5] studied the influence of deforming austenite at 1000 and 1550 F in a 4340 steel. Deformation of 72 per cent increased the yield strength of 280,000 psi and the tensile strength of 310,000 psi after tempering at 450 F, improvements of 19 and 13 per cent, respectively.

The only reported work done on property evaluation of warm-forged castings was that of Murphy, Clark and Rostoker,[6] who intensively investigated three non-ferrous alloys. For the 220 aluminum alloy, improvements of 100 per cent in the yield strength were accomplished by cold forging the metal section 40 per cent. Ductility was reduced to 7 per cent but was still judged adequate. For the AZ92 magnesium alloy, forging improved the yield strength 40 per cent with the test bar deformed 20 per cent, and no ductility loss was experienced.

It is apparent from this résumé that no work has been done on the forging of steel castings at temperatures which would lead to strain-induced or modified reactions. The literature cited for wrought steel investigations clearly indicates the possibilities for improvements in the mechanical properties of steel castings by such treatment. The data which follow are extracted from existing research in this area under the sponsorship of the Manufacturing Methods Div. of the Air Materials Command, Wright-Patterson Air Force Base, Contract No. AF 33(600)-36387.

TECHNIQUES AND EXPERIMENTAL METHODS

The mechanical property evaluation has been done on rectangular cross-section test bars sliced from the large block casting shown in Fig. 1. These test bars are uniform in composition, soundness and grain structure. To ensure this quality, the bars were taken only from the central portion of the block casting where end effects were no longer felt, and then were x-rayed to 2 per cent sensitivity at right angles to the plane of the saw cut.

The test bars have been solution treated or austenitized, quenched to isothermal conditions (temperature dependent on alloy composition) and press forged a controlled amount at this temperature by the dies which were heated or cooled accordingly. A view of the heated dies is shown in Figs. 2 and 3.

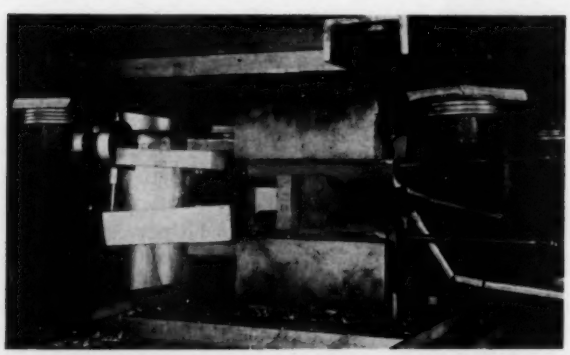


Fig. 2 — Front view of press forging arrangement.

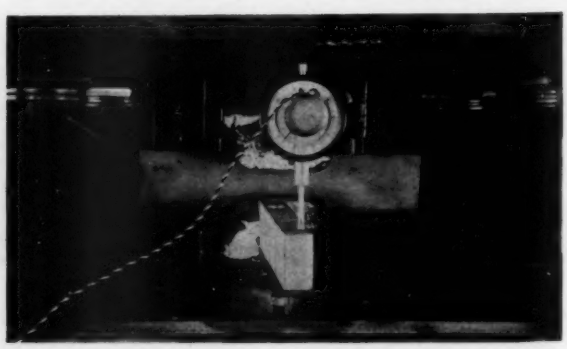


Fig. 3 — Side view of press forging arrangement.

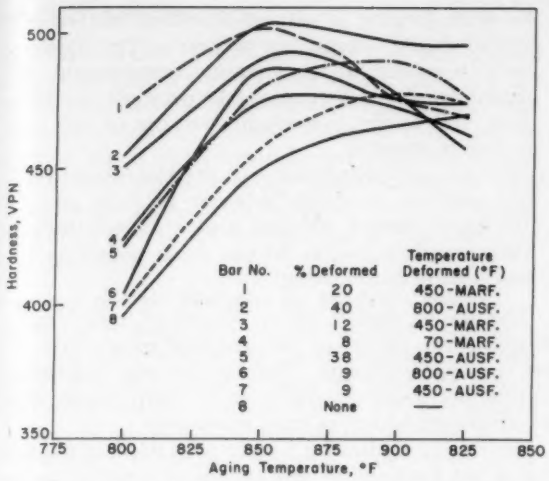


Fig. 4 — Aging reaction of 17-4PH after deforming.

They are then immediately quenched to previously selected temperatures. The time allowed for each of these operations can be critical for certain easily transformed steels. The forging effort was applied only to the two faces of the bar in the reduced gage length.

Work is in progress of three steel compositions representative of three basic families—the AISI 4340 alloy steel, 17-4PH precipitation hardening stainless and CF-8C (347) austenitic stainless. Aging of a 17-4PH steel which had been press forged at three temperatures caused a distinct increase in hardness. In addition to the hardening effect, deformation also caused a shift in the optimum hardening temperature. These results are shown in Fig. 4. That tensile properties were also affected, is shown by the data of such properties vs. deformation in Table 1.

The remarks following summarize the observed effects of press-forging 17-4PH:

a) Press forging at room temperature of a principally martensitic structure results in a marked increase in mechanical properties for relatively small amounts of deformation.
b) By press forging a similar structure, at 450 F, or alternatively press forging an austenitic structure (ausforming), the mechanical properties obtained by room temperature deformation may be approached. However, the approach is at a slower rate, and it must be noted that for the ausforming a combination of lower yield and higher elongation is obtained.
c) Reheating at 800 F and press forging at this temperature results in an embrittled structure. This is analogous to temper embrittlement experienced in other chromium-rich steels.
d) The aging temperatures required to develop the optimum mechanical properties are lowered by press forging prior to aging, as anticipated from the hardness results.

The curves shown in Figs. 5, 6, 7 and 8 graphically repeat this data to show more clearly the rate of approach to optimum values.

TABLE 1 — MECHANICAL PROPERTIES OF 17-4PH STAINLESS STEEL PRESS FORGED AND AGED

Deformation, %	0.2% Yield Stress, psi	UTS, psi	Elong., % 1 in.	Elong., % 2 in.	Aging Temp., F
	Base Properties				
—	158,000	175,000	9	5	925
	Press Forged at Room Temperature				
4.1	193,000	202,000	5.1	2.8	860
5.1	187,000	202,000	4.3	2.4	860
10.0	198,000	208,000	4.5	2.5	860
11.0	194,000	204,000	5.3	2.8	860
	Press Forged at 450 F				
4.2	193,000	205,000	6.5	3.3	860
11.5	197,000	206,000	5.0	3.0	860
15.5	190,000	203,000	4.7	2.5	860
20.0	—	198,000	5.0	2.8	860
11.4	186,000	203,000	6.0	3.8	880
20.0	194,000	202,000	5.7	3.5	880
10.7	182,000	200,000	5.7	3.0	900
23.8	196,000	208,000	5.0	3.0	900
	Ausformed at 450 F				
6.4	170,000	191,000	20.0	10.8	900
12.5	180,000	187,000	10.0	6.3	900
13.2	177,000	190,000	11.0	5.8	900
23.6	181,000	192,000	10.0	6.3	900
23.9	—	192,000	19.2	9.5	900
23.0	184,000	192,000	9.0	4.8	860
39.6	183,000	206,000	9.5	5.7	860
	Ausformed at 800 F				
8.6	182,000	202,000	7.2	3.9	860
18.8	162,000	200,000	5.0	2.5	860
19.5	182,000	200,000	6.2	4.0	860
38.9	184,000	203,000	13.5	7.5	860
39.3	184,000	204,000	18.5	10.0	860

Press Forging Deformation

An 18/8 stainless steel, made to the CF8C specification, was deformed by press forging at three selected temperatures. This treatment was followed by tempering at previously determined optimum temperatures, as selected by hardness studies. Table 2 shows the increase in mechanical properties consequent upon press forging at −80 F room temperature.

The data presented in Table 2 is summarized:

a) The mechanical properties of both the high- and low-carbon stainless steel are significantly improved

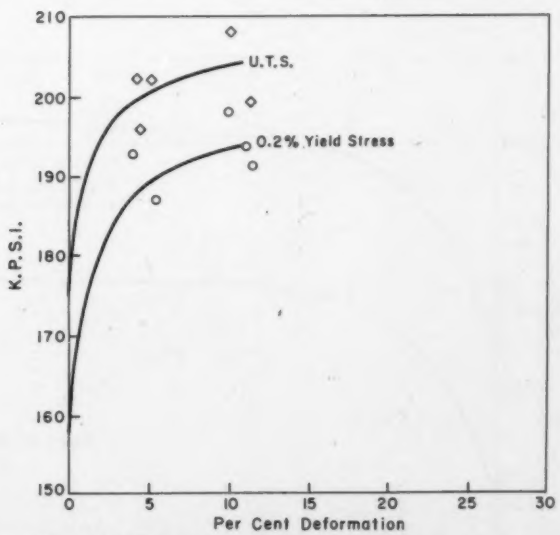


Fig. 5 — Mechanical properties of 17-4PH press forged at room temperature and aged at 860 F.

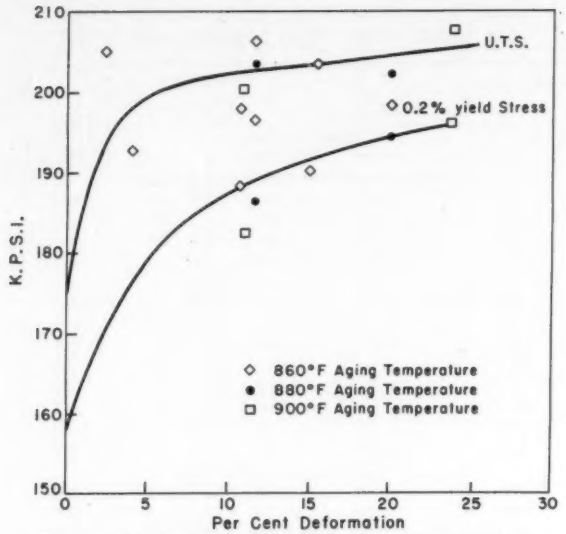


Fig. 6 — Mechanical properties of 17-4PH reheated to 450 F, press forged and aged at various temperatures.

by press forging at all temperatures, but in particular at the sub-zero temperatures. The improvement is related to the amount of martensite produced by the deformation/temperature combination, and to the subsequent working of this low-carbon martensite.

b) The elongations obtained in the low-carbon stainless steel are markedly superior to those seen in the higher carbon stainless steel, although the ultimate tensile strengths do not reach such high values as those of the latter.

The graphs, Figs. 9, 10, 11, 12 and 13, more clearly illustrate the effect on the ultimate tensile and 0.2 per cent yield strength. The various post treatments of bars previously forged at room temperature show no advantage for any of these treatments.

Finally, strengthening by the formation of martensite is not confined to any particular location in the bar cross-section. Photomicrographs, shown in Figs. 14 and 15, indicate that slip plane orientation is more critical than location of the individual grains.

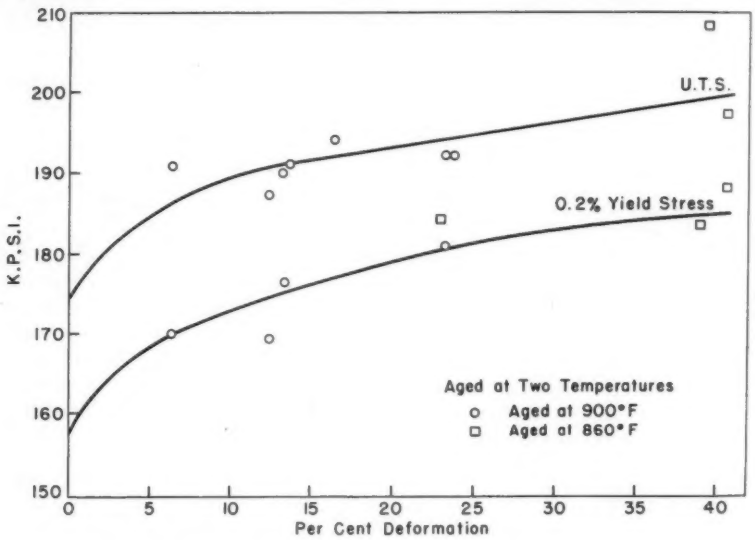


Fig. 7 — Mechanical properties of 17-4PH ausformed at 450 F by press forging.

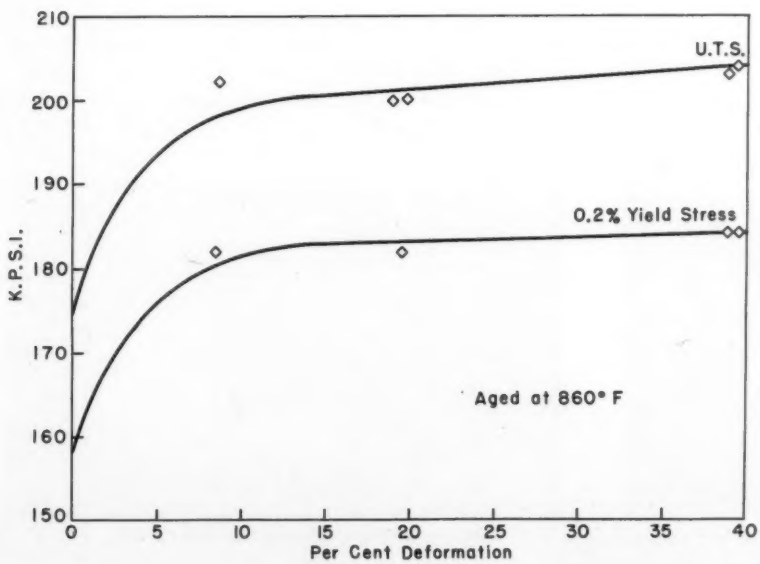


Fig. 8 — Mechanical properties of 17-4PH ausformed at 800 F by press forging.

PROCESS APPLICATION

The conversion of this laboratory duplex process to a useful production method will require the development of more knowledge. The problems fall into two broad categories—namely, those of most concern to the designer of the cast-forged component, and those of most concern to the manufacturer of such a component. As with all methods, there are important points of mutual interest that will demand coordinated effort.

Problems of the Designer of Forged Castings

It is the responsibility of the designer to make a part functional within the space limitations allowable, to provide ways of making the necessary attachments, to know the forces active on the part and to construct a shape which will withstand those forces when constructed from a material with known properties.

In order to use a new material intelligently, it is imperative that the mechanical and physical properties be known and be reproducible. Information currently available for this duplex process is inadequate in both respects, and the present work will fill only

TABLE 2 — MECHANICAL PROPERTIES OF 18/8 STAINLESS STEEL (CF8C) LOW CARBON AND HIGH CARBON, AFTER PRESS FORGING AND TEMPERING

Carbon Content, %	Deformation, %	0.2% Yield Stress, psi	UTS, psi	Elong., % 1 in.	Elong., % 2 in.
		Base Properties*			
0.09	—	32,000	74,500	44	35
0.03	—	30,600	70,200	85	65
		Press Forged at Room Temperature**			
0.09	5	54,000	78,000	26	18
	18	96,500	107,000	11	5.5
	39	129,000	155,000	4	2.5
0.03	5	50,200	79,000	71	50
	17.4	87,600	104,000	46	27
	36.8	114,000	137,000	23	10
		Press Forged at —80 F+			
0.09	7.0	45,200	79,500	33	21
	20.8	144,000	147,000	6	3
0.03	2.5	37,100	71,000	65	43
	17.0	98,500	122,500	40	23
	32.0	162,000	171,000	21	9

* Not tempered.
\+ Tempered at 650 F.
** Tempered at 800 F.

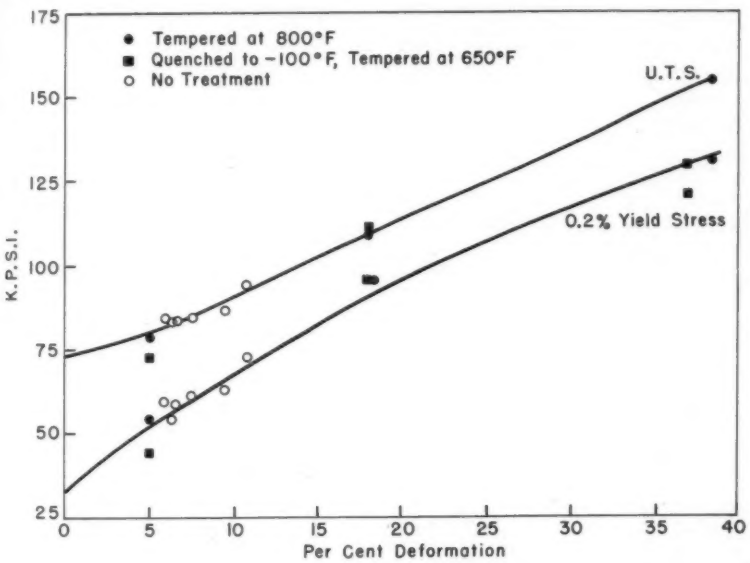


Fig. 9 — Mechanical properties of 18/8 stainless steel CF8C (high carbon) press forged at room temperature.

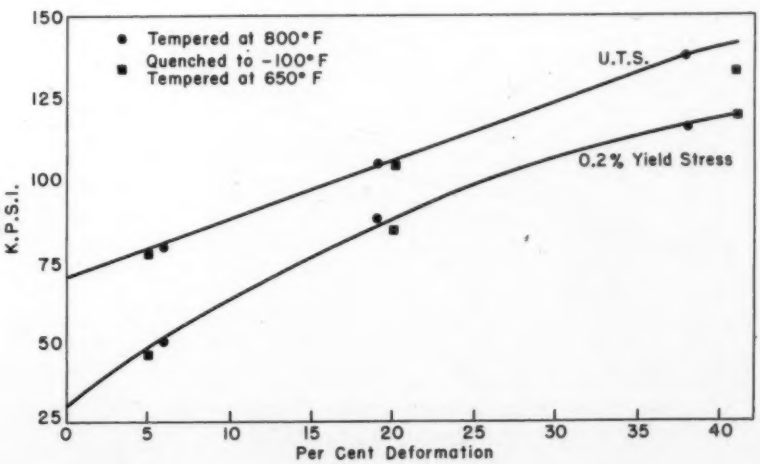


Fig. 10 — Mechanical properties of 18/8 stainless steel CF8C (low carbon) press forged at room temperature.

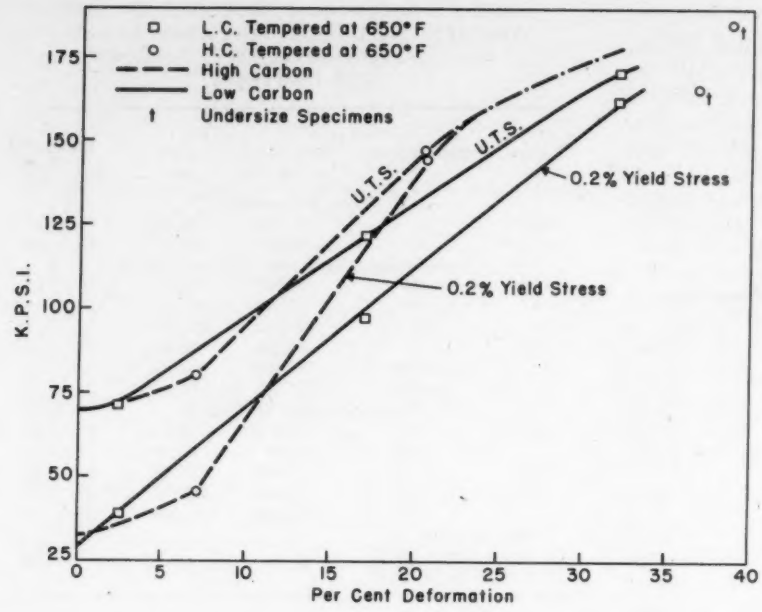


Fig. 11 — Mechanical properties of 18/8 stainless steel CF8C (high and low carbon) press forged at —80 F.

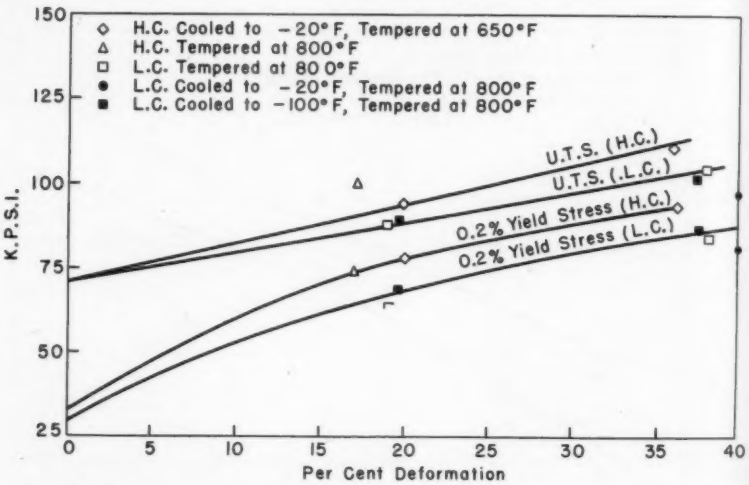


Fig. 12 — Mechanical properties of 18/8 stainless steel CF8C (high and low carbon) press forged at 450 F.

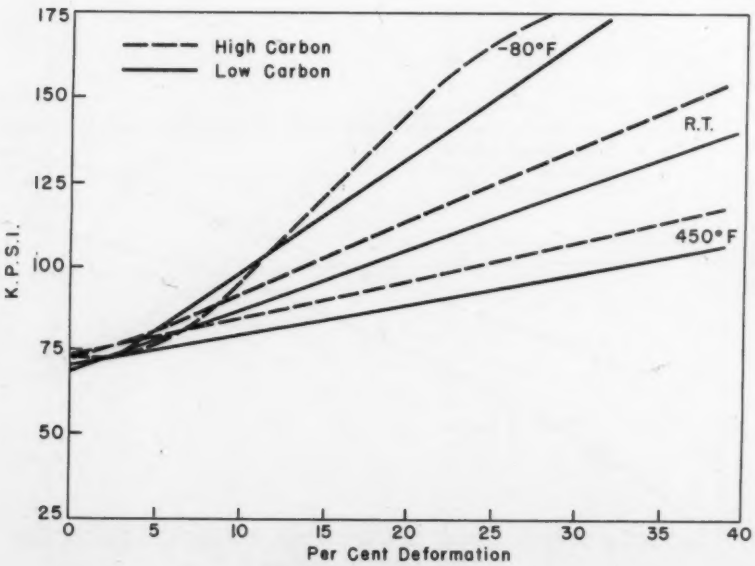


Fig. 13 — Press forging temperature effect upon ultimate tensile strength of 18/8 stainless steel CF8C.

some of the gaps. In spite of this condition, the problems of alloy selection and treatment are susceptible to some rational analysis so that likely candidate metal compositions can be selected for this process and their properties evaluated by testing.

When considering steels, the basic strengthening mechanisms are work hardening, precipitation hardening and martensite or bainite transformation hardening. Of these the latter two are most effective and can be found together in certain alloys. The effects can be additive.

In order to plan a press forging, heat treatment sequence and select an alloy amenable to improvement, the isothermal transformation diagram—or, better yet, the continuous cooling curve—is required. The curves must provide sufficient time to quench the casting to the forging temperature without getting undesirable transformation products. A ferrite nose, past which the casting can be quenched easily, and a pronounced gap between the ferrite and bainite noses is necessary.

The allowable time before transformation begins is barely sufficient to perform the quenching and forging operations in 4340 alloy. Most of the moderately alloyed steels have transformation curves of this general configuration, while the retardation of some high alloys, such as 17-4PH and 18/8 is so great that no isothermal diagram has been determined.

For the low alloy materials the application of strain causes a shift in position of the various boundary lines of the isothermal transformation curve. In general, these changes are toward shorter times and higher temperatures for any reaction. Bhattacharyya and Kehl[7] have illustrated this affect for various steels, including 4340. The start of martensite transformation is also elevated by the application of stress.

Problems in the Manufacture of Forged Castings

The basic configuration produced by the designer is altered to yield the best shape for forging. The problems encountered here are similar to those faced for conventional forgings and include:

1) Provision of forging draft.
2) Avoidance of hollow-sections and undercuts.
3) Estimation of metal flow.
4) Calculation of forging pressures.
5) Allowance for springback.

The choice of forging method probably is one point of divergence from conventional practice. Since the ductility of metal below the recrystallization temperature is much less than for hot forging, slow forging methods would be favored to give the effect of greater ductility. In addition, the dies used in this process may be heated electrically and, in the interest of reasonable heater element life, minimum impact loads are desirable. For these reasons press forging methods are favored over drop forging.

The die design can be troublesome if high die temperatures are required or if high flow stress are encountered. At warm temperatures the flow stress of most steels is more than 20 per cent greater than the 0.2 per cent offset strength, and it increases as the forging deformation increases. This imposes size limitations on a part and calls for high strength, nondeforming dies and high capacity presses. Deformations greater than the ductility level at the deforming temperature should not be attempted.

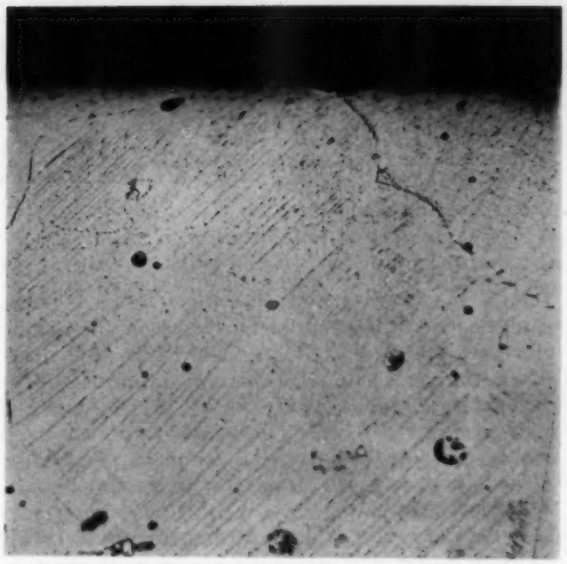


Fig. 14 — Grain orientation effect upon formation of martensite at site adjacent to press forged surface. Kalling's reagent etch. 250×.

Materials to resist deformation, ranging from cobalt-base heat-resistant superalloys to chilled white iron, have been used successfully in their cast form. If the part to be forged has varying metal sections which still require the same percentage deformation, multistage forging may be necessary. This should be avoided if possible.

Good Design

The cast forging blank is designed so that the forging deformation will result in a finished part of the

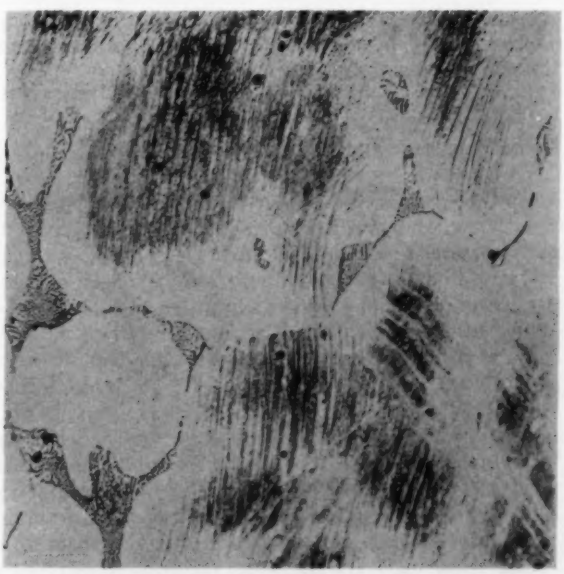


Fig. 15 — Grain orientation effect upon formation of martensite at site in center of press forged section. Kalling's reagent etch. 250×.

correct size and shape. Fortunately, good casting design and good forging design have much in common. These points should be observed when casting the forging blank:

1) Uniformity of metal sections.
2) Easy blending of different sections.
3) Directional solidification.
4) Accessibility of metal masses.
5) Avoidance of stress concentration.
6) Low cost production.

Since the forging process by its nature is more restrictive regarding possible shapes than the casting process, casting of a forging blank should be relatively simple. The most important problems will be the elimination of centerline shrinkage and control of grain size. At the working temperatures employed closure of voids would not be complete, so that the advantages noted by Sims[1] for imperfect castings would not be important.

A reasonably fine-grained casting, free from coring and segregation, will help minimize cracking and nonuniform surface conditions during deforming. The fine-grained structure may possess slightly higher ductilities and does possess better impact resistance, but would be primarily important to minimize segregation, although a higher flow stress will be encountered during deforming.

CONCLUSIONS

Experimental evidence pointing toward worth-while improvements in mechanical properties of castings by press forging at selected temperatures has been presented and discussed. The potential advantages are thought to result because of strain-induced changes in metallurgical structure, principally in the orientation, location and size of martensite, and precipitation hardening nuclei.

Mechanical improvements, when compared to the same composition hardened by heat treatment, where applicable, include higher ultimate tensile strength, yield strength, proof strength and hardness, without the same degree of ductility loss.

The directionality of properties, as customarily encountered in forgings made from billet stock, would be minimized by the cast preform.

The closure of defects in the casting would probably constitute a minor gain only.

Localized application of forging stress can be used to strengthen specific problem areas. It is conceivable that such forging could be done with simple flat dies.

Die material requirements call for specific compositions if working with heated dies or high flow stress cast-forged parts. Casting of such dies is feasible and practicable.

Limitations on design are more akin to forgings than castings, so that subsequent machining often would be indicated. Use of uniform sections and simple shapes is in the interest of process economy.

ACKNOWLEDGMENT

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REFERENCES

1. C. E. Sims, "Cast and Forged Steels Compared," *Steel,* June 4 1945.
2. G. T. Harris and W. H. Bailey, "Effect of Warm Working on an Austenitic Steel (G18B)," British Iron & Steel Inst. Special Report No. 43, pp. 60-67, 1952.
3. E. M. H. Lips and H. Van Zuilen, "Improved Hardening Technique," *Metal Progress, vol. 66,* p. 103, August 1954.
4. D. J. Schmatz and V. F. Zackay, "Mechanical Properties of Deformed Metastable Austenitic Ultra-High Strength Steels," A.S.M., 1958.
5. E. B. Kula, J. M. Dhosi, "Effect of Deformation Prior to Transformation on the Mechanical Properties of 4340 Steel," A.S.M., 1958.
6. A. Murphy, L. Clark and W. Rostoker, "Improvement of Castings by Press Forging," AFS TRANSACTIONS, *vol. 66,* p. 105, 1958.
7. S. Bhattacharyya, G. L. Kehl, "Isothermal Transformation of Austenite under Externally Applied Tensile Stress," *A.S.M. Transactions, vol. 47,* 1955.

CONTINUOUS CARBON INJECTION

By J. E. Wilson and R. C. Shnay

ABSTRACT

The total carbon content of gray iron melted in an acid cupola depends on a multitude of factors, by far the most important being the carbon content of the metallic charge. The pig iron usually has a higher carbon content than any other component of the charges, so most foundries depend primarily on changes to the per cent pig iron to adjust the final carbon content. Since this material is the most expensive component, and may be in short supply during periods of national emergency, there is an obvious need for an alternative method of controlling final carbon content particularly at the higher levels.

Many types of gray iron castings require a high carbon content to ensure that sufficient fluidity is available to produce a sound casting. In other cases, a high carbon content is necessary to produce the required properties. In all cases carbon control is essential, since this element has the greatest effect on mechanical properties and machinability.

This project was divided into four phases:

1) To obtain or devise dispensing equipment for the process.
2) To solve the problem of blockages and lance failures.
3) To determine practical methods of control.
4) To establish plant operating procedures for the process.

While the principles are simple and the basic requirements were readily established, many small points had to be cleared up before uniform and reliable operation could be assured.

INTRODUCTION

Control of carbon content in an acid cupola by methods, other than adjustments to the pig iron charge, can be divided into two general categories. Material such as carbon blocks, graphite electrodes, lump silicon and calcium carbide have been added to the coke bed. However, consistent results were not obtained.[1] Ladle additions of graphite powder have been found to produce an inoculating effect, but no more than 0.10 per cent to 0.15 per cent carbon could be recovered.[2] The advantages of ladle treatment are the inoculating effect and also the speed with which compositional adjustments can be made.

In a basic cupola operation satisfactory carbon control can be achieved by adjustments to the blast or slag composition. High carbon contents achieved in this way are obtained at the expense of a decrease in melting rate. Although the problem of carbon control is different in a basic cupola than in an acid operation, there is still a need for a method of obtaining high carbon contents consistently with no decrease in melting rate.

The most successful solution to this problem is the carbon injection process. This was first described by Spangler and Schneidewind.[1] These authors reported carbon increases of up to 1.9 per cent with recovery efficiences ranging from 50 per cent to 100 per cent depending on the injected material. A high degree of reproducibility was obtained and data were presented to show that gray iron treated in this manner showed "a greater degree of uniformity and predictability of properties." The authors concluded that carbon injection produced the same effect on mechanical properties as an inoculation treatment.

Although Spangler and Schneidewind were primarily concerned with the metallurgical advantages of carbon injection, there is also a decided economic advantage in that at least a part of the pig iron charge can be replaced by iron or steel scrap. The resulting metal is at least equivalent if not higher in quality than if it had been produced by entirely orthodox means.

Injection Process

In the carbon injection process, fine particles of graphite or other carbonaceous material are carried by a stream of relatively inert gas, such as nitrogen, through a graphite tube or lance immersed in the liquid metal. The carbon particles are delivered to a point below the surface of the metal where they are readily dissolved, and in this way high recoveries are obtained. The equipment and general method are similar to those used for desulfurizing by the injection of calcium carbide, lime or aluminum magnesium alloys.

Although the carbon injection process was first reported publicly by Spangler and Schneidewind,[1] the technical feasibility of this process was independently established by J. E. Rehder in 1952. Trials then showed the technical possibilities and a patent search was made preparatory for patent application. However, patent attorney opinion was that while injection of carbon into cast iron was novel, much older patents dealing with carburizing liquid steel made it

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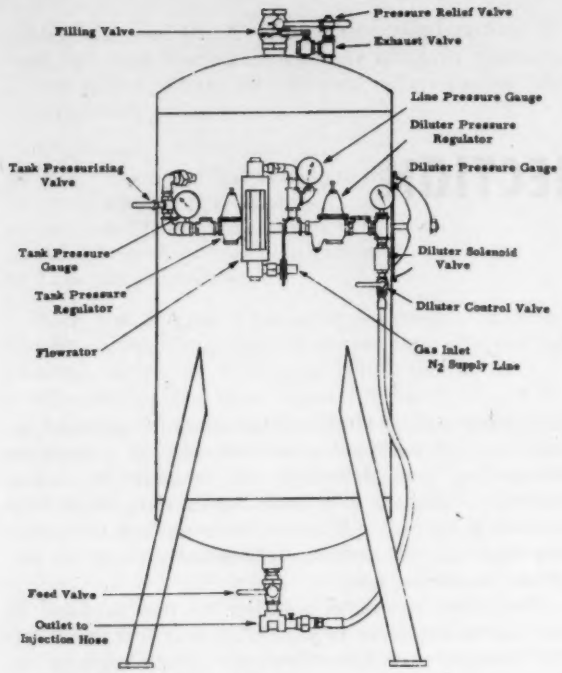


Fig. 1 — Details of injector.

doubtful that a patent could be obtained or defended if attacked.

Because of pressure of other work further development was slow, but in 1954 preliminary plant tests were carried out by A. L. Cooper and P. Biron of National Iron Div., Canada Iron Foundries Ltd. to establish carbon injection on a continuous basis. These tests were carried out in addition to regular production duties and progress was slow. After about three months the work was suspended partly due to the pressure of regular duties, but also because of the difficulty in maintaining uniform feed rates and the problem of rapid lance erosion.

The Research and Development Div. was organized in 1957, and the work on carbon injection was resumed, although a United States company[3] had been granted a U.S. patent on carbon injection of cast iron.

The experience of the previous investigators, and an examination of the conditions necessary for metallurgical and economical efficiency, led to the subdivision of this project into four distinct phases.

1) To obtain, or if necessary devise, dispensing equipment capable of stable operation at feed rates of 2 to 3 lb/min.
2) To solve the problem of lance failures and blockages and limit lance consumption to a gradual and continuous process.
3) To determine practical methods of control so that the final carbon content of the metal could be held within suitable narrow limits despite fluctuations in the melting rate and the composition of the metal leaving the tap hole.
4) To establish a procedure for operating the carbon injection process in the plant.

The first phase of the project, to find suitable dispensing equipment, was carried out at National Iron Div., and a start was also made on finding a solution to the problem of lance failure and blockage. In Aug. 1957, the program was transferred to the Delavaud Shop in the Trois Rivieres plant. While trying to determine a suitable test setup at Trois Rivieres a solution was found to the lance problem. The test program then proceeded to the third phase and control procedures were established by Dec. 1957.

Operational equipment was purchased and installed during April 1958. Operating procedures were then developed and the plant personnel were trained to use the equipment. The equipment and process were formally handed over to the plant staff in May 1958, and carbon injection has been a regular part of the operation since that date.

EQUIPMENT AND MATERIALS

Dispensing Equipment

Basically the dispensing equipment is simple, and consists of a pressure vessel, mixing assembly and a delivery system. At the low feed rates of 2 to 3 lb/min used in this investigation, the net pressure in the delivery system is low, since the metal itself exerts a back pressure due to the ferrostatic head which increases with immersion depth. This becomes an important factor when low feed rates are used since the net pressure may not be any more than 1 psi.

Because of this low pressure, any line obstruction due to surface roughness, sharp bends in the hose or changes in section at the junction of a fitting and the hose, can cause blockage and stop the flow of graphite. The low pressures necessary for these feed rates introduce additional problems in the design of gauges and regulators.

After some experimentation an American company was able to supply suitable equipment. Diagrams of this equipment are shown in Fig. 1.

Injection Lances

The injection lances are graphite "fluxing tubes" with an outer diameter of 2 in., an inner diameter of ½-in. and length of 9 ft. The inner diameter is threaded at one end to enable connection with the hose from the dispenser. A fitting is inserted at the end of the hose leading from the dispenser so that it can be threaded into the lance. A suitable fixture was mounted in the well or forehearth to hold the lance in position.

Graphite

Electric furnace graphite was used which had the following approximate sieve analysis:

Retained on 20 mesh screen, %	5
Retained on 65 mesh screen, %	80
Retained on 100 mesh screen, %	15

Later tests showed that equivalent results were obtained with an alternative electric furnace graphite having the following approximate sieve analysis:

Retained on 10 mesh screen, %	5
Retained on 65 mesh screen, %	80
Retained on 100 mesh screen, %	15

The particle size and particle size distribution were found to be quite critical in obtaining satisfactory

flow. Coarser particles tended to give unstable operation with wide fluctuations in feed rate while finer particles did not flow properly and produced blockages in the system.

The nitrogen used in these tests was the dry oil-pumped grade.

TESTS AT NATIONAL IRON DIVISION

The major object of this series of tests was to determine whether the dispenser was capable of continuous operation at a feed rate of 2 to 3 lb of graphite/min. A secondary objective was to find a solution to the problem of lance erosion and blockage.

The dispenser was placed on a platform scale located on the charging floor to one side and slightly forward of the cupola. The graphite delivery hose was lowered through a hole in the charging floor directly in front of the cupola so that the hose was suspended over the front slagging spout. Injection was carried out in an enlarged well located forward of the slag dam. The inside dimensions of this well were approximately 20 in. x 20 in. x 20 in. A hood was designed to fit over the well, and a hole was cut in the hood to accommodate the lance. A simple hoist consisting of a ratchet winch pulley and cable was installed to raise and lower the lance. A sketch of the complete setup is shown in Fig. 3.

The first two tests were carried out merely to check the installation, and the charge was chosen to obtain a total carbon content of about 3.5 per cent at the tap hole. Carbon pickup was about 0.17 per cent in the first test and 0.39 per cent in the second. The next test was a dry run for purposes of calibration. The graphite was discharged into a bucket. Different settings were used, and the amount of graphite discharged/min was determined for each setting. This procedure cannot lead to a complete calibration since the pressure due to the head of liquid metal varies with immersion depth and can only be estimated in a dry run, but it did help to establish suitable settings for the operation.

During subsequent tests the pig iron was reduced by 50 per cent, and final carbon contents were held in the range of 3.65 per cent to 3.80 per cent.

These tests showed that the dispensing equipment could be operated at feed rates as low as 1.75 lb/min for an indefinite period. The only interruptions were due to lance erosion and blockages, which occurred at 15 to 20 min intervals.

Lance Tests

Coatings of refractory cement and sodium silicate were tested to see if they could improve lance life. In addition, tests were carried out with the tapered lances. None of these measures produced any significant improvement. Alternative lance materials were also investigated and discussed with representatives of two refractories companies, but none were found that warranted testing.

The appearance of the used lances suggested that failures were caused by a cavitation or pitting type of erosion, which led to the formation of pinholes. When the pinholes extended through the wall, the gas pressure was released, but at this stage the holes were still too small to allow passage of the graphite.

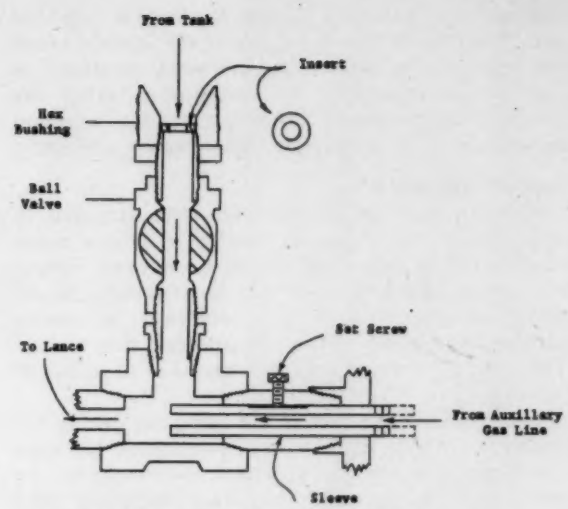


Fig. 2 — Cross-section of mixing assembly.

The graphite then accumulated in the lance since there was no pressure to force it through, and the subsequent blockage led to failure of the lance. The most likely cause of this type of erosion was the turbulence of the liquid metal along the walls of the lance.

The above hypothesis was verified by leaving a lance immersed in the well for several hours with no flow of nitrogen or graphite. This exposure resulted in only a minor amount of lance consumption due to a smooth and continuous type of attack.

The next step should have been to do tests with varying rates of nitrogen flow to find the minimum flow rate which would still allow satisfactory injection. Unfortunately, the equipment was being used near the lower limit of nitrogen flow rate and the allowable range was too narrow for a conclusive test.

Drop tests and metallographic examination were

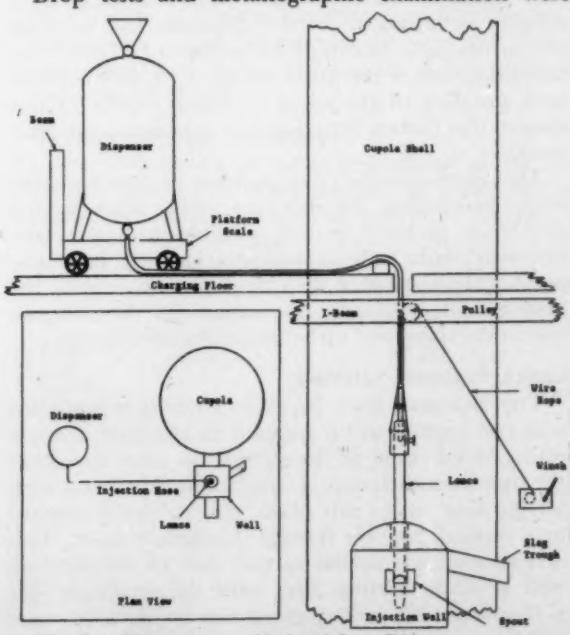


Fig. 3 — Test setup at National Iron Div.

done on pipe produced during the carbon injection tests. The level of carbon content of the injected metal was considerably higher than normally produced so that it was impossible to determine whether the observed improvements were due to the injection process as such, or merely the higher carbon content.

Carbon Recovery

The tap hole, spout and well were arranged in such a way that it was impossible to obtain metal samples before injection, therefore, accurate figures for carbon recovery could not be obtained. Nevertheless, some estimates were obtained by taking samples immediately before the injection tests started. The results indicated a carbon recovery of about 80 per cent.

Tests could not be carried out every day during this period because of production requirements. On some occasions the equipment was used successfully by the plant staff to overcome temporary operating difficulties. The procedure in these cases was to station a man at the injection well to inspect and manipulate the lance. At first two lances were used changing after every blockage. A new lance was inserted and in the meantime the blocked portion of the first lance was removed so that it could be put back in service after the second lance failed or became blocked.

Another method used was to break off the defective portion of the lance by striking it against the bottom or side of the well. Neither of these procedures could be considered practical for a permanent operation.

TESTS AT TROIS RIVIERES

The first objective of these tests was to solve the problem of lance failures and blockages. Once this was done, the next step was to determine suitable dispenser settings for a consistent carbon increase from 3.30 per cent to 3.65 per cent. Before any of this work could be done, it was necessary to establish a suitable arrangement for the dispenser and lance so that satisfactory tests could be performed without undue disruption of the plant operation. A well had not been installed in the spout of either cupola in this shop so that carbon injection had to be done in a forehearth.

The solution to the lance problem became apparent while establishing the injection setup. After solving the lance problem it was possible to inject continuously with little interruption for an indefinite period. Methods were then developed to control the carbon content despite fluctuations in the composition at the spout and variations in the melting rate.

Lance Problem Solution

The Delavaud Shop at Trois Rivieres is equipped with two cupolas which are used on alternate shifts. A tilting drum type of forehearth is used to collect the iron from each cupola. The north cupola was used for the first test in this plant with the lance inserted in a vertical position through forehearth cover. This arrangement was similar to that used in the injection well at National Iron Div., with the exception that a chain-operated trolley crane was available to raise and lower the lance.

Because the forehearth must be tilted to transfer iron to the ladle, it was necessary to cut a slot across the forehearth cover to avoid disturbing the lance. The results of this test were approximately the same as at National Iron Div.

Slant Immersion

The south cupola had to be used for the second test, and it was found that the hose was too short to immerse the lance in a vertical position. At this stage it became apparent that the problem of lance failures and blockages might be solved by immersing the lance at an angle of about 45 degrees. A test proved this was indeed the case and excellent results were obtained. The reasoning that led to this solution is illustrated by the schematic diagram shown in Fig. 4. Figure 5 shows one lance after vertical immersion and another after slant immersion.

At first the slant immersion was obtained by suspending the lance from one end, the other end floating below the surface of the metal in the forehearth. This procedure was not satisfactory and a fixture was welded to the end plate of the forehearth which held the lance in the correct position at an angle of 45 degrees. Some mechanical troubles were experienced with this at first, but were overcome in later tests. Figure 6 shows the final design of the lance fixture, and Fig. 7 shows the lance and fixture in service.

Lance consumption with slant immersion averaged about 10 in./hr, as compared to 24 in./hr with vertical immersion. This represents a savings of approximately 9 cents per ton, but the continuity of operation resulting from slant immersion is of far greater importance.

Operating Problems

Many operating problems, mainly of a minor nature, were encountered during these tests. Methods to overcome these problems were devised and described in operating instructions issued to the plant.

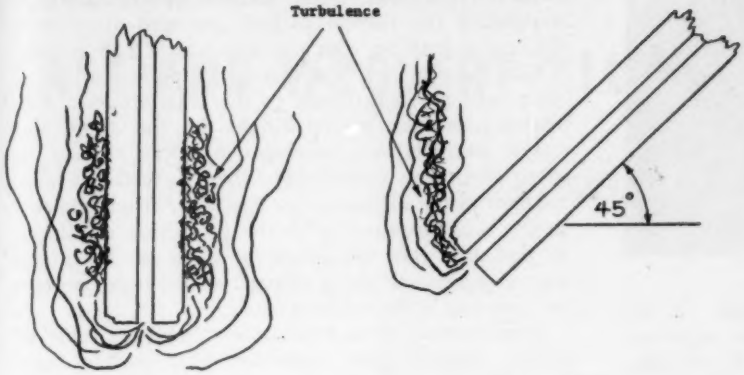
Moisture Conditions. Moisture entering the dispenser is absorbed by the graphite and results in blockages at or near the feed valve. Initially the dispenser was located between the two cupolas slightly above the level of their wells. The heat from the cupola and the spray from its shell cooling system led to considerable difficulty with wet graphite due to condensation and also due to spray from the cooling water entering the dispenser during filling. The dispenser was then moved to the charing floor.

Even in this location, there was still occasional condensation inside the dispenser, particularly during cold weather. Methods were developed to blow out the moisture so that blockages did not occur.

Hose and Fittings. It was found that ordinary rubber hose could not be used to carry the graphite from the dispenser to the lance. The rough inner surface and sharp bends led to blockages during service. Fittings had to be specially machined to ensure that there were no lips on which the graphite could build up since such accumulations also led to blockages.

Regulators. The regulators furnished with the dispenser were found to deteriorate quickly due to the

Fig. 4 — (below) — Schematic diagram illustrating solution to the lance erosion problem. Left view shows vertical immersion; right view shows slant immersion.



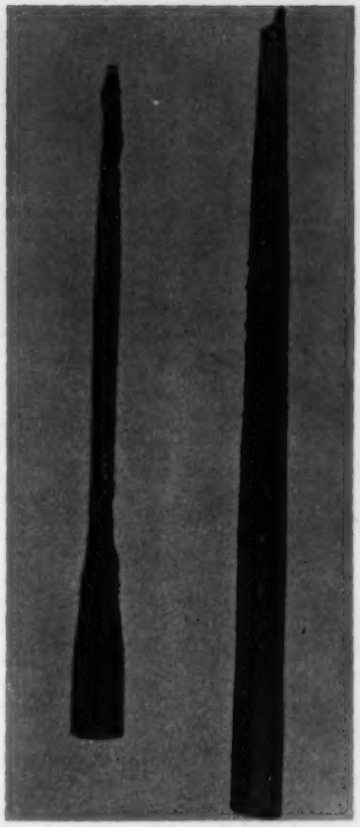


Fig. 5 — (right) — Left view shows erosion after vertical immersion; right view shows erosion after slant immersion at approx. 45 degree angle for the same amount of immersion time.

abrasion of the rubber valve cones by the graphite. Several other types of regulators were tried and a more suitable type was specified.

Disc Insert — Aperture Size. Initial experiments were carried out with a ¼-in. aperture. However, it was found that $\frac{3}{16}$-in. aperture led to a more stable operation.

Position of the Diluter Sleeve. The position of the diluter sleeve was found to be critical. If it is advanced beyond the critical position the graphite builds up around the sleeve and the feed rate is limited to the amount of graphite picked up by the nitrogen from the sleeve opening. If the diluter sleeve is withdrawn from the critical position, the venturi effect is lost and the graphite does not flow into the hose. The optimum position of the tip of the diluter sleeve was found to be located at the center line of the feed valve piping or slightly ahead of it.

CARBON CONTROL

For a given initial carbon content the control of final carbon depends on the following quantities:

a) *Feed Rate.* The weight of graphite powder dispensed by the equipment per min.
b) *Carbon Recovery.* The ratio of the weight of injected carbon dissolved in the metal to the weight of graphite injected, expressed as a percentage.

These quantities are affected by various operational factors, some of which affect more than one quantity.

The feed rate depends primarily on the tank pressure and flowrator settings. At any given values of these settings, the feed rate varies with the length of hose, the difference in elevation between the dispenser and the forehearth and finally the immersion depth. Because of the back-pressure due to the ferrostatic head (height of metal above the bottom tip of the lance), the net pressure in the hose and lance decreases with increasing immersion depth and leads to a decrease in feed rate. The length of hose and the difference in elevation are determined when setting up the equipment and are usually left unchanged.

Carbon recovery depends on the per cent carbon in the powder, the immersion depth and the chemical analysis of the metal before treatment. The carbon content of the graphite powder used in these tests was about 99 per cent so that this factor could be neglected. With increasing immersion depth the graphite has a greater opportunity to dissolve in the metal. This results in increased recovery and tends to compensate for the loss in feed rate caused by the increased head of metal.

Carbon Content Variation

The fluctuations in metal temperature at the forehearth were not serious enough to cause any appreciable variation in carbon recovery. Changes in

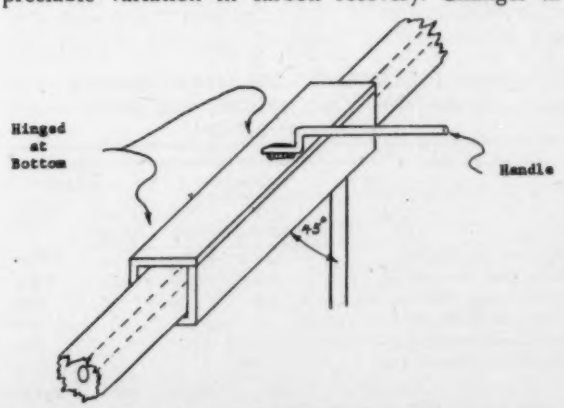


Fig. 6 — Final design of lance fixture.



Fig. 7 — Lance and fixtures in service.

carbon content were the major variations in the chemical composition of the metal before injection. The per cent carbon varied as much as 0.30 per cent, but the increased recovery rate at lower carbon levels tended to even out these fluctuations to give self-regulation.

The actual carbon pickup depends on the feed rate and carbon recovery. Therefore the major operating factor that affects carbon pickup is the immersion depth, which acts on both components. It was found that an average immersion depth of 18 in., with a maximum variation of ±4 in., resulted in quite uniform values for carbon pickup and good control of the final carbon content.

The best method of maintaining this control of the final carbon content is of course chemical analysis. However, in most plants the results from the analytical laboratory usually arrive too late for effective action. Chill and fluidity tests taken every 15 min, although affected to a lesser degree by factors other than total carbon content, were found adequate for short term control if backed up by regular chemical analyses taken at least once each hr.

Carbon injection at the low feed rates used in these tests is usually accompanied by higher carbon recoveries than when a high feed rate is used. The geometry of the forehearth may also be important since it affects the degree of mixing. However, these two factors are difficult to assess and are not usually changed to any appreciable extent in any given operation.

Carbon Recovery

Long term tests were not carried out to obtain accurate estimates of carbon recovery, although this was determined at frequent intervals during the course of the investigation. With slant immersion the carbon recovery ranged from about 82 per cent to about 95 per cent. After suitable procedures were established the carbon pickup averaged about 0.30 per cent to 0.40 per cent at a feed rate of 2.75 lb/min.

The melting rate was changed frequently during these tests by adjustments to the blast volume. This was necessitated by conditions in the shop, but simple methods were developed to adjust the carbon injection process accordingly. These consisted of small changes in the nitrogen flow rate and/or adjustments to the tank pressure. With the prevailing conditions in this shop each change of ¼ psi in the tank pressure compensates for a change of about 5 per cent to 7 per cent in the melting rate.

Good control of carbon content has been achieved during regular plant operation with carbon injection. The average variation of carbon contents for 20 successive 8 hr shifts was 0.18 per cent, and experience since has shown that carbon content control is now consistently better than before, due to the self-regulating action.

TABLE 1 — STATEMENT OF TYPICAL CUPOLA OPERATION COSTS BEFORE AND AFTER CARBON INJECTION*

	Before Carbon Injection		After Carbon Injection	
	Tons Charged	Total Cost	Tons Charged	Total Cost
Pig iron, $64/net ton	50.0	$3200	15.0	$ 960
Scrap iron, $40/net ton	49.4	1976	84.1	3364
Ferrosilicon, $150/net ton	0.6	90	1.2	185
Nitrogen, $1.30/100 ft³				18
Graphite powder, $250/net ton				109
Injection lances, $14 ea.				14
	100	$5266	100	$4650

*Acid Cupola Melting at 20 tons/hr.

DISCUSSION

It has been established by other research workers, namely Spangler and Schneidewind,[1] that this process has an inoculating effect which tends to increase the strength of the final product. Improvements in properties could not be definitely verified in these tests since the final carbon content is now at a considerably higher level than it used to be. This move to higher carbon contents was made to improve the quality of the product. Regardless of inoculating effect, a better product is now being made at a lower cost.

At present the limiting factor in application of carbon injection is the increase in sulfur content as the pig iron charge is decreased. Plans are now underway for simultaneous desulfurization and carbon injection which will increase the savings with no decrease in quality.

ECONOMIC CONSIDERATIONS

A typical breakdown of costs is presented in Table 1.

A saving of $6.16/net ton charged is realized, and the final carbon content would be the same at about 3.60 per cent to 3.70 per cent.

With simultaneous desulfurization the pig iron could probably be cut to 10 per cent and the net savings would then be about $6.60/net ton of metal charged. Further savings could be obtained by developing methods to ensure fuller utilization of the lances and also methods to use a cheaper gas such as compressed air, but these could only produce minor savings in relation to those expected from desulfurization.

REFERENCES

1. G. E. Spangler and R. Schneidewind, "Increasing Carbon Contents of Cast Iron by Ladle Injection," AFS TRANSACTIONS, *vol. 63*, pp. 533-539, 1955.
2. H. T. Argus, "Carbon Pickup," *B.C.I.R.A. Journal of Research and Development, vol. 4*, No. 6, June, 1952.
3. Air Reduction Co.

CAST AND WROUGHT ALLOY STEELS

elevated temperature properties

By R. K. Buhr and W. A. Morgan

ABSTRACT

The cast and wrought properties at room and elevated temperatures are reported for two known and two new low alloy high strength steels. Three of the steels display secondary hardening, and the mechanism involved in this phenomenon was investigated.

The castings are shown to possess many desirable properties over their wrought counterpart, especially when used at elevated temperatures where the differences in mechanical properties become quite small. In the unnotched condition, longitudinal fatigue properties are superior to those found in castings, but in the notched condition the cast and longitudinal fatigue properties are nearly identical.

Secondary hardening was investigated using electron microscopy and electrolytic carbide extraction techniques. These tests verify that secondary hardening is due to the precipitation of fine alloy carbides.

INTRODUCTION

Recently great interest has been shown in the properties of low alloy alloy steels for use at temperatures up to 1100 F (590 C). The most useful alloys exhibit secondary hardening on tempering the quenched alloy. These steels can be tempered at much higher temperatures than the standard A.I.S.I. 4340 type steels and still have equivalent hardnesses, and it has been claimed that more complete stress relief is thus obtained in a cast or wrought part.

A large amount of information has been published on the properties of the wrought steels at elevated temperatures, but the cast steels have not been investigated so extensively. This paper presents the results of some initial work being carried out on the room and elevated temperature mechanical properties of some low alloy steels in the cast condition. The properties of these same compositions in the wrought condition have also been determined so that comparisons may be made.

In order to obtain a more complete understanding of the relationship of the properties of these cast steels to their microstructure and heat treatment, a brief account is given of some experimental work on the electron microscopy of these steels after tempering at different temperatures.

EXPERIMENTAL PROCEDURE

Melting and Casting

The heats were made in a basic-lined direct arc electric furnace of 550 lb capacity. Double slag procedure was employed, and all heats were deoxidized in the ladle with 2 lb/ton of aluminum. Half of each heat was poured into ingots, while the remainder was poured into a standard four-leg keelblock casting.

Chemical Composition

The analyses of the steels used in this investigation are shown in Table 1.

Forging and Rolling

The ingots were cropped and then forged from the 5½-in. ingot diameter to a 4½-in. square billet, and then to a 2¼-in. thick slab. This slab was subsequently rolled to a 1-in. thick plate. The reduction obtained by this hot working was approximately 4 to 1.

Machining

Tensile, impact and fatigue bars were rough machined from either the keelblock legs or the 1-in. plate (longitudinal direction). At least 0.030 in. of stock was left on the diameter of the tensile and fatigue bars, while the Charpy impact bars were left unnotched. The bars were heat treated, and then finish machined to the final dimensions. Drawings of the Krouse fatigue bars and the tensile test bar used

TABLE 1 — ANALYSES OF STEELS USED

Element, %	Heat A	Heat F	Heat H	Heat J
C	0.38	0.39	0.38	0.38
Mn	0.61	0.65	0.47	1.12
Si	0.19	0.44	0.78	0.22
S	0.019	0.014	0.019	0.019
P	0.010	0.008	0.012	0.012
Cr	0.95	1.34	4.97	2.04
Ni	1.68	—	—	—
Mo	0.36	0.47	0.88	1.71
V	—	0.32	0.57	0.15

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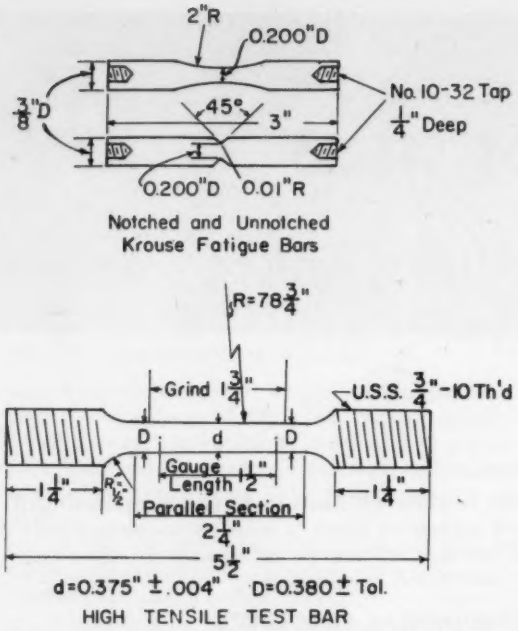


Fig. 1 — Drawings of fatigue bars and tensile test bar used.

are shown in Fig. 1. A standard Charpy V notch impact bar was used for the impact tests.

Jominy hardenability tests were performed on standard 1-in. diameter specimens, while the dilatometer specimen was 5/16-in. diameter and 2.570 in. long.

Heat Treatment

In order to determine the optimum tempering temperature for heats F, H and J, 1x1x2 in. samples from each of these heats were austenitized at either 1950 F (1060 C) (heats F and J) or 1900 F (1035 C) (heat H), and either air cooled (heats H and J) or oil quenched (heat F), and then tempered at different temperatures. The hardness tempering temperature curves obtained are shown in Fig. 2.

The rough machined test bars were heat treated as follows:

Heat A —	Normalized	1650 F	
	Reheated to	1475 F	Oil quenched
	Drawn at	750 F	5 hr — air cool
	Redrawn at	750 F	2 hr — air cool
Heat F —	Normalized	1700 F	
	Reheated to	1950 F	Oil quenched
	Drawn at	1150 F	5 hr — air cool
	Redrawn at	1150 F	2 hr — air cool
Heat H —	Normalized	1700 F	
	Reheated to	1900 F	Air cool
	Drawn at	1050 F	5 hr — air cool
	Redrawn at	1050 F	2 hr — air cool
Heat J —	Normalized	1700 F	
	Reheated to	1950 F	Air cool
	Drawn at	1125 F	5 hr — air cool
	Redrawn at	1125 F	2 hr — air cool

Jominy hardenability bars were austenitized at the same temperatures as those used in the heat treatment and quenched in a jet of water as in the standard procedure. The results are plotted in Fig. 3.

Dilatometer curves were obtained for each heat at heating rates of 150 C (302 F) per hr along with cooling rates of 150 C (302 F) per hr and 40 C (104 F) per hr. The dilatometer used was of the recording type with an expected accuracy of ± 2 C.

MECHANICAL TEST RESULTS

The tensile results both at room and elevated temperatures are shown in Table 2. Charpy V notch impact results at room temperature and at −40 F are listed in Table 3. Table 4 shows the endurance limit for the four steels in the cast and wrought condition for both notched and unnotched bars.

The critical points obtained from the dilatometer curves are given in Table 5.

The Shepherd fracture grain size of the steels were estimated by four different observers and are reported as averages in Table 6. There was close agreement in the individual estimations.

DISCUSSION

The purpose of this paper is to present and compare some of the room and elevated temperature properties of several alloy steels in the cast and wrought condition, and to explain briefly the mechanism of their strengthening.

Hardness-Tempering Temperature Curves

The resistance to tempering of both cast and wrought alloys was similar. Steel *H* gave the highest hardness on tempering at 1000 F (540 C), but at 1100 F (590 C) and 1200 F (650 C) steels *F*, *H* and *J* differed by only about five points Rockwell C. For

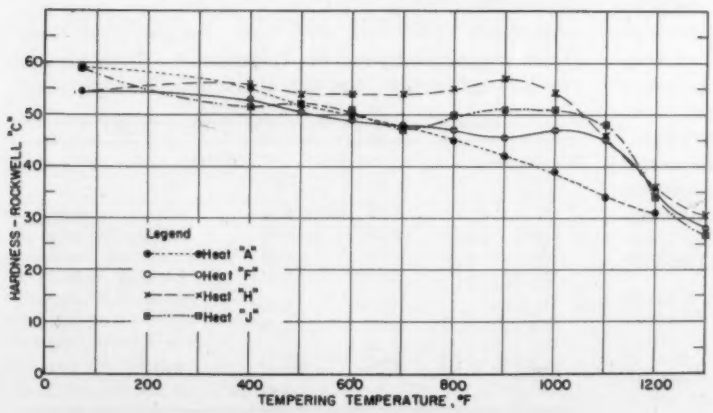


Fig. 2 — Hardness-tempering-temperature curves for the four heats investigated.

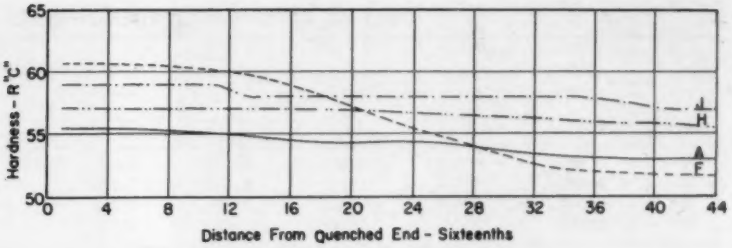


Fig. 3 — Jominy hardenability curves for heats *A*, *F*, *H* and *J*.

TABLE 2 — ROOM AND ELEVATED TENSILE PROPERTIES OF HEATS A, F, H AND J

Heat	Test Temp. F	U.T.S. Kpsi		0.2% Proof Stress		Elongation, %		R. A., %	
		Cast	Wrought	Cast	Wrought	Cast	Wrought	Cast	Wrought
A	70	213.1	215.1	202.0	205.4	5.65	10.0	19.6	41.85
	700	178.3	176.8	156.2	154.4	10.65	15.65	41.05	62.9
	1000	104.65	104.9	86.2	86.6	17.0	20.3	53.1	63.8
F	70	183.0	185.15	164.3	165.65	6.35	12.6	10.8	36.85
	1000	125.3	129.1	108.9	112.8	16.35	16.65	53.55	60.6
	1100	108.75	112.5	93.1	94.2	16.35	15.35	55.6	46.75
H	70	236.5	246.15	205.5	207.75	1.6	9.3	3.6	27.7
	1000	167.6	170.5	140.45	136.85	6.35	16.35	11.15	51.25
	1100	142.25	139.95	112.75	107.2	11.3	19.3	17.4	52.7
J	70	207.1	206.85	187.2	187.1	3.65	8.0	5.5	17.4
	1000	136.8	140.25	119.1	120.8	11.3	12.55	23.4	32.1
	1100	116.2	112.55	96.95	95.75	15.0	18.0	30.25	43.1

resistance to softening at temperatures in the range 1100-1200 F (590-650 C), steels *F*, *H* and *J* are similar, and all are superior to steel *A*.

Room and Elevated Temperature Tensile Tests

At room temperature, the ultimate tensile strengths and 0.2 per cent proof stresses of the cast and wrought alloys are almost equivalent. The elongation and reduction of area of the cast alloys are lower in all instances. The elongation and reduction of area values reported for the wrought steels were obtained on longitudinal samples from the forged and rolled plates. Transverse ductilities of wrought products would be expected to be lower.

At elevated temperatures 1000 and 1100 F (540 and 590 C) again, the ultimate tensile strengths and 0.2 per cent proof stresses do not differ appreciably. In steels *F* and *J* the elongation and reduction of area of the cast alloys have improved to such a degree that these properties approximate those of their wrought equivalents. In steel *H*, however, this large improvement in the tensile ductility has not occurred. Although this steel possesses the highest tensile strengths at 1000 F (540 C) and 1100 F (590 C), it should be noted that steels *F* and *J* have been tempered at temperatures 75-100 F (42-55 C) higher than *H*.

Fatigue Properties

Evans, Ebert and Briggs[1] have compared the fatigue properties of plain carbon and low alloy steels with tensile strengths of up to 170,000 psi. They found that the endurance ratio (fatigue strength divided by the ultimate tensile strength) for unnotched wrought steels was greater than the endurance ratio for cast steel. However, the endurance ratio for notched samples was about the same for both cast and wrought steels, having a value of about 0.27 to 0.32. The results of the present work are substantially in agreement with the results of these authors.

TABLE 3 — CHARPY V NOTCH IMPACT PROPERTIES OF HEATS A, F, H AND J

Heat	Charpy V Notch Impact — ft-lb			
	Room Temperature		—40 F	
	Cast	Wrought	Cast	Wrought
A	9.5	22.5	9.5	14.5
F	11.0	7.0	4.5	4.0
H	4.0	9.5	3.0	5.0
J	5.5	9.0	3.5	4.0

For the notched fatigue tests, however, the cast steel exhibits some superiority over the wrought. The ratio of the unnotched endurance limit to the notched endurance limit indicates a greater notch sensitivity in fatigue of the wrought steels. In comparing the secondary hardening steels, *F*, *H* and *J*, it is apparent that all steels, even though they differ in tensile properties, exhibit approximately the same endurance limits in the cast and wrought conditions for both notched and unnotched specimens.

The difference in the strength reduction factors of the cast and wrought steels does not appear to be a function of the grain size of the steels, since there is little apparent difference in grain size. This effect may be due to the directionality inherent in wrought steels.

Secondary Hardening Mechanism

Recent work[2,3] has shown that secondary hardening may be attributed to the formation of a coherent type alloy carbide precipitate at the temperature corresponding to the peak hardness in the hardness-tempering-temperature curve. This type of precipitate creates a high degree of lattice strain and inhibits slip processes in the steel. It has also been shown[4] that vanadium carbides are present in steels containing about 0.3 per cent vanadium on tempering at 1275 F (690 C).

TABLE 4 — FATIGUE PROPERTIES OF HEATS A, F, H AND J

Heat	Endurance Limit (E.L.) psi at 107 cycles				Endurance Ratio = E.L./U.T.S.				Strength Reduction Factor = Unnotched E.L./Notched E.L.	
	Unnotched		Notched		Unnotched		Notched			
	Cast	Wrought	Cast	Wrought	Cast	Wrought	Cast	Wrought	Cast	Wrought
A	77,000	81,000	30,000	26,000	0.362	0.376	0.141	0.121	2.56	3.12
F	64,000	105,000	55,000	49,000	0.355	0.568	0.300	0.265	1.16	2.14
H	68,000	113,000	60,000	50,000	0.288	0.459	0.254	0.203	1.13	2.26
J	62,000	107,000	43,000	51,000	0.300	0.517	0.208	0.246	1.44	2.11

TABLE 5 — CRITICAL POINTS FOR HEATS A, F, H AND J

Heat	Heating rate, C(F)/hr	Ac_1,C(F)	Ac_3,C(F)	Cooling rate, C(F)/hr	Ar_3,C(F)	Ar_1,C(F)
A	150(302)	720(1328)	787(1449)	150(302)	675(1247)	595(1103)
	150(302)	722(1332)	787(1449)	40(104)	680(1256)	605(1121)
F	150(302)	775(1427)	813(1494)	150(302)	723(1333)	680(1256)
	150(302)	775(1427)	813(1494)	40(104)	736(1357)	700(1292)
H	150(302)	833(1531)	875(1607)	150(302)	748(1347)	700(1292)
	150(302)	833(1531)	875(1607)	40(104)	780(1436)	760(1400)
J	150(302)	750(1382)	800(1472)	150(302)	382(720)	315(599)
	150(302)	750(1382)	800(1472)	40(104)	402(756)	313(595)

TABLE 6

Heat	Shepherd Fracture Grain Size	
	Cast	Wrought
A	6	5-6
F	4	4
H	5	5
J	3-4	3

Samples of steel were tempered above and below the secondary hardening peak temperature, and carbon replicas of the surface were examined with an electron microscope. The electron photomicrographs are shown in Figs. 4 to 9, inclusive.

It is clearly evident that at 1000 F (540 C) no separate alloy carbide is visible, and at this stage the Fe_3C has started to redissolve. Tempering at this temperature has resulted in the complete transformation of martensite to ferrite and cementite. After tempering at 1100 F (590 C) the iron carbides have been replaced by complex chromium-molybdenum carbides and vanadium carbide. At 1200 F (650 C) the carbides are more distinct and have increased in size, while at 1300 F (705 C) the carbides are easily

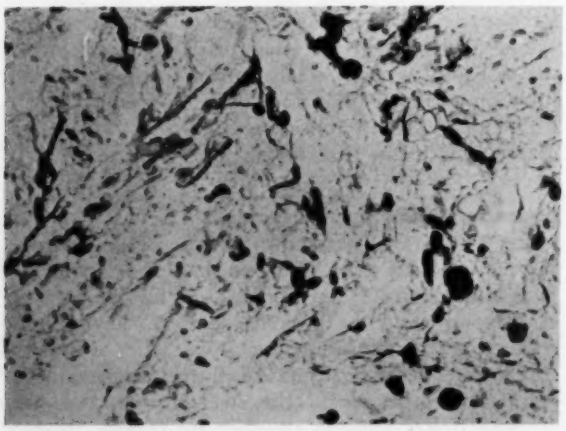


Fig. 4 — Electron photomicrograph of steel *F* sample after tempering 5 hr at 1000 F (540 C). The gray spots are cementite particles. 15,000 ×.

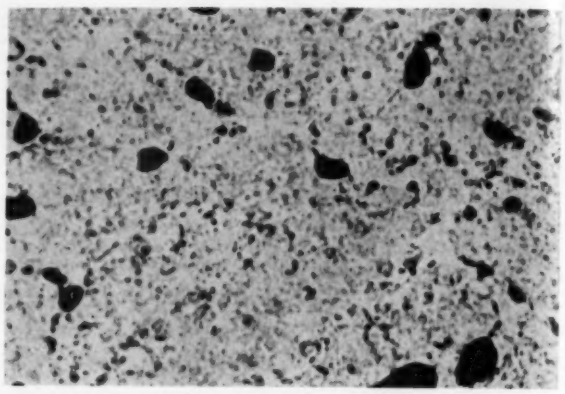


Fig. 5 — Electron photomicrograph of steel *F* sample after tempering 5 hr at 1100 F (590 C). Cloud-like vanadium carbides can be seen. 25,000 ×.

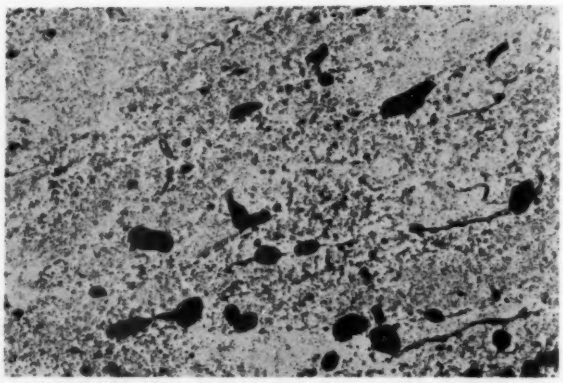


Fig. 6 — Electron photomicrograph of steel *F* sample after tempering for 5 hr at 1200 F (650 C). Fine vanadium carbides can be seen. 15,000 ×.

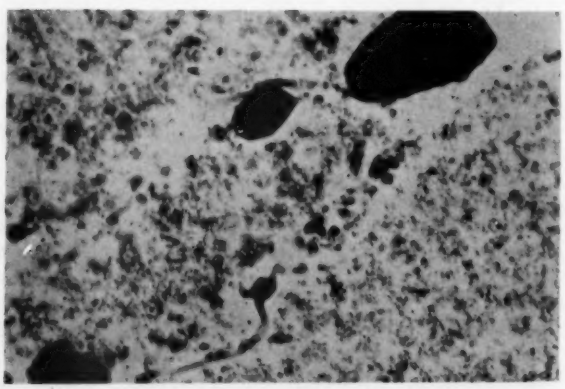


Fig. 7 — Electron photomicrograph of same sample as in Fig. 6. Tempering results at 1200 F (650 C) are shown in the formation of distinct vanadium carbides (small gray plates and thin dark lines). 50,000 ×.

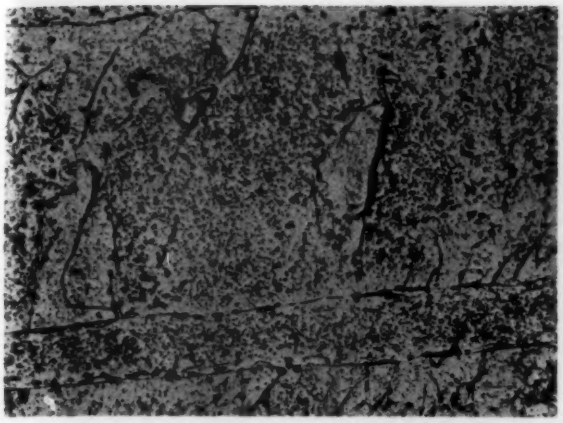


Fig. 8 — Electron photomicrograph of steel *F* sample after tempering at 1300 F (705 C) for 5 hr. Carbides within ferrite grains and concentrated at the grain boundaries can be seen. 15,000 ×.

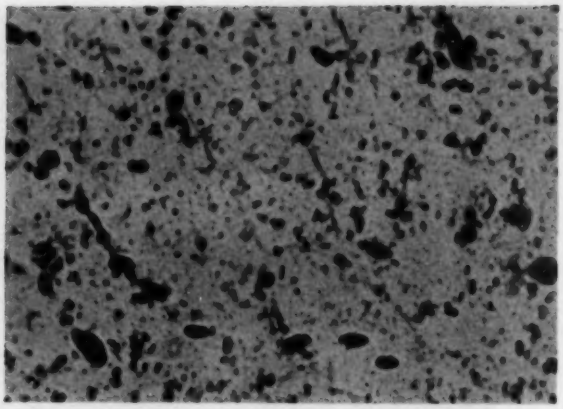


Fig. 9 — Electron photomicrograph of same sample as in Fig. 8. Thickness of the lines and darkness of the plates shows growth of carbides at tempering temperature of 1300 F (705 C). 50,000×.

resolved. These carbides tend to be concentrated at the grain boundaries (originally martensite needles) and also within the ferrite grains.

In order to obtain a quantitative picture of the carbides present in the steels after various austenitizing and tempering temperatures, samples of the steels were subjected to electrolytic carbide separation treatment. The results of the austenitizing and tempering temperatures effect on the weight of undissolved carbides are shown in Figs. 10 and 11. The need for the high austenitizing temperature is clearly shown in Fig. 10, in order to dissolve all of the stable alloy carbides.

Figure 11 shows that a minimum quantity of carbide remains after tempering at 1000 F (540 C), which corresponds approximately with the peak hardness in the hardness tempering temperature curves. The maximum amount of residue obtained at 850-900 F (455-480 C) corresponds to the minimum hardness in the hardness-tempering-temperature curves. Analysis of this residue shows it to be mainly Fe_3C. As the tempering temperature increases, the amount of undissolved carbide also increases, as is shown in Fig. 11, i.e., the coherent precipitate is forming a separate phase in the form of a noncoherent-type alloy carbide.

The maintenance of a high hardness and hot strength in the steels *F*, *H* and *J* may, therefore, be attributed to the presence of finely dispersed alloy carbides. As these carbides coalesce, the hardness decreases rapidly.

The microstructures of the cast and wrought steels used in this investigation are shown in Figs. 12 to 15, inclusive, at a magnification of 750 ×. The carbides are not resolvable at optical microscope magnifications, and require electron microscopy for resolution.

Further work is planned, to investigate modifications of standard deoxidation practices in an attempt to improve the room temperature cast ductility properties. Tests will also be carried out on the effect of vacuum ladle-degassing of the steels prior to casting.

CONCLUSIONS

The results reported indicate that these secondary hardening-type steels exhibit many desirable cast properties that could be utilized for service at elevated temperatures.

Although room temperature cast ductility is somewhat inferior to that of the wrought condition, this

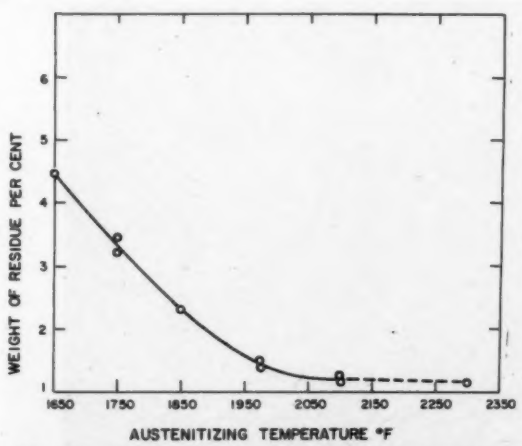


Fig. 10 — Graph showing austenitizing temperature vs. weight of residue (undissolved carbides) for steel *F*. Maximum resolution of carbides is obtained by increasing the austenitizing temperature to about 1950 F (1065 C).

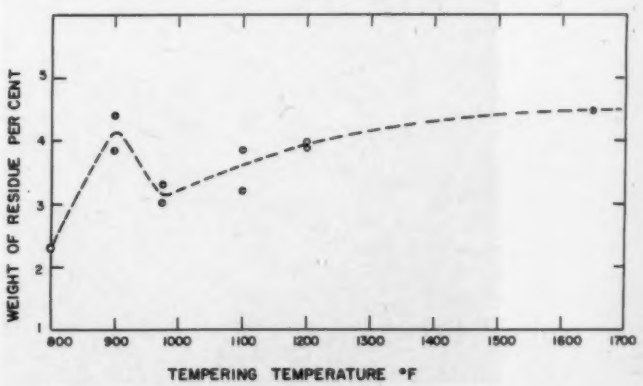


Fig. 11 — Graph showing tempering temperature vs. weight of residue (precipitated carbides and undissolved carbides). The point on the ordinate is the weight of residue in the as-quenched condition.

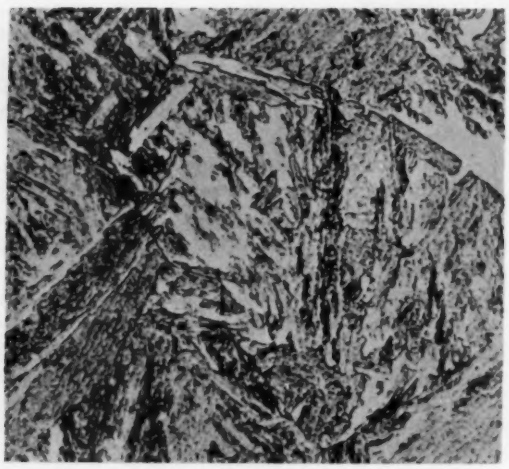


Wrought — Rc 41.0.

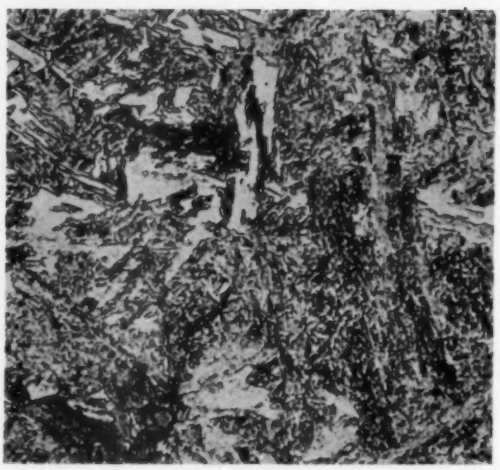


Cast — Rc 43.5.

Fig. 12 — Microstructures of cast and wrought steels from heat *A*. 2 per cent nital etch. 750 ×.

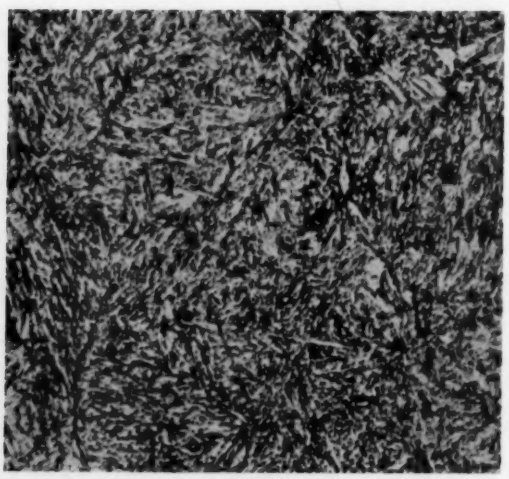


Wrought — Rc 41.0.

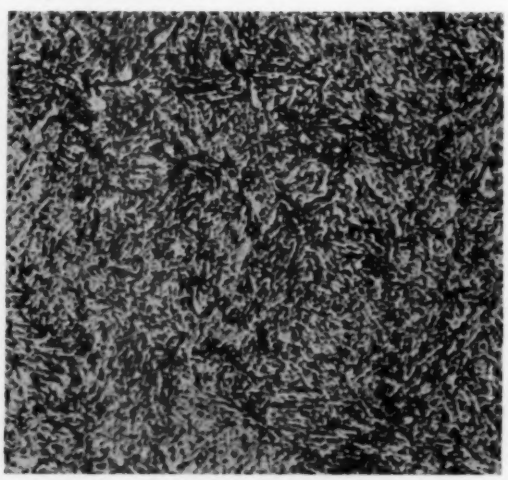


Cast — Rc 42.0.

Fig. 13 — Microstructures of cast and wrought steels from heat *F*. 2 per cent nital etch. 750 ×.

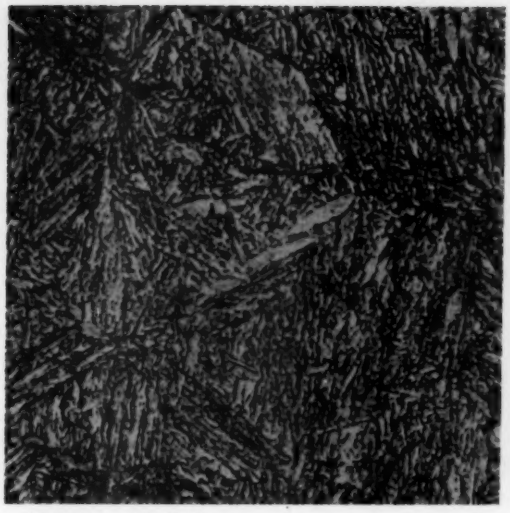


Wrought — Rc 49.5.

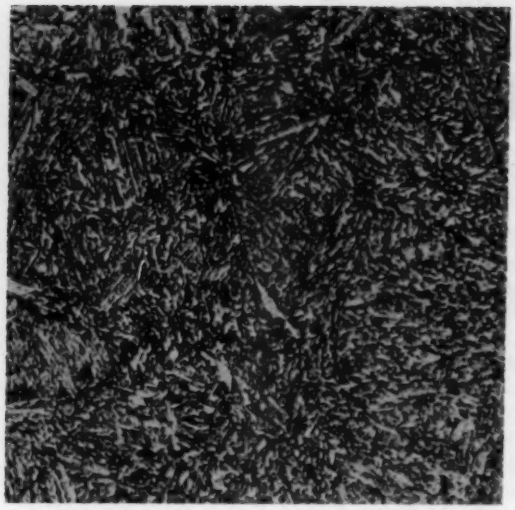


Cast — Rc 48.0.

Fig. 14 — Microstructures of cast and wrought steels from heat *H*. 2 per cent nital etch. 750 ×.



Wrought — Rc 42.5.



Cast — Rc 43.0.

Fig. 15 — Microstructures of cast and wrought steels from heat *J*. 2 per cent nital etch. 750 ×.

property becomes quite comparable at the higher temperatures for which these steels are designed.

Unnotched fatigue properties are much better for the wrought steel, but it has been shown that the wrought steels are much more notch sensitive than the cast steels. The fact that castings have no directionality properties is another favorable feature.

It has been confirmed that the secondary hardening in these steels is due to the precipitation of fine alloy carbides.

REFERENCES

1. F. B. Evans, L. J. Ebert and C. W. Briggs, "Fatigue Properties of Comparable Cast and Wrought Steels," *A.S.T.M. Proceedings, vol. 56,* pp. 979-1011, 1956.
2. Walter Crafts and John L. Lamont, "Secondary Hardening of Tempered Martensitic Alloy Steel," *A.I.M.M.E. Transactions,* Institute of Metals Div., *vol. 180,* pp. 475-512, 1949.
3. E. Smith and J. Nutting, "The Tempering of Low-Alloy Creep-resistant Steels Containing Chromium, Molybdenum, and Vanadium," *Journal of the Iron and Steel Institute, vol. 187,* pp. 314-329, Dec., 1957.
4. L. D. Tote and R. S. Zeno, "The Effect of Vanadium on the High and Low Temperature Mechanical Properties of 1 Cr-1 Mo Cast Steel," AFS TRANSACTIONS, *vol. 66,* pp. 425-432, 1958.

STICKINESS IN CORE SAND MIXTURES

Final Report
AFS Sand Division
Committee 8-K

ABSTRACT

The major problems encountered in stickiness are dealt with. A outline of a method for determining the cause and possible corrective action necessary is given. Although the Retained Sand Test is not a standard AFS test, its reliability in certain uses is noted. Precautions necessary with this test are given.

INTRODUCTION

Stickiness in core sand mixtures has long been recognized as a problem by the foundry industry, but relatively little information was available on this subject. In an effort to alleviate this situation, a Core Stickiness Committee[1] was organized in 1953 under the direction of the American Foundrymen's Society. The purpose of this committee was to define stickiness, investigate methods for its evaluation and, if possible, recommend a suitable laboratory test procedure to measure the comparative severity of stickiness.

The first progress report by Wayne H. Buell was published in the AFS TRANSACTIONS in 1954. Progress reports two and three[2] by Donald S. Mills were published in 1956.

The current committee[3] has reviewed all of the work accomplished to date, and this report is submitted as a summation and finalization of this project. It has been concluded that the only known method of determining stickiness related to core sand mixtures is the "Retained Sand Method." This test is conducted by visual inspection or actual measurement of the quantity of sand which adheres to the core box surface after a pre-determined number of cores have been produced. The actual test procedure is outlined in detail in the aforementioned progress reports.

The results of the Committee work indicate that trials conducted in the laboratory can be correlated with the results obtained similarly in production. Even though correlations were obtained, the overall results indicate that this test has limited practical value and is, therefore, not recommended as a standard AFS test. The major reasons for this conclusion are:

1) The test is subject to error mainly because of the small quantities of materials involved which multiplies any errors in the testing procedure.
2) The laboratory tests must be performed in a similar manner to the methods employed in production or the results may be unreliable. These methods would include using the same core box material, blowing the test cores, if blowing is used in production, etc.
3) There may be less time and expense involved if stickiness is determined in production and the proper approach is used to correct the condition.

If this test is to be used for the purpose of evaluation, the following precautions must be observed whether the test is performed in the laboratory or in production:

1) The test is relative and is only useful in comparing one mixture to another.
2) The mixes to be compared must be prepared in precisely the same manner in order to compare results.
3) The preparation and treatment of the core box surfaces must be standardized and carefully observed throughout the test.
4) The mechanism for producing the cores must be standardized and observed such as air pressure, drawing facilities, etc.
5) The test must be repeated a sufficient number of times to eliminate reasonable doubt of error.

NO STANDARD TEST

It was quite apparent to the committee after reviewing the qualifications for conducting this test, that it would not be acceptable as a standard. Although several other types of tests were evaluated during the course of the investigation, these, too, possessed unsatisfactory qualifications and were not acceptable. It must be noted, however, that the "Retained Sand Test," when performed in strict adherence to the procedure, is reliable, and can be used where necessary to compare the relative stickiness of materials or mixtures. It is not felt that a standard test can be subjected to the precise qualifications and procedures required and still be generally used in industry and relied upon for control work.

Although the results of the committee work did not evolve a Standard Testing Procedure, a great deal of knowledge was gained on this subject through the laboratory and production tests which were performed. In order to preserve this knowledge, it was agreed to present this paper as the culmination of the work performed. The following outline on stickiness

is presented as an aid in overcoming this problem in production. This outline is an effort to summarize all of the findings, and present them in a form which can be used by operating personnel when stickiness is encountered.

Each major core process is covered, and the items listed are known to effect stickiness in varying degrees and should be evaluated when investigating a stickiness problem. The majority of the factors listed have been investigated and proved by the results of the committee work to effect stickiness, and the possible remedies are presented as a step to overcome the specific problem. The first series of items listed must be considered in any type of core making process used, and the items listed under the three types of processes apply specifically to these processes. This outline is not presented as a "cure-all" for the problem but only as a guide in conducting a thorough evaluation.

FACTORS AND TREATMENT RELATED TO STICKINESS

A. Factors Affecting All Core Making Processes Currently In Use.

1) *Sand.* Stickiness is considered to be a surface phenomenon in any type of process. The ultimate mulling practice is to surround each grain of sand with the particular binder in use. The total surface area per unit weight of sand is, therefore, the area of each sand grain to be covered multiplied by the total number of sand grains in a particular measure.

   In considering sand as a possible cause of stickiness, it must be remembered that the surface area per unit weight of sand is dependent on the grain size, shape and distribution of the sand in use. It is, therefore, logical to assume that any changes in the type of sand used may be responsible for stickiness. Most cases of stickiness due to sand may be eliminated by reviewing the size, shape and distribution so that it is consistent and by adjusting the moisture content, if used, and the binders to fit the surface area of the sand.

2) *Mulling.* Both the mulling time and procedure must be considered relative to stickiness. During the course of establishing the proper mulling cycle, not only the optimum strength values should be considered, but also the stickiness and flowability characteristics of the mixture should receive equal attention. Both the sequence of the additions to the mix and the mulling time for each material must be considered as they will effect stickiness.

3) *Additives.* All sand additives will change the total surface area of the mix. In order to compensate for this change, the moisture content, if used, and the binder content must be adjusted to develop the full properties of the mix. The additives, themselves, must also be investigated as some types of materials in use to overcome specific core problems are in themselves inherently sticky.

4) *Core box composition.* Stickiness may be related to the type of materials used in the construction of the core box. Some materials react quite differently to different types of mixtures. Stickiness should certainly be a consideration when the choice of core box material is to be made.

5) *Equipment.* Stickiness is also related to the mechanics of the core making process. One specific example is that the sand must feed properly through the blow machine hopper and into the core box. Unless a consistent flow of sand is maintained, compressed air will be blown directly onto the core box surface. It appears that this action destroys the surface film on the box and localized sticking will occur.

6) *Variations in sand temperature.* Any changes in sand temperature could be the cause of stickiness. In most processes, cool sand is most desirable for physical properties, handling, storage and particularly stickiness. Other processes utilize hot sand to develop optimum properties. In either case, it is not the actual sand temperature that is important since the process can be adopted to the condition. Variations, however, must be avoided since the process cannot be continually changed to compensate for this variable.

B. Oil-Cereal Binders.

Problems found with this process, but because of the type of binders used, do not affect the other processes.

1) *Water – Cereal Ratio.* If sufficient water is employed to develop the optimum green and dry properties with this process, the core sand mix may be expected to have poor flowability and be relatively sticky. Because of this fact, in normal core blowing practice, the moisture content is usually and necessarily reduced to promote flowability and prevent stickiness. The cereal-water ratio in some production mixes is, therefore, such that the mixture is lacking water with respect to some green properties, but particularly to the dry strengths.

   These values may be improved, within limits by increased and proper mulling time and procedure. The general rule is to use as much water as can be permitted without causing excessive stickiness and poor flowability.

   The problem of excessive sand temperatures would also be most apparent in this section since the moisture loss due to heat will change the water-cereal ratio and promote stickiness.

2) *Cleaning and conditioning core boxes.* The objective of conditioning a core box is to build up a thin film which will provide lubricity. The conditioning should not create an excessively wet condition because this may lead to other forms of stickiness. The core box cleaning operation should be designed to remove the dirt or other foreign materials but leave the lubricating film. This process is generally in the form of a solvent material used to clean the core box or a mild abrasive.

3) *Release Agents.* The normal purpose of release agents is to provide increased flowability in the storage hoppers and the blowing machines. It is also used directly on the core box to provide

lubricity, but it must be remembered that excessive amounts used either in the mix or on the core box can be detrimental. The sequence of addition of the release agent in the mulling cycle is also important.

The normal practice is to add the release near the end of the mixing cycle for the best overall results. The actual time is not as important as is the consistency of the cycle throughout the day's production.

4) *Handling and Storage of the Mix.* With some types of binders, the oxidation and loss of solvent may influence the stickiness of the mix. The relative tendency of the mix to air dry should be considered in relation to the storage and handling systems. Location of the storage hoppers relative to heat from core ovens, shape of the hoppers influencing flowability and quantity of sand mixed ahead of production are some of the factors which must be investigated.

C. Water-Soluble Binders.

The type of binders included are the silicates, some phenolic, urea and aldehyde base resin binders, sugar based materials and certain water soluble by-products that are being used as binders. The entire class of materials is notably lacking in lubricity and, therefore, stickiness is a primary problem.

These binders require a high moisture content to develop adequate dry properties. Due to the higher moisture requirement, the mixes tend to be sticky and lack flowability. In most all cases, a finer sand with a wide grain distribution will help to relieve or control the flowability and sticking problem. The general approach is to use a release agent in the sand to achieve the necessary lubricity. It has been found with the resin type materials, particularly, that a release agent of some type is essential, but the quantity must be controlled because this type of material will effect the other properties of the mix.

D. Shell Cores and Molds.

The process also requires a release agent usually in the mixture and on the core box surface. The temperatures used must be considered in the selection of the release since the ability to withstand heat is a primary concern. Normally waxes or some types of metallic stearates are used in the sand mix, and some waxes and silicones are used on the core box surface. It is also well to note that release agents used in the core sand mix has other functions which must be considered.

In some cases the prevention of caking in the hoppers or machines is a function of the release as well as reducing shear strength while the sand is being mixed. There is no set pattern at present on the order of addition or the time cycle in adding these materials to the mix, but these factors will influence the effectiveness of the release in the mix.

SUMMARY

This report has dealt with the major problems encountered in stickiness, and has outlined a method of determining the cause and possible corrective action necessary. The "Retained Sand Test" has not been submitted as an AFS standard because of the limitations and precise qualifications for performing the test. It has been noted, however, that this test can be used to correlate production results where necessary.

There is no simple solution to stickiness because of the numerous variables involved, but this property can be controlled through proper application of the known variables which affect this condition. Any further work on this subject will be referred to the appropriate AFS Technical Committee if a new approach to a test procedure is forthcoming.

ACKNOWLEDGEMENT

During the course of the committee work, there has been considerable time and effort expended on this subject. Acknowledgement is made to the following: Harry W. Dietert Co., A. L. Graham, R. Glass and T. Daksiewicz; Aristo Corp., W. H. Buell and M. Dombrowski; Lanhoff Grain Co., J. E. Huss; National Engineering Co., T. V. Linabury; General Motors Corp., D. S. Mills, H. J. Hughes, Jr., A. L. Bryant, S. McComb.

REFERENCES

1. AFS Core Stickiness Sub-Committee (8-C-f), W. H. Buell, *Chairman*, A. L. Graham, J. E. Huss, T. V. Linabury, D. S. Mills.
2. AFS Core Stickiness Sub-Committee (8-C-f), D. S. Mills, *Chairman*, A. L. Bryant, W. H. Buell, A. L. Graham, T. V. Linabury, J. R. Young.
3. Core Stickiness Committee (8-K), J. R. Young, *Chairman*, W. H. Buell, A. L. Graham, H. J. Hughes, Jr., T. V. Linabury, D. S. Mills.

MAGNESIUM CASTING ALLOY EK31XA

By K. E. Nelson

ABSTRACT

Magnesium casting alloy EK31XA (Mg + 3.2 per cent didymium + 0.6 per cent zirconium) is rated with the commercial magnesium-rare earth metal-zirconium alloys as second in castability to the magnesium-aluminum-zinc alloys. This alloy combines good room temperature tensile properties with excellent tensile strength at elevated temperatures up to about 600 F. These properties, in addition to excellent creep resistance at elevated temperatures for times up to 10 hr, should prove attractive to rocket and missile designers.

Its long term properties (creep strength up to 1000 hr) should be of interest in applications operating up to about 450 F. EK31XA alloy can be welded by argon-arc or heli-arc methods, and requires no special procedures in applying surface protecting treatments.

INTRODUCTION

The magnesium-rare earth metal system of alloys has received considerable attention from investigators interested in both wrought and casting alloys. In the United States, Leontis pioneered these investigations on the binary alloys,[1,2] writing two papers which point out the effects of varying additions of the individual rare earth elements on their tensile and creep properties. Ternary and quaternary alloys of this system have also been evaluated. Pashak and Leontis[3] are reporting on wrought alloys while Leontis and Feisel[4] presented properties of casting alloys.

This latter paper shows the advantages of magnesium-didymium*-zirconium alloys over other combinations of rare earth metals in magnesium-zirconium alloys. A patent issued[5] covers magnesium alloys containing essentially at least 85 per cent magnesium, at least 0.4 per cent zirconium and from 0.5 to 4.0 per cent rare earth metal, of which neodymium constitutes at least 50 per cent of the rare earth metal, lanthanum plus cerium do not exceed 25 per cent together and the balance consisting of other rare earth metals.

A more recent publication by Payne and Bailey[6] presents information on alloys containing magnesium-rare earth metal-silver-zirconium, in which the rare earth metal was added as didymium. One composition has been given the English designation "MSR" or, using the A.S.T.M. system of nomenclature, QE22 alloy. This magnesium alloy, containing silver, didymium and zirconium, is claimed by the English to possess a yield strength comparable to some of the better aluminum alloys, with an ultimate strength sufficiently high in relation to yield strength to permit full advantage of these high yield values in design.

Limited information available on QE22 alloy indicates that while the yield and ultimate strength of EK31XA-T6 is lower than QE22 at room temperature, they become superior to QE22 properties as testing temperature increases.

As shown by Leontis and Feisel,[4] the optimum combination of strength and ductility in magnesium-didymium-zirconium alloys is obtained at about 3 per cent didymium. Accordingly, it was decided to evaluate further a composition containing nominally magnesium—3.2 per cent didymium and 0.6 per cent zirconium. This paper discusses castability, heat treatability, tensile and creep properties and physical properties of the alloy which is designated EK31XA-T6, according to the A.S.T.M. system of nomenclature.

The chemical composition of EK31XA, the recommended heat treatment and some of the physical properties of this alloy are listed in Tables 1 and 2. The term castability, as used in this paper, refers to the ability of an alloy to fill a sand mold satisfactorily and to form a casting relatively free of deleterious defects.[7]

A discussion of the techniques of alloy preparation and casting production, as well as testing methods, is appended to the paper.

DISCUSSION OF RESULTS

Castability

The measure of the degree of castability of EK31XA alloy is limited to the observations made during the pouring of the castings in the laboratory and production foundries, and to the quality of these castings as shown by various inspection techniques. The pouring times required to fill the molds were not significantly different for EK31XA alloy as compared to the commercial magnesium-rare earth metal-zirconium alloys.

Castability studies made on several magnesium-rare earth metal-zirconium alloys[7] indicated that "cope pitting" was a characteristic of these alloys.

*Rare earth metal consisting essentially of 85% neodymium and 15% praseodymium.

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Control of turbulence of the metal stream before it enters the mold cavity has been found necessary to minimize the cope pitting condition and to minimize the formation of oxide skins. Because the severity of the pitting in EK31XA alloy is comparable to that found in the commercial magnesium-rare earth metal-zirconium alloys, the gating used to control turbulence with the commercial magnesium-rare earth metal-zirconium alloys should prove satisfactory with EK31XA alloy.

Two types of segregation have been observed in castings of the magnesium-rare earth metal-zirconium and magnesium-thorium-zirconium alloys,[21,22,23] namely:

1) Gravity-type segregation presumed to be metallic zirconium, and usually present as small dense inclusions observed by radiography.
2) Eutectic-type segregation usually observed as a dense lined-up formation in a radiograph.

Both types of segregation have been observed in alloy EK31XA, and while some casting designs are susceptible to such segregation proper handling of the molten metal and correct gating and risering will go a long way toward its elimination and result in good quality castings.

The magnesium-rare earth metal-zirconium alloys have a relatively low microshrinkage tendency. For some reason as yet not explained, EK31XA is inferior to the ternary magnesium-rare earth metal-zirconium alloys EK30A and EK41A in its tendency toward microporosity. It might be classified as comparable to such alloys as AZ92A and AZ91C in this respect. "Swirl" type microporosity noted in ZK51A alloy[24] has been seen in alloy EK31XA. It can be expected that design limitations will be encountered with simpler castings in EK31XA alloy than with EZ33A alloy when freedom from microshrinkage is a requisite.

Draws are defined as surface shrinkage defects usually found at the junction between a massive section and a web. This condition results in a surface depression or separation in the casting wall while it is in the semi-solid state. EK31XA is comparable to the commercial magnesium-rare earth metal-zirconium alloys in draw tendency.

"Surface sinks" occur in a massive part of a casting solidifying under poor conditions of feeding.[24] The rather narrow solidification range of EK31XA results in an incidence of surface sinks comparable to HK31A alloy, higher than would be found in magnesium-aluminum-zinc alloys and somewhat worse than found using EZ33A alloy.

Shrinkage factors for EK31XA alloy are comparable to the values used for EZ33A alloy. As a rule of thumb, a shrink factor of $\frac{5}{32}$-in./ft can be used for green sand and $\frac{9}{64}$-in./ft on sand cores. As is true with all alloys, the casting design and location of gates and risers will also affect the actual shrinkage.

TABLE 1 — CHEMICAL COMPOSITION AND HEAT TREATMENT OF EK31XA ALLOY

Element	Composition Range, %	Heat Treatment
Didymium*	2.5-4.0	975±5 F for 12 hr in an atmosphere of 1.0-1.5% SO_2. Rapidly cool to room temperature. 400 F for 16 hr.
Zirconium	0.40-1.0	
Total Other Impurities (Maximum)	0.30	
Magnesium	Balance	

*Essentially 85% neodymium and 15% praseodymium.

TABLE 2 — PHYSICAL PROPERTIES OF ALLOY EK31XA-T6

Specific Gravity (68 F)	1.79
Density, lb/cu in.	0.065
Approximate Melting Range	
Liquidus	1197 ±5 F
Solidus	1040 ±5 F
Electrical Resistivity in Microhm, cm	
At 68 F	7.2
At 400 F	10.7
At 500 F	12.2
Thermal Conductivity in cal./cm²/cm/°C/sec.	
At 68 F	0.23
At 400 F	0.26
At 500 F	0.26
Coefficient of Thermal Expansion (68-212 F), /°F	0.0000145
Corrosion Rate in milligrams/cm²/day (range)	0.24-0.29*

*Limited data determined in 3% NaCl solution using a 14-day alternate immersion test.[20]

Heat Treatment

EK31XA is a precipitation hardening alloy, and requires a solution heat treatment followed by an artificial aging treatment in order to develop maximum properties. This alloy is quite sensitive to quenching rate subsequent to solution heat treatment, and requires a relatively rapid quench in order to develop optimum properties.

Various quenching media were used in obtaining cooling rates from the solution heat treat temperature of separately-cast test bars and slope castings. Figure 1 shows cooling rates obtained on separately-cast test bars, while Figs. 2 and 3 record the rates in thin and thick sections of slope castings. Tensile properties obtained with the various quenching conditions are given in Tables 3 and 4.

In order to eliminate bars containing microporosity, and to present a truer picture of the effect of section size and cooling rates on tensile properties, the *C* location[11] in the slope casting was used for thick sections, while *I* and *J* locations[11] were used for thin sections. Generally, the bar from the *C* location was free of microporosity. Some variations in elongation and tensile strength are caused by oxide inclusions in the bar.

It can be seen, by comparing the tensile properties in Tables 3 and 4, that the thinner wall cast test bars can be given an adequate quench more readily than the thicker sectioned parts such as the slope castings. A cooling rate from the solution heat treat temperature to 400 F of 1 F/sec is adequate for acceptable properties.

Creep Limits

Creep limits of 100 hr were obtained on bars sectioned from thick sections of slope castings which had been quenched at different rates. Results in Table 5 show the sensitivity of creep strength to cooling rate.

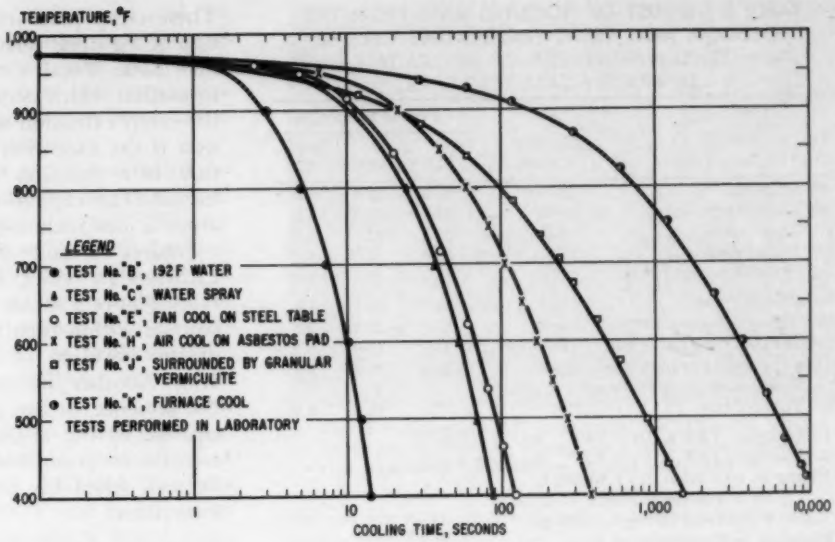


Fig. 1 — Quenching media effect on the cooling rate of EK31XA separately cast test bars.

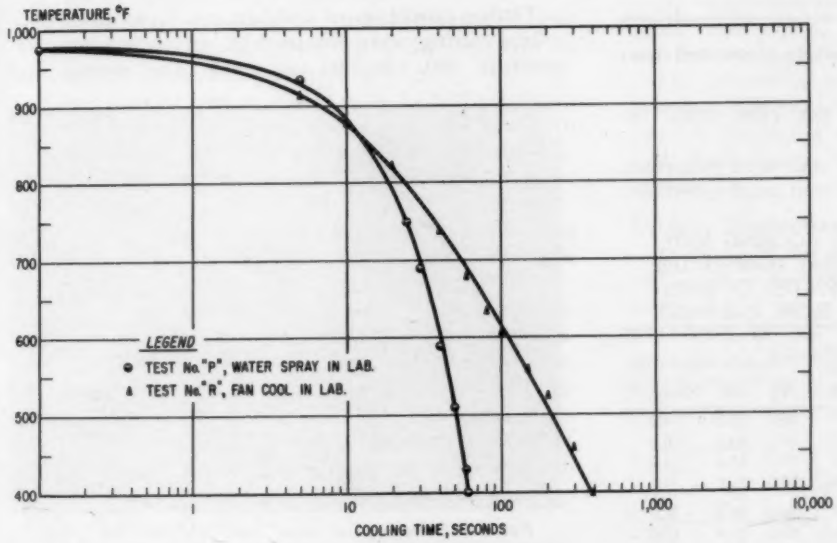


Fig. 2 — Quenching media effect on the cooling rate of EK31XA slope castings, thin section.

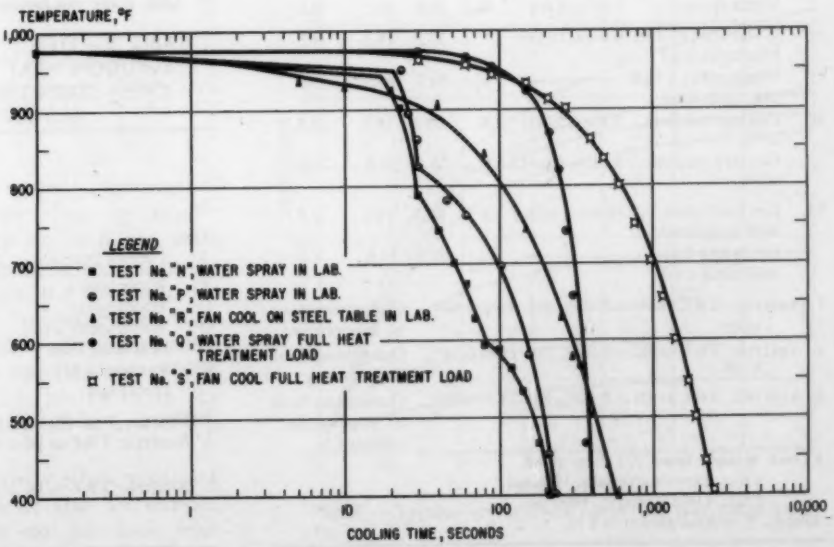


Fig. 3 — Quenching media effect on the cooling rate of EK31XA castings, thick sections.

TABLE 3 — EFFECT OF COOLING RATE FROM THE SOLUTION HEAT TREAT TEMPERATURE ON THE TENSILE PROPERTIES OF EK31XA-T6 SEPARATELY-CAST TEST BARS

Test No.	Type of Quench	Cooling Rate*	Tensile Properties TS	YS	Elong., %
A[1]	Ice water	<6 sec	36.4	22.6	6.0
B[1]	192°F water	14 sec	33.2	21.8	4.0
C[1]	Water spray	1 min, 25 sec	35.5	23.7	3.5
D[2]	Water spray	2 min, 39 sec	33.6	20.3	4.5
E[1]	Fan cool on steel table	2 min	35.4	22.4	4.0
F[2]	400 F oil bath	4 min, 40 sec	34.1	21.1	5.0
G[1]	Air cool on steel table	5 min	37.0	23.4	5.5
H[1]	Air cool on asbestos pad	6 min, 15 sec	33.5	21.9	3.5
J[1]	Surrounded by granular vermiculite	25 min	31.2	20.2	5.0
K[1]	Furnace cool	3 hr	26.2	15.3	6.0

1. Analysis: TRE as Di = 3.49%, Sol. Zr = 0.57%.
2. Analysis: TRE as Di = 3.35%, Sol. Zr = 0.54%.
*Time to cool from 975 F to 400 F.
TS = Tensile strength, 1000 psi.
YS = Yield strength, 1000 psi.
Elong., % = Elongation in 2 in.
Quenching done in laboratory.

The three conventional creep limits presented may be defined as follows:

1) Limiting stress to give 0.1 per cent creep extension.
2) Limiting stress to give 0.2 per cent total extension.
3) Limiting stress to give 0.5 per cent total extension.

TABLE 4 — SECTION SIZE AND COOLING RATE FROM THE SOLUTION HEAT TREAT TEMPERATURE EFFECT ON THE TENSILE PROPERTIES OF BARS SECTIONED FROM EK31XA-T6 SLOPE CASTINGS

Test No.	Type of Quench	Cooling Rate*	Section Size, in.	Tensile Properties TS	YS	Elong., %
L[1]	74 F Water	<30 sec	3	36.7	20.0	8.0
	74 F Water	<30 sec	5/16	36.9	20.5	6.0
M[1]	170 F water	<30 sec	3	33.2	19.8	4.0
	170 F water	<30 sec	5/16	38.6	21.2	9.5
N[1]	Water spray	3 min, 15 sec	3	36.2	20.1	6.0
	Water spray	————	5/16	40.9	21.8	12.0
P[2]	Water spray	3 min, 27 sec	3	31.2	19.2	4.7
	Water spray	1 min, 2 sec	5/16	36.4	20.7	6.0
Q[3]	Water spray a full heat treat load	6 min, 15 sec	3	34.8	19.5	4.5
	Water spray a full heat treat load	————	1/4	34.7	20.5	3.0
R[2]	Fan cool on steel table	9 min, 35 sec	3	29.4	18.7	3.3
	Fan cool on steel table	6 min, 40 sec	5/16	33.4	20.5	4.0
S[3]	Fan cool a full heat treat load	40 min, 30 sec	3	26.5	14.5	4.0
	Fan cool a full heat treat load	————	1/4	29.0	16.6	4.0

1. Analysis: TRE as Di = 3.40%, Sol. Zr = 0.53%. Quenching done in laboratory.
2. Analysis: TRE as Di = 3.35%, Sol. Zr = 0.54%. Quenching done in laboratory.
3. Analysis: TRE as Di = 3.16%, Sol. Zr = 0.52%. Quenching done in production foundry.

*Time to cool from 975 F to 400 F.
TS = Tensile strength, 1000 psi.
YS = Yield strength, 1000 psi.
Elong., % = Elongation in 2 in.

These creep parameters were obtained by interpolation of log-stress versus log-extension plots of the original data. Total extension represents the total deformation which occurs during both the loading and the creep extension testing period, while creep extension is the extension at constant stress and temperature is a function of time after the specimen is loaded. The expansion of the sample due to temperature is not included in any of these measurements.

Figures 4 and 5 show the microstructure of EK31XA-T6 which has been quenched at different rates. Figure 5 shows that there is a rather coarse precipitate within each grain which is associated with a drastic lowering of the strength properties of the alloy. Another indication of inadequate quenching is the presence, at the grain boundaries, of rods of precipitate having appreciable width rather than thin, continuous grain boundaries. Similar structures can be seen when the alloy is exposed above 500 F for long times.

Mechanical Properties

Tensile properties of test bars machined from individual castings were obtained by testing at room temperature, 300, 400, 500 and 600 F. The average and

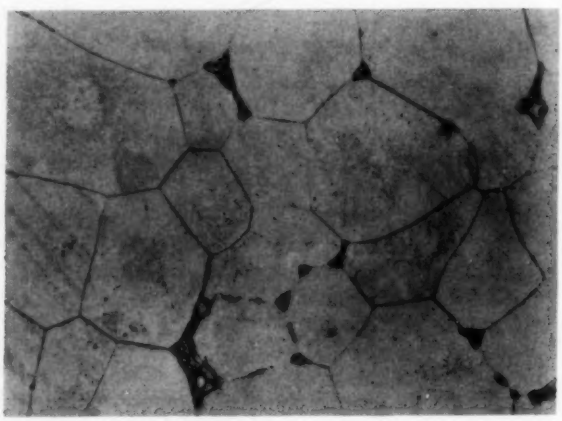


Fig. 4 — Structure of sand cast EK31XA-T6 alloy which has received a rapid quench from the solution heat treat temperature. Glycol etch. 250 ×.

TABLE 5 — EFFECT OF COOLING RATE FROM THE SOLUTION HEAT TREAT TEMPERATURE ON THE CREEP STRENGTH OF BARS SECTIONED FROM EK31XA-T6 SLOPE CASTINGS

Test No.	Type of Quench	Cooling Rate*	100-Hr Creep Limits, 1000 psi 0.1% Creep Extension	0.2% Total Extension	0.5% Total Extension
T[1]	Water spray	3 min, 39 sec	9.7	8.0	13.0
U[1]	Water spray	4 min, 4 sec	10.7	9.0	13.3
Q[2]	Water spray a full heat treat load	6 min, 15 sec	10.5	8.7	13.9
S[2]	Fan cool a full heat treat load	40 min, 30 sec	7.5	7.0	9.2

*Time for 3-in. section to cool from 975 F to 400 F.
1. Analysis: TRE as Di = 3.16%, Sol. Zr = 0.52%. Quenching done in laboratory.
2. Analysis: TRE as Di = 3.16%, Sol. Zr = 0.52%. Quenching done in production foundry.

range of values at each testing temperature and the melt analysis from which the castings were poured are shown in Table 6. These castings had received an adequate quench from the solution heat treat temperature. Some of the lower properties were influenced by oxide skins and microporosity.

The average tensile properties and the standard deviation values,[7] as well as the number of values used for each determination, are listed in Table 7. The results of tension tests in Tables 7 and 8 show that bars sectioned from castings are roughly comparable in strength to separately-cast bars. A comparison of average tensile properties of separately-cast test bars of various magnesium alloys is made in Table 9.

Tension stress-strain measurements taken on EK31XA-T6 test bars at room and at elevated temperatures, obtained at a testing speed of 0.005 in./in./min, are shown in Fig. 6. There is little effect of strain rate on the strength of EK31XA-T6 alloy through 400 F, as shown in Figs. 7 and 8. Effects of strain rate are important in applications where loads are rapidly applied to the structure.

The effect of exposure for various lengths of time at elevated temperatures on the tensile strength of this alloy is shown in Figs. 9 and 10.

Creep tests on bars sectioned from EK31XA-T6 castings were made for times up to 1000 hr. Table 10 presents 100-hr and 1000-hr creep limits of EK31XA-T6 alloy, with a comparison of creep limits of other magnesium casting alloys being made in Table 11. Effects of exposure at elevated temperatures prior to standard creep testing at 400 F and 500 F are small, as shown in Fig. 11.

In order to provide the designer with some indication of allowable stress, and resulting deformation at a given temperature for short-time applications, isochronous stress-strain curves for EK31XA-T6 alloy are shown in Figs. 12-17. As explained in an earlier publication,[19] these figures are prepared from short-time creep data. The short-time creep strength of EK31XA-T6 is compared with other magnesium alloys in Figs. 18 and 19.

The effect of temperature on the modulus of elasticity of EK31XA-T6 alloy is reported in Table 12.

Laboratory-produced test castings were used to perform bearing and shear tests at 75, 400 and 600 F, as well as hardness measurements at room temperature. Results are shown in Table 13.

Welding and Straightening

Limited experience on the repair welding of EK31XA alloy castings indicates they are readily weldable by argon-arc and heli-arc methods using EZ33A welding rod. Castings should be welded in the T4 (solution heat treated) or in the T6 (solution heat treated plus aged) temper. In unrestrained sections, preheating is required. Thin-wall or restrained sections may require a short preheat at 400 F. Whether the casting is welded in the T4 or in the T6 temper, a post-weld treatment of 16 hr at 400 F is mandatory.

Straightening of EK31XA castings when required, should also be done in the T4 or the T6 temper. Correction of minor deflections can be done with

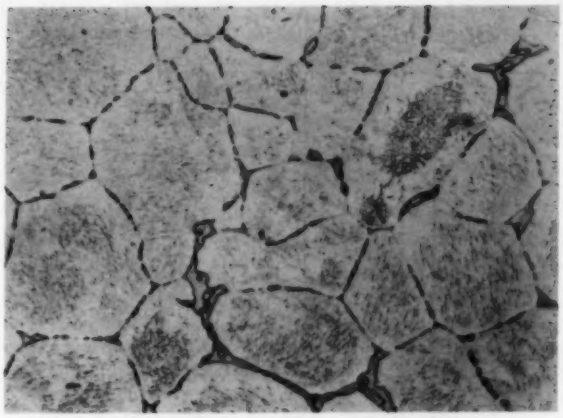


Fig. 5 — Structure of sand cast EK31XA-T6 alloy which has received an inadequate quench from the solution heat treat temperature. Glycol etch. 250 ×.

the casting at room temperature, while major straightening should be carried out at 400 F. All straightening operations should be followed with a regular aging treatment of 16 hr at 400 F. Stress relaxation tests have not been made with this alloy.

The sensitivity of the properties of this alloy to exposure at elevated temperatures requires that close controls be placed over exposure time and temperature, both during welding and during straightening operations.

Surface Protection

As shown by the corrosion rate reported in Table 2, resistance of EK31XA alloy to saline attack is equivalent to the better magnesium alloys. Applications considered in this composition should be carefully reviewed to determine the optimum protective system that can be applied. This alloy is similar to the commercial magnesium-rare earth metal-zirconium al-

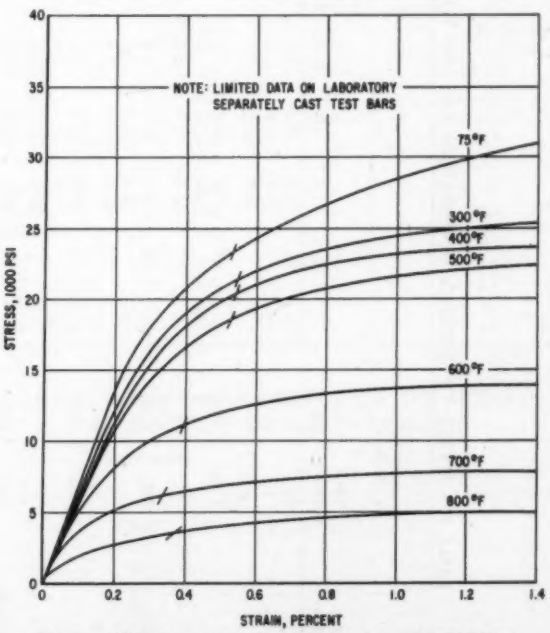


Fig. 6 — Tension stress-strain curves for cast EK31XA-T6.

TABLE 6 — EFFECT OF TESTING TEMPERATURE ON THE TENSILE PROPERTIES OF BARS SECTIONED FROM INDIVIDUAL CASTINGS — EK31XA-T6 ALLOY

				Testing Temperature, F																			
				75				300				400				500				600			
Casting Number	TRE as Di, %	Sol. Zr, %		No. of Bars	TS	YS	E, %	No. of Bars	TS	YS	E, %	No. of Bars	TS	YS	E, %	No. of Bars	TS	YS	E, %	No. of Bars	TS	YS	E, %
A	3.04	0.58	Ave.	25	30.5	20.0	4.0					16	24.7	17.7	11.8	16	19.2	15.1	22.8	16	13.2	9.6	57.1
			range		25.4	17.6	2.5						21.9	15.9	6.0		17.4	13.9	13.0		11.3	8.3	28.0
					−36.3	−21.7	−7.0						−27.0	−18.5	−22.0		−21.3	−16.6	−38.0		−14.9	−10.8	−80.0
B	3.23	0.56	Ave.	10	32.2	20.4	4.2					9	27.3	19.0	15.9	8	20.2	16.2	36.6	7	14.1	10.9	53.7
			range		29.4	17.6	2.5						26.1	18.3	7.0		19.1	14.7	23.0		13.0	9.9	22.0
					−36.6	−21.4	−6.0						−28.2	−19.8	−20.0		−21.1	−17.1	−43.0		−14.6	−11.7	−92.0
C	3.23	0.56	Ave.	22	33.0	21.0	4.2					6	25.2	17.8	11.8	6	19.5	15.0	25.3	6	14.0	10.2	55.5
			range		26.0	19.6	2.0						22.2	15.8	7.0		19.0	13.8	12.0		13.3	8.7	27.0
					−35.9	−22.9	−6.0						−27.4	−20.0	−17.0		−20.0	−16.2	−36.0		−15.5	−12.2	−68.0
D	3.23	0.56	Ave.	8	29.9	20.1	2.8					8	25.3	17.1	14.4	8	20.0	15.2	27.0	8	13.2	10.3	72.6
			range		27.7	19.4	2.0						24.5	16.3	10.0		18.6	14.2	23.3		12.5	9.4	58.0
					−32.4	−21.4	−4.0						−26.0	−17.7	−19.0		−21.1	−15.9	−33.0		−14.4	−12.4	−108.0
Slope Casting "L"	3.40	0.53	Ave.	9	35.7	20.2	6.0																
			range		32.7	17.9	4.7																
					−36.9	−20.9	−8.0																
Slope Casting "M"	3.40	0.53	Ave.	9	34.5	19.9	5.3																
			range		31.3	17.8	4.0																
					−38.6	−21.2	−9.5																
Slope Casting "N"	3.40	0.53	Ave.	9	36.1	20.5	7.1																
			range		30.4	18.4	3.3																
					−40.9	−21.8	−12.0																
Slope Casting "P"	3.35	0.54	Ave.	9	32.9	20.0	5.0																
			range		29.4	19.1	4.0																
					−36.4	−20.9	−6.5																
Slope Casting "Q"	3.16	0.52	Ave.	10	34.7	20.5	4.5	4	29.3	19.0	11.2	4	29.1	19.6	13.8	4	25.4	17.8	24.2	4	16.2	12.6	41.0
			range		28.9	19.5	1.5		28.6	18.3	10.0		28.2	18.6	8.0		24.3	17.2	20.0		14.9	12.2	26.0
					−39.0	−22.1	−7.5		−30.0	−19.6	−13.0		−30.6	−20.6	−18.0		−26.3	−18.3	−30.0		−18.5	−12.9	−55.0
Slope Casting "R"	3.35	0.54	Ave.	9	30.3	19.3	3.8																
			range		27.3	18.6	2.7																
					−33.4	−20.5	−4.7																

Slope casting designations refer to the type of quench described in Table 5.
TS = Tensile strength, 1000 psi.
YS = Yield strength, 1000 psi.
E, % = Elongation in 1 or 2 in.

The average values were obtained arithmetically.
Bars held 10 min at temperature before testing.

TABLE 7 — AVE. TENSILE PROPERTIES AND THE STANDARD DEVIATION OF BARS SECTIONED FROM EK31XA-T6 CASTINGS

Property*	Testing Temp., F	No. of Tests	Ave.	Standard Deviation
TS	75	127	32.6	3.2
YS	75	127	20.2	1.0
E, %	75	127	4.6	1.5
TS	300	4	29.3	—
YS	300	4	19.0	—
E, %	300	4	11.2	—
TS	400	47	26.1	1.9
YS	400	47	18.2	1.2
E, %	400	47	13.8	5.3
TS	500	42	20.2	—
YS	500	42	15.5	1.3
E, %	500	42	26.7	10.1
TS	600	41	13.8	1.1
YS	600	41	10.4	1.2
E, %	600	41	58.0	19.0

* TS = Tensile strength, 1000 psi.
YS = Yield strength, 1000 psi.
E, % = Elongation in 1 or 2 in.

Note: Where no standard deviation is listed, the average value for that property is the arithmetic average of the individual values.
Bars held 10 minutes at temperature before testing.

TABLE 9 — COMPARISON OF AVERAGE TENSILE PROPERTIES OF SEPARATELY-CAST TEST BARS OF VARIOUS MAGNESIUM ALLOYS

Composition	Testing Temp., F	Properties* TS	YS	E, %
AZ92A-T6	75	40	23	2
	200	37	21	25
	300	28	17	35
	400	17	12	36
	500	11	8	33
ZH62A*-T5	75	40	27	8
	200	33	23	20
	300	26	20	24
	400	19	15	28
	500	14	10	30
EZ33A-T5	75	23	16	3
	200	23	15	5
	300	22	14	10
	400	21	12	20
	500	18	10	31
	600	12	8	50
HK31A-T6	75	31	16	6
	200	29	16	8
	300	27	15	12
	400	24	14	17
	500	23	13	19
	600	20	12	22
HZ32A-T5	75	29	15	6
	200	26	14	15
	300	22	12	23
	400	17	10	33
	500	14	9	39
	600	12	8	38
EK31XA-T6	75	35	22	4.5
	400	29	20	17
	500	21	17	28
	600	15	12	49

TS = Tensile strength, 1000 psi.
YS = Yield strength, 1000 psi.
E, % = Elongation in 2 in.
*Bars held 10 min at temperature before testing.

TABLE 8 — TENSILE PROPERTIES OF SEPARATELY-CAST TEST BARS POURED AND HEAT TREATED WITH CASTINGS EK31XA-T6 ALLOY

Testing Temp., F	No. of Bars	TS Range	TS Ave.	YS Range	YS Ave.	E, % Range	E, % Ave.
75	46	31.3-38.0	35.1	18.5-24.2	21.8	2.0-7.0	4.5
400	9	22.5-32.1	28.6	16.5-22.7	20.3	13.0-22.0	17.2
500	7	18.7-22.8	20.6	14.9-19.4	17.4	22.5-37.5	28.4
600	8	11.8-18.6	15.2	8.6-16.2	12.5	40.0-73.0	48.9

TS = Tensile strength, 1000 psi.
YS = Yield strength, 1000 psi.
E, % = Elongation in 2 in.
Bars held 10 min at temperature before testing.

TABLE 10 — CREEP RESISTANCE OF EK31XA-T6 BARS SECTIONED FROM CASTINGS*

Testing Temp. F	No. of Test Bars	0.1% Creep Extension	0.2% Total Extension	0.5% Total Extension
Range of Creep Limit Values, 1000 psi., 100-hr				
400	24	8.1-11.7	6.4-9.4	10.0-13.1
500	14	1.9- 3.6	2.3-3.8	3.1- 4.4
600	10	0.6- 1.3	0.8-1.5	1.2- 1.8
Range of Creep Limit Values, 1000 psi., 1000-hr				
400	11	3.9-6.7	4.4-6.5	6.8-8.6
500	11	0.8-1.6	1.2-2.2	1.9-2.9
600	8	0.3-0.7	0.5-0.9	0.7-1.1

*Bars from laboratory test casting and from castings made in the production foundry.

TABLE 11 — COMPARISON OF RANGES OF 0.2% TOTAL EXTENSION CREEP LIMITS IN 100 HR OF BARS SECTIONED FROM VARIOUS MAGNESIUM ALLOY CASTINGS

Composition	Testing Temp., F	Number of Test Bars	Range Stresses Required to Produce 0.2% Total Extension in 100 hr, 1000 psi.
EZ33A-T5	400	41	5.8- 9.0
HK31A-T6	400	24	7.8-10.6
HZ32A-T5	400	31	6.5- 9.4
EK31XA-T6	400	24	6.4- 9.4
HK31A-T6	500	9	5.4- 7.0
HZ32A-T5	500	30	5.4- 7.5
EK31XA-T6	500	14	2.3- 3.8
EZ33A-T5	600	33	0.9- 1.6
HK31A-T6	600	25	2.1- 3.6
HZ32A-T5	600	42	2.8- 4.9
EK31XA-T6	600	10	0.8- 1.5

TABLE 12 — TEMPERATURE EFFECT ON THE MODULUS OF ELASTICITY OF EK31XA-T6 ALLOY

Temperature, F	Modulus* (psi x 10^6)
75	6.4
300	6.2
405	5.9
505	5.8
600	5.3
615	5.2
700	4.6

*Ave. of three tests.

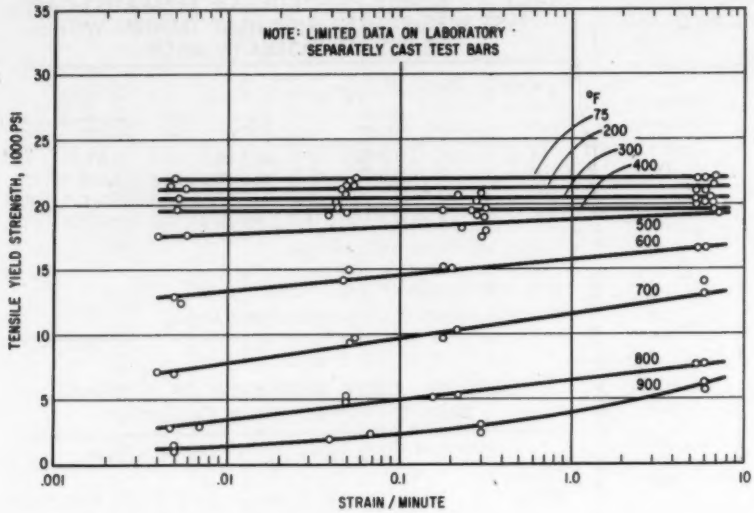


Fig. 7 — Strain rate effect on yield strength of sand cast EK31XA-T6 alloy.

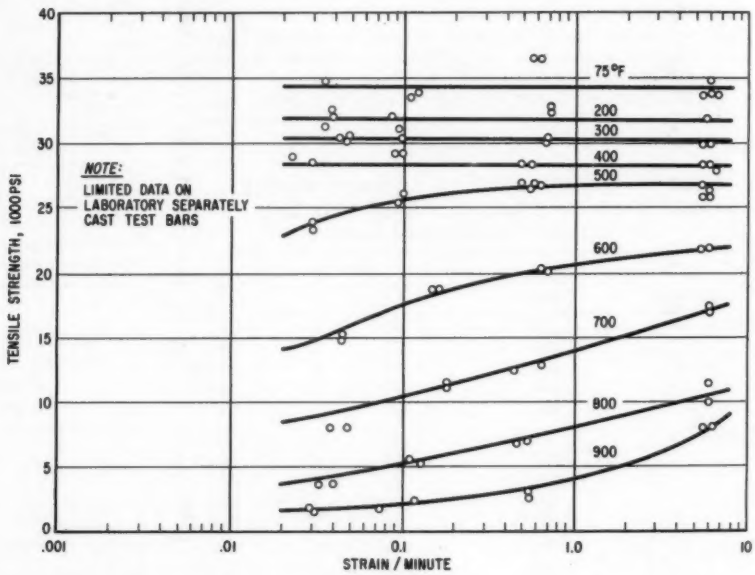


Fig. 8 — Strain rate effect on tensile strength of sand cast EK31XA-T6 alloy.

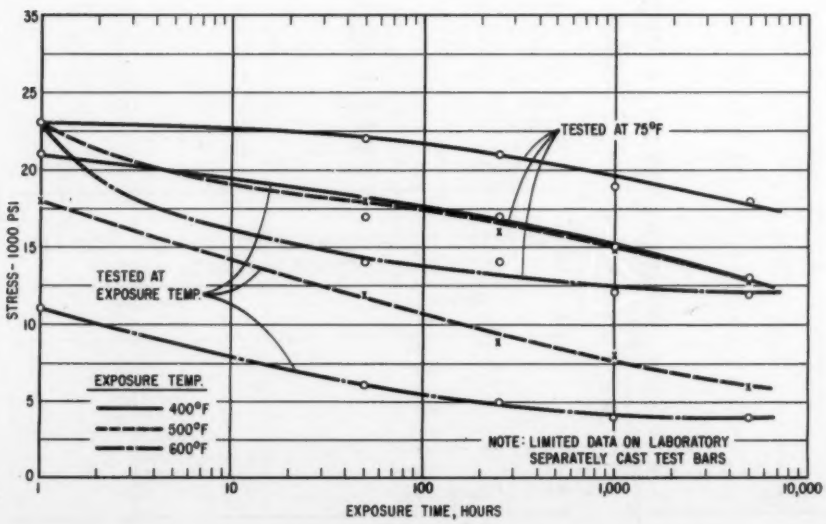


Fig. 9 — Exposure effect on yield strength of EK31XA-T6 alloy at 75, 400, 500 and 600 F.

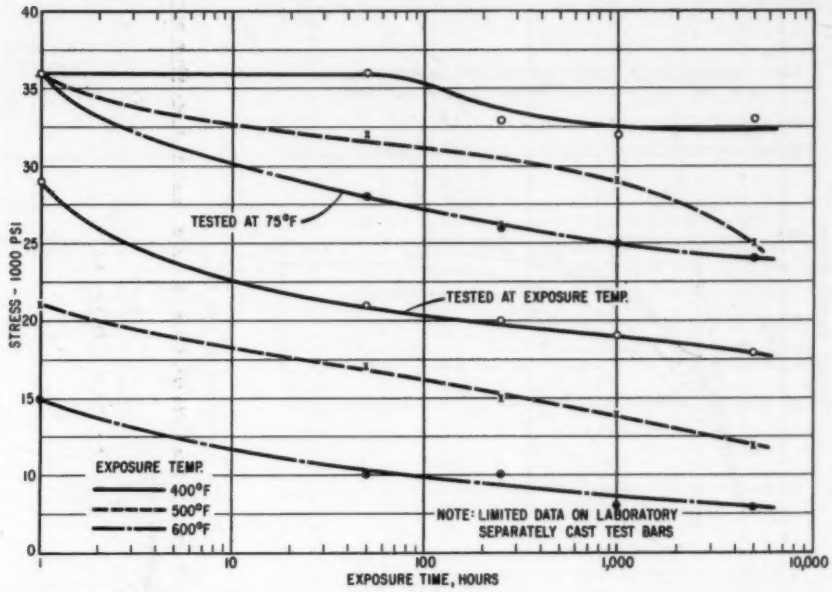


Fig. 10 — Exposure effect on tensile strength of EK31XA-T6 alloy at 75, 400, 500 and 600 F.

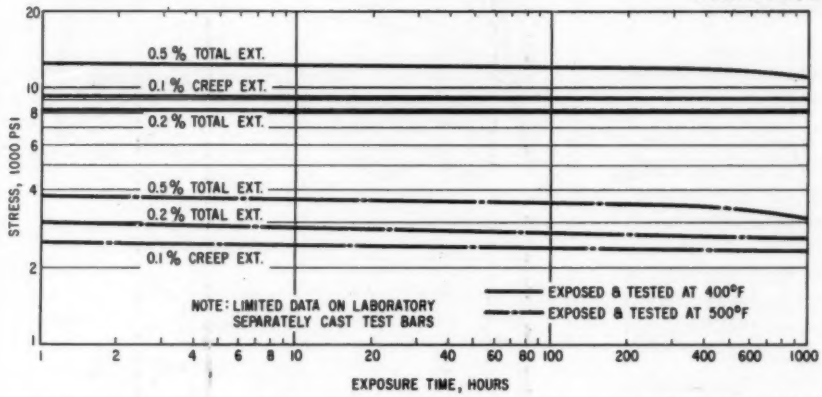


Fig. 11 — Exposure at elevated temperatures effect on 100 hr creep limits of EK31XA-T6 alloy.

loys in response to chemical treatment. The following treatments can be satisfactorily applied:

1) Chrome pickle (MIL-M-3171 Type I).
2) Galvanic anodize (MIL-M-3171 Type IV).
3) Sealed chrome pickle (MIL-M-3171 Type II).
4) Anodize (AMS 2478).
5) Anodize [(MIL-C-13335, (ORD) and AMS 2476)].

Dichromate treatment (MIL-M-3171 Type III) cannot be used on this alloy.

CONCLUSIONS

EK31XA alloy is classified in terms of castability with the magnesium-rare earth metal-zirconium alloys, even though limitations in design will be met with simpler castings in this alloy than with EZ33A alloy.

The excellent tensile properties at room and at elevated temperatures, as well as the good creep strength exhibited for short times (up to 10 hr), suggest EK31XA-T6 alloy will find applications in rockets and missiles. Its creep strength for times up to 1000 hr is favorable for applications operating up to about 450 F.

Standard procedures recommended for the chemical treatment of commercial sand-cast magnesium alloys can be used with EK31XA-T6 alloy. When properly chemically treated and painted, castings in this alloy should perform satisfactorily in service.

While this paper relates specifically to sand-cast EK31XA alloy, permanent mold castings should be readily prepared with properties equivalent to those reported herein.

A recommended composition range and heat treat-

TABLE 13 — BEARING, SHEAR AND HARDNESS VALUES FOR EK31XA-T6 SAND CASTING ALLOY

Testing Temp., F	Bearing Strength, 1000 psi* (3/16 in. Dia. Pin)	
	Bearing Yield	Bearing Ultimate
75	48	68
400	43	57
600	26	37
	Shear Strength, 1000 psi** (1/8 in. Dia. Pin)	
75	23	
400	19	
600	11	
	Rockwell "E"	Brinell (500 kg)
Hardness	74	63.8

*2d edge distance — 8d width.
**Double pin shear.
NOTE: These are limited laboratory data.

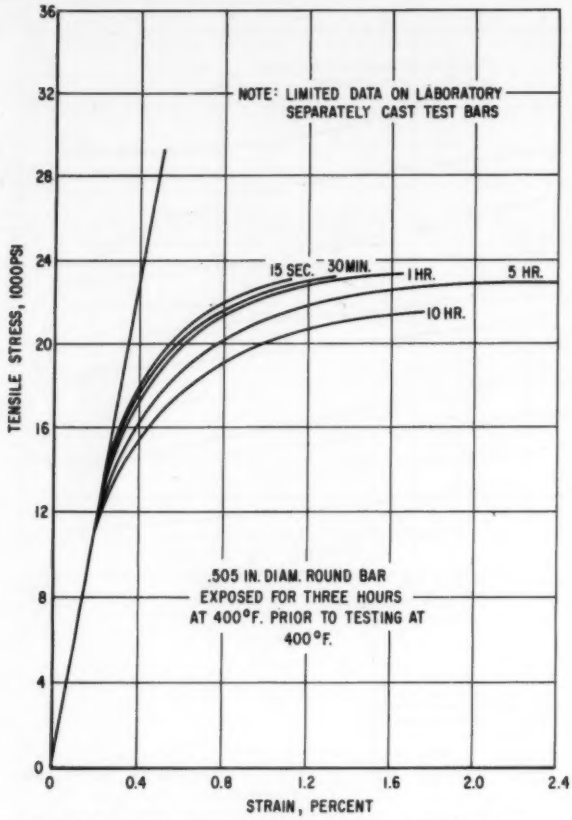


Fig. 12 — Isochronous stress-strain curves at 400 F for cast magnesium alloy EK31XA-T6.

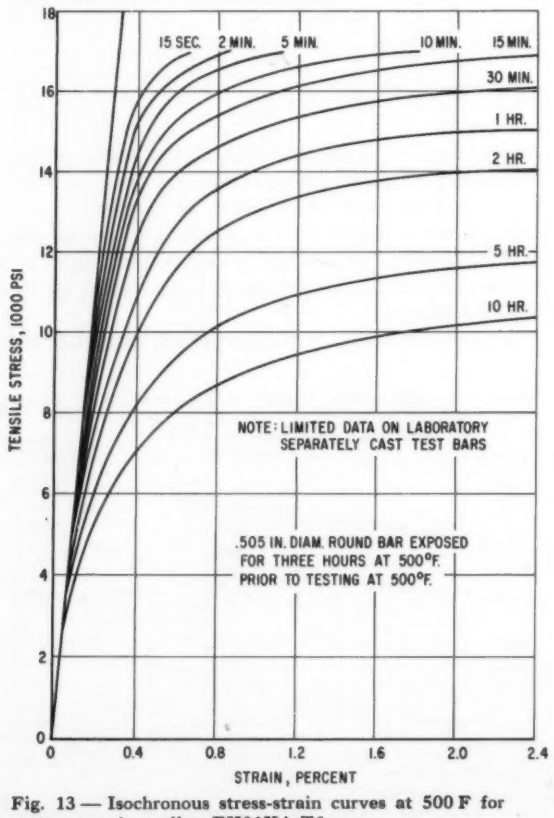


Fig. 13 — Isochronous stress-strain curves at 500 F for cast magnesium alloy EK31XA-T6.

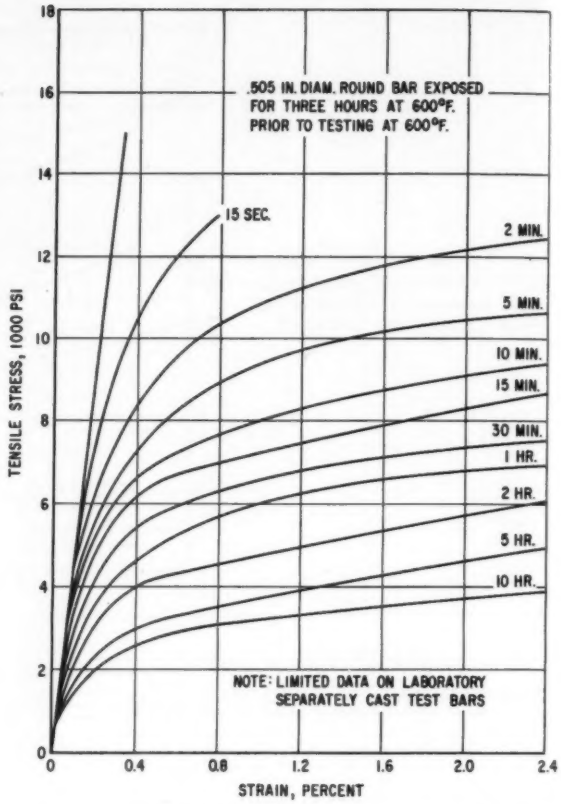


Fig. 14 — Isochronous stress-strain curves at 600 F for cast magnesium alloy EK31XA-T6.

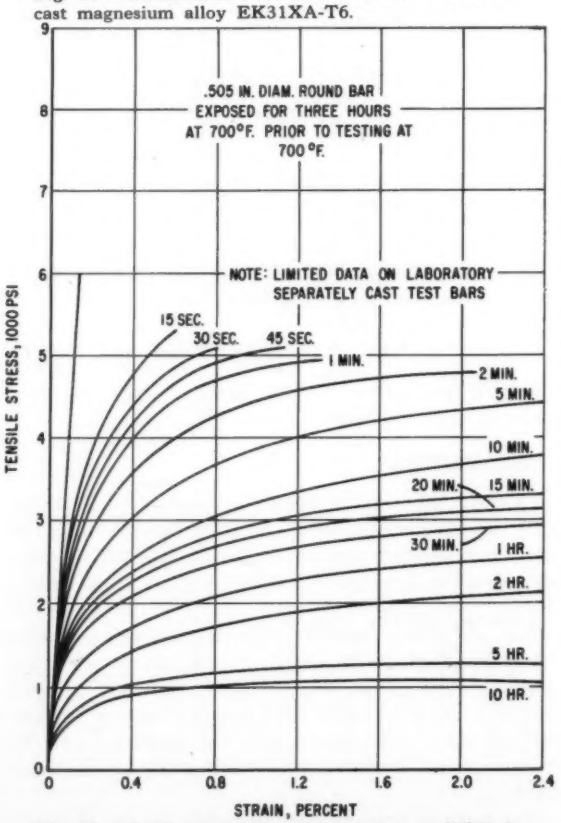


Fig. 15 — Isochronous stress-strain curves at 700 F for cast magnesium alloy EK31XA-T6.

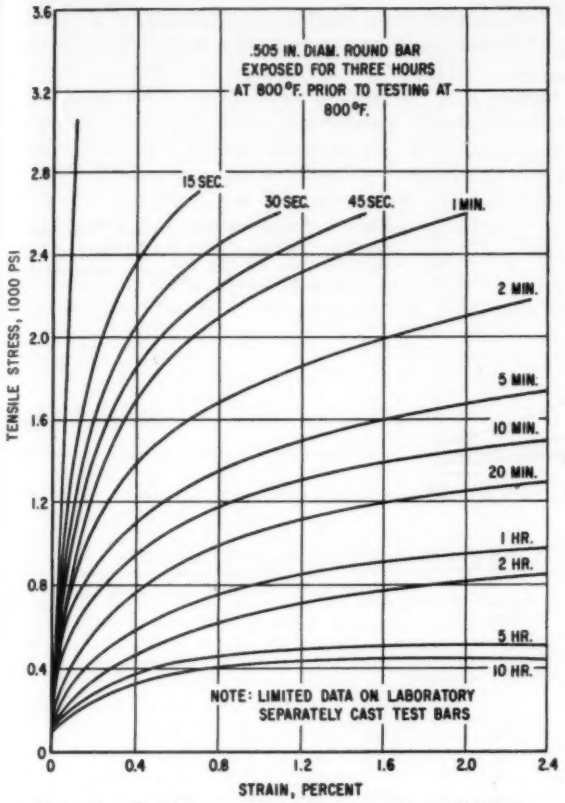


Fig. 16 — Isochronous stress-strain curves at 800 F for cast magnesium alloy EK31XA-T6.

ment is given for EK31XA-T6 alloy in Table 1. Table 14 reports typical and guaranteed tensile properties of this alloy, and compares them with properties of other magnesium casting alloys.

APPENDIX A

PROCEDURE

Preparation of the molten alloy is similar to that used for the standard magensium casting alloys con-

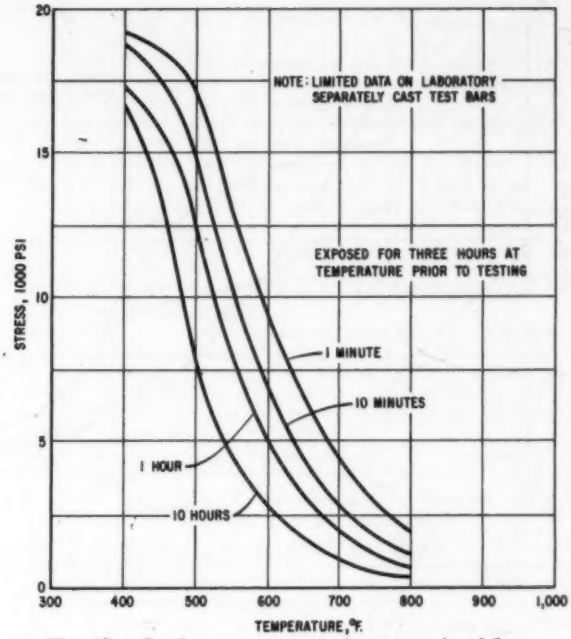


Fig. 17 — Isochronous stress-strain curves for 0.5 per cent total extension for cast magnesium alloy EK31XA-T6.

taining rare earth metals and thorium.[7,8,9] In these alloys, as in EK31XA alloy, serious losses of zirconium may be encountered if precautions are not taken to prevent contamination with harmful impurities such as aluminum and silicon. It can also be pointed out that contamination with cerium and lanthanum, such as found in mischmetal used in the commercial magnesium-rare earth metal alloys, could result in a lowering of properties.

Melts were prepared using either virgin materials or foundry run-around scrap with additions of didymium and zirconium to make up losses of these elements incurred during remelting. Didymium was added as a magnesium-didymium hardener analyzing between 10 and 15 per cent didymium. The total rare earth content as didymium and the zirconium con-

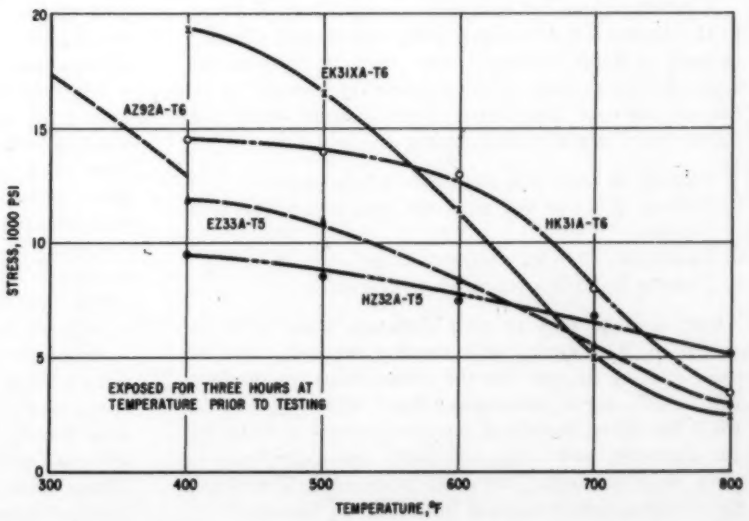


Fig. 18 — Comparison of various magnesium casting alloys isochronous stress-strain curves for 0.5 per cent total extension in a 15 sec test.

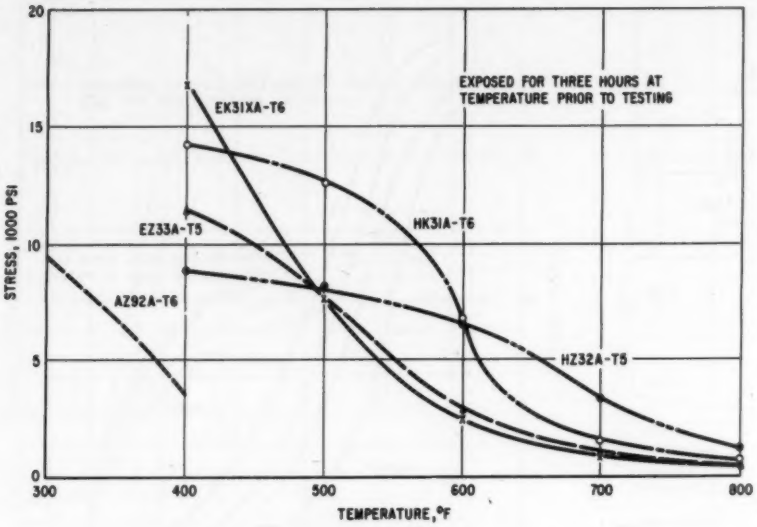


Fig. 19 — Comparison of various magnesium casting alloys isochronous stress-strain curves for 0.5 per cent total extension in a 10 hr test.

TABLE 14 — MECHANICAL PROPERTIES OF VARIOUS MAGNESIUM SAND CASTING ALLOYS

	Separately-Cast Test Bars										Bars Sectioned from Castings					
	TS		YS		E, %		Shear, 1000 psi.	Typical Bearing Strength 1000 psi		Brinell Hardness	TS		YS		E, %	
Alloy	Typ.	Min*	Typ.	Min*	Typ.	Min*	Typ.	Ult.	Yd.	No.	Ave.**	Min*	Ave.**	Min*	Ave.**	Min*
AZ92A-T6	40.0	34.0	21.0	18.0	2	Not req.	20	80	65	84	25.5	17.0	16.0	13.5	—	—
ZH62A-T5	40.0	35.0	25.0	22.0	6	4	23	72	49	70	31.5	26.5	17.5	15.5	1	—
EZ33A-T5	23.0	20.0	15.0	14.0	3	2	20	57	40	50	15.0	13.0	12.5	11.0	0.5	—
HK31A-T6	32.0	27.0	15.0	13.0	8	4	21	61	40	55	23.0	19.0	11.7	10.5	1	—
HZ32A-T5	30.0	27.0	14.0	13.0	7	4	20	60	37	57	23.0	19.0	11.7	10.5	1	—
EK31XA-T6	35.0	31.0	22.0	19.0	4.5	2.5	23	68	48	64	26.0	22.5	17.5	16.5	1.0	—

TS = Tensile strength, 1000 psi.
YS = Yield strength, 1000 psi.
E, % = Elongation in 2 in.

*Guaranteed properties based upon one test.
**Guaranteed properties based upon four to ten tests.

NOTE: Guaranteed properties on EK31XA-T6 are tentative.

tent were analyzed by wet chemical methods,[10] with the zirconium contents being reported as "soluble," indicating that portion soluble in dilute acid solution. The usual impurities in each melt were determined spectroscopically.[10]

A casting intended to evaluate sensitivity of alloys to the formation of surface sinks, draws and porosity as well as slope castings[11] were used in conjunction with castings poured in the production foundry in order to evaluate castability. Four castings were prepared in the production foundry:

1. Casting A, 86-lb housing used on helicopters.
2. Casting B, 14-lb airframe component used on helicopters.
3. Casting C, 10-lb jet engine casting.
4. Casting D, 30-lb aircraft wheel casting.

These castings vary in wall thickness from 0.200 in. to 2.0 in. The gating and risering methods used on these casting designs for the commercial magnesium-rare earth metal-zirconium alloys were applied to EK31XA alloy. Standard separately-cast test bars, ½-in. diameter, were also cast from the melts used to pour these castings. The grain sizes were determined by the comparison method described by George.[12]

The castings were inspected in the usual way for foundry defects. Heat treatments were performed in both laboratory and production furnaces. The solution heat treatments were done in a protective atmosphere of air plus greater than 1 per cent SO_2 gas. The parts were then quenched as described in this paper. Precipitation heat treatement was done in regular furnace atmospheres.

After heat treatment the slope castings were sectioned, as described by Pearson,[11] while the other castings were sectioned as completely as possible from thick and thin sections, in order to obtain a complete picture of their room and elevated temperature tensile properties and elevated temperature creep strength. These sections were machined to conform to standard A.S.T.M. specified shapes for testing, or, when this was not possible, to dimensions listed as acceptable to the A.S.T.M.

Standard separately-cast test bars were used in determining effects of exposure at elevated temperatures on the room temperature and elevated temperature tensile properties of this alloy. Creep tests of 100 hr were made using separately-cast test bars. These tests were performed after exposures without applied stress for various times at elevated tempera-

tures. The exposure treatments were carried out in electrically heated Laboratory furnaces controlled to ± 5 F.

Tensile,[7,13] creep,[7,14] bearing[15] and shear[16] strength determinations have been described. Fenn and Gusack[17] report on methods of obtaining effects of strain rate on tensile properties, while Fenn[18] describes the determination of room and elevated temperature modulus values. Clapper[19] describes procedures for obtaining short-time creep data and the plotting of isochronous (equal-time) stress-strain curves for each test temperature. Alternate immersion testing in 3 per cent NaCl solution has also been described.[20]

REFERENCES

1. T. E. Leontis, *A.I.M.E. Transactions, vol. 185,* pp. 968-983 (1949).
2. T. E. Leontis, *A.I.M.E. Transactions, vol. 191,* pp. 987-993 (1951).
3. J. F. Pashak and T. E. Leontis, to be published.
4. T. E. Leontis and D. H. Feisel, *A.I.M.E. Transactions, vol. 209,* pp. 1245-1252, (1957).
5. A. C. Jessup, E. F. Emley, P. A. Fisher, U. S. Patent No. 2,549,955, Magnesium Elektron, Ltd. (April 24, 1951).
6. R. J. N. Payne and N. Bailey, *Metallurgia,* pp. 67 and 68, (August, 1958).
7. K. E. Nelson and F. P. Strieter, AFS Transactions, *vol. 59,* pp. 532-545, (1951).
8. K. E. Nelson, AFS Transactions, *vol. 61,* pp. 315-326, (1953).
9. Bulletin No. 141-27 "Crucible Melting of Magnesium Alloys;" The Dow Chemical Co., Midland, Mich.
10. Bulletin No. 141-41 "Magnesium Laboratory Methods;" The Dow Chemical Co., Midland, Mich.
11. W. E. Pearson, AFS Transactions, *vol. 64,* pp. 376-384, (1956).
12. P. F. George, AFS Transactions, *vol. 57,* p. 133, (1949).
13. A. A. Moore and J. C. McDonald, *A.S.T.M. Proceedings, vol. 46,* p. 970, (1946).
14. T. E. Leontis and J. P. Murphy, *A.I.M.E. Transactions, vol. 166,* p. 295, (1946).
15. A. A. Moore and J. A. Gusack, *A.S.T.M. Proceedings, vol. 56,* pp. 834-841, (1956).
16. A.S.T.M. Tentative Specification No. B 316-58T, "Aluminum-Alloy Rivet and Cold Heading Wire."
17. R. W. Fenn, Jr. and J. A. Gusack, *A.S.T.M. Proceedings, vol. 58,* pp. 685-696, (1958), to be published.
18. R. W. Fenn, Jr., A.S.T.M. Special Technical Publication No. 196, pp. 63-75, (1956).
19. R. B. Clapper, *A.S.T.M. Proceedings, vol. 58,* (1958), to be published.
20. J. D. Hanawalt, C. E. Nelson and J. A. Peloubet, *A.I.M.E. Transactions, vol. 147,* pp. 273-298, (1942).
21. H. M. Skelly and D. C. Sunnocks, AFS Transactions, *vol. 62,* pp. 481-489, (1954).
22. K. E. Nelson, Discussion to Skelly and Sunnocks' Paper, AFS Transactions, *vol. 62,* pp. 489-491, (1954).
23. K. E. Nelson, AFS Transactions, *vol. 63,* pp. 596-605, (1955).
24. K. E. Nelson and W. P. Saunders, AFS Transactions, *vol. 64,* pp. 364-372, (1956).

STRUCTURAL VARIABLES INFLUENCING MECHANICAL PROPERTIES OF HIGH STRENGTH CAST STEELS

a progress report

By M. C. Flemings, R. Green and H. F. Taylor

ABSTRACT

The initial phase of what is planned as a continuing study into the effect of solidification variables on the structure and properties of high strength cast steels is described. Plate castings of a modified 4330 alloy were cast in several different molding materials to obtain widely varying conditions of solidification. Cooling rates during solidification (at test bar locations) were varied from 300 F to less than 20 F/min. Thermal gradients during solidification were varied from 350 F to less than 70 F/in.

The differences in solidification conditions did not affect the macrostructure (columnar, original austenite grains) appreciably, but changed the microstructure in important ways. The internal structure of the dendrites became much finer with increasing cooling rate; also increasing cooling rates resulted in finer, more evenly distributed inclusions. Slow cooling rates tended to result in coarser agglomerated inclusions. Voids (microporosity) were evident in the central portion of all plates; the voids were most marked in an unchilled casting poured in dry sand.

Mechanical properties (including tensile and impact) were measured in various locations in the plates solidified under different thermal conditions. Properties were measured in the quenched and tempered condition. Tensile and yield strengths were essentially constant in all plates, but large variations in values of elongation and reduction of area were obtained depending on solidification conditions. These variations were attributed to differences in 1) microporosity, 2) microsegregation and 3) inclusion size and distribution.

As a quick check upon the effect of the most drastic conditions of chilling, a brief investigation was made of the structure and properties from metal melted by arc welding and then allowed to rapidly re-solidify.

INTRODUCTION

At present, high strength steel castings of a reasonably high degree of intricacy are obtainable from a few specialized steel foundries; these castings possess tensile strengths of the order of 200,000 psi, and have adequate ductility and impact strength for Ordnance applications. The production of intricate, thin walled castings with tensile strengths approaching 300,000 psi appears to be a practical possibility in the near future, and strengths in excess of this may someday be feasible (again with adequate ductility and impact strength).

Design engineers and metallurgists have not overlooked the potentialities of such high strength castings, and there is a growing demand to 1) develop techniques for producing steel castings with higher tensile strengths, while retaining adequate ductility and impact strength and 2) develop techniques which make it practical to cast more complex shapes in high strength steels.

Many of the problems presently encountered in producing high strength steel castings are related to the solidification mechanism of the alloy; these include hot tearing, shrinkage defects, some surface defects and segregation of nonmetallics and alloying elements. It should be possible to reduce or eliminate these problems by scientific control of the mode of solidification, together with careful control of such variables as melt chemistry, melting practice and heat treatment.

Results reported herein describe the first year's work at M.I.T. on solidification of high strength steel under simulated production conditions. This is a progress report of a continuing investigation supported by the Ordnance Department through Rodman Laboratory, Watertown Arsenal.

Studies to date have been concentrated primarily on the effect of mode of solidification on 1) microsegregation, 2) microshrinkage and 3) inclusion count and distribution. Solidification variables were 1) cooling rate and 2) thermal gradients. Tensile and impact properties were measured on test bars solidified under various thermal conditions. Results of these investigations, together with a survey of the literature are reported herein.

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LITERATURE SURVEY

Solidification and Segregation of Low Alloy Steel Castings

Microsegregation. Low alloy steel castings solidify, as do all cast alloys, by the formation and growth of dendrites. As the dendrites grow, carbon and other alloying elements are segregated into their liquid interstices and between the arms of individual dendrites. When, finally, the last metal freezes in the interstices of the dendrites, this metal contains the richest concentration of alloying elements. This micro-scale inhomogeneity is termed "microsegregation," or "coring."

Depending on thermal gradients during solidification and on nucleation,[1] the structure of any portion of a steel casting may be either of columnar or equiaxed dendrites, but the "coarseness" of each individual dendrite (distance between dendrite arms) is dependent only on rate of solidification.[2] A relatively coarse structure is obtained within individual dendrites when cooling is slow; when cooling is more rapid the dendrites consist of many more branches and sub-branches, and their internal structure is much "finer."

Thus, it is to be expected that the microstructure at the center of a large steel casting should be considerably different from that at the surface, and should also differ from the microstructure of a relatively thin casting; this has been observed experimentally by numerous investigators.[3-7]

Macrosegregation. Another type of segregation which can occur in steel castings is macrosegregation. This is the bulk movement of alloying elements from the surface of the casting towards the center ("positive" segregation), or the bulk movement of alloying elements from the center of a casting towards the exterior of the casting ("negative" segregation). It has been shown that in steel castings up to about 8 in. in section thickness, macrosegregation of elements other than carbon is negligible.[6,8] The segregation of carbon is usually slight and negative,[6,8] except under specialized conditions of solidification.[9]

Microsegregation Effect on Mechanical Properties

Carbon diffuses extremely rapidly in iron at elevated temperatures, and so in plain carbon steels simple austenitizing treatments are probably adequate to eliminate significant microsegregation. In low alloy steels, elements other than carbon segregate (including chromium, nickel and molybdenum); these elements diffuse much less rapidly than carbon, and so microsegregation in low alloy steels is quite difficult to eliminate by subsequent heat treatment.

Even after homogenization treatments at quite high temperatures, concentration gradients often remain in low alloy steels; these gradients (microsegregation) may have important effects on the hardenability and mechanical properties of a steel casting.[7]

When coring remains after heat treatment, the alloy-poor areas in the center of dendrite spines are not as hardenable as the dendrite surfaces, and may therefore decompose to high temperature constituents of iron carbide while the alloy-rich areas will transform entirely to martensite. Other microstructural effects are also possible, depending on the composition of the steel and on the specific heat treatment employed.[10-12]

Mechanical properties which are likely to be affected to the greatest extent by any variation in microstructure are ductility, reduction in area, impact strength and the yield-tensile ratio.[11] For example, Hollomon and Jaffee[13] have shown the importance of having a fully tempered martensitic structure to assure good notch-bar impact properties and a high yield-tensile ratio.

In any cast alloy system, the coarse dendritic structure obtained by slow cooling is less amenable to homogenization treatment than the finer structure obtained by rapid cooling.[6,14] Thus, it is to be expected that the central portions of a steel casting would exhibit lower mechanical properties than the exterior of the casting due to microsegregation effects alone. A number of investigators[6,8,15,16,17] have shown that mechanical properties (particularly reduction of area and elongation) are lower in heavy cast steel sections than in light sections.

Section Size Effect

However, it is not always possible to attribute these variations in mechanical properties directly to variations in microsegregation or to changes in any other single variable, since many factors are affected by increasing section size (or by different locations within a section). These include, microsegregation, macrosegregation, porosity, variation in inclusion count and distribution and effect of section size on heat treatment variables.

Studies of the effect of increasing homogenization temperatures and times on the mechanical properties of cast steels of various section sizes are one direct indication of the importance of microsegregation in determining properties of a cast steel (since the only beneficial effect of these treatments is considered to be the minimization of the microsegregation). Instances[18-20] have been cited where no change in hardenability was observed after increasing homogenizing times and temperatures, and no appreciable change in mechanical properties resulted. In other cases,[7,21] however, increased homogenization temperatures have been found to be definitely beneficial to mechanical properties of steels.

In work by Wallace, Savage and Taylor,[6] comparative mechanical properties were determined at the center and the surface of a large steel casting (5 in. in diameter). Substantially lower properties were obtained in the center of the casting than at the surface, and these lower properties were attributed to 1) microsegregation and 2) agglomeration of inclusions at the center. By inserting a thin core at the center of their casting around the location of the central test bar, transfer of metal from the central test bar to the exterior of the casting was prevented.

The core prevented agglomeration of inclusions at the center, and increased the tensile strength of the central specimen. Although tensile ductility was improved by the use of the core it still remained appreciably less than that in the outer areas. This

was taken as an indication that the interdendritic segregation due to the slower cooling rate was, of itself, deleterious to the mechanical properties of the cast steel.

Solidification and Microporosity

Another feature of the solidification process which may have a bearing on the properties of cast steel is porosity. As freezing progresses, it sometimes becomes difficult for liquid metal to feed into the voids which form as a result of solidification contraction. These voids may exist as large cavities or as a series of smaller voids between adjacent dendrites. The latter form of porosity is particularly pronounced in alloys which freeze with a wide liquid-solid zone ("mushy" zone). Also, gas which is liberated upon solidification of the metal exerts a positive pressure in the liquid channels and hinders feeding.[22]

If the cavities thus formed in cast steel are disposed along the centerline and are visible to the naked eye they are termed "centerline shrinkage;" finer, more randomly dispersed porosity is variously called "interdendritic shrinkage," "microshrinkage" and "microporosity."

Bishop and Pellini[23] have outlined the factors affecting centerline shrinkage in steel castings, and methods for elimination of this shrinkage. For example, they have shown that it is possible to feed a 1 in. thick plate which is 4½ in. long if a riser is placed at one end of the plate. The plate so risered is sound to within commercial nondestructive radiographic practice; i.e., 2 per cent definition.

The design rules formulated by Bishop and Pellini, therefore, provide for elimination of centerline shrinkage but do not necessarily guarantee complete absence of microporosity. In fact, it has been shown by Walther, Adams and Taylor[24] that to eliminate microporosity completely may require quite high thermal gradients during solidification.

The importance of microporosity upon the mechanical properties of steel castings is not well understood. Briggs and Gezelius[8] have attributed the low mechanical properties at the center of heavy cast steel sections to several factors, one of these being "low density." They do not, however, definitely attribute the low density to the presence of porosity. Several recent investigations[21,25] have shown that microporosity can affect tensile ductility seriously.

Whether or not microporosity is present to any marked degree in well fed steel castings is, however, a matter that has received little attention from investigators. At present, the most that can be said is that to minimize the effect of microporosity, casting design should be such as to insure that directional solidification takes place.[25]

Solidification and Inclusions

The effect of sulfur and phosphorus in lowering the elongation, reduction of area and impact properties is widely known. Sulfur is especially deleterious to these properties, and whenever possible the sulfur level of high strength steels should be kept below 0.01 per cent.[21] The damaging effect of sulfur is apparently due to formation of sulfide inclusions which, along with inclusions of oxides, silicates and occasionally nitrides or carbonitrides, are found at grain boundaries in cast steel. The form and type of these inclusions are strongly influenced by the deoxidation practice used.

Sims and Dahle[26] were the first to correlate inclusion characteristics with the degree of deoxidation employed. They classified inclusions into three types, 1, 2 and 3. Type 2 inclusions are a form of grain boundary "stringers" which seriously reduce ductility and impact strength and which are undesirable in cast steel; Types 1 and 3 are more globular in nature and less deleterious to properties.

Type 1 inclusions are most favorable for optimum mechanical properties, but melting practices which achieve these inclusions usually do not completely deoxidize the steel. Type 3 inclusions are only slightly less desirable than Type 1, and it is this type of inclusions that is sought in most melting practices. The effect of various deoxidation practices on inclusions and properties of cast steels have been summarized in a recent paper by Sims,[27] and the theory underlying these effects are given in an earlier paper by the same author.[28]

Wallace, Savage and Taylor[6] have shown that the size and distribution of inclusions are affected not only by the melting practice but also by solidification variables. They found that while inclusions at the exterior of their heavy (5 in. diameter) steel castings were fine and randomly dispersed, inclusions at the center tended to be coarser and to agglomerate (frequently in chain-like arrangements).

High Strength Steels

The past decade has seen a tremendous increase in the need for high strength steel castings as well as forgings.[29,31] However, research efforts on high strength steels have been confined for the most part to the alloying and heat treatment of forged steels,[28-30] and only limited work has been carried out with cast steels.[21,32]

Probably the first approach to high strength steels (in forgings) was to employ the standard 4340 analysis but to temper it at 400 to 450 F instead of the standard tempering treatment in excess of 800 F. By doing so, the strength level was advanced from 190,000 to 270,000 psi in a single step.[28] However, most high strength forged steels in use today have been tailored to provide increased strength with little or no decrease in established values of ductility and toughness; essentially all cast alloys have also been "tailor-made" to these ends.

In development of compositions and heat treatments for low alloy steel, it has been necessary to avoid tempering the quenched steel in the so-called "500 F embrittlement range." When impact test values are plotted against tempering temperature, impact strength is usually a minimum near 500 F, although by the addition of silicon the embrittlement range can be raised to over 700 F.

Some steel analyses employ silicon additions to permit tempering at temperatures as high as 600 F, to allow more complete relief of residual quenching stresses with consequent improvement in yield strength and ductility. Typical modifications of com-

TABLE 1 — CHEMICAL ANALYSES OF EXPERIMENTAL HEATS

Heat	Modified 4330 Analyses							
	C	Mn	Si	Ni	Cr	Mo	P	S
A	0.27	1.12	0.57	1.80	0.93	0.42	0.006	0.012
B	0.24	1.20	0.62	1.79	0.91	0.41	0.007	0.009
C	0.26	1.08	0.58	1.79	0.90	0.40	0.009	0.009
D	0.28	1.12	0.59	1.83	0.94	0.41	0.007	0.009
E	0.27	1.10	0.56	1.80	0.90	0.41	0.008	0.011
F	0.28	1.15	0.42	1.91	0.96	0.42	0.007	0.009
G	0.29	1.12	0.51	1.87	0.96	0.43	0.010	0.017
Aim	0.27	1.10	0.55	1.85	0.90	0.40	—	—

positions of alloys of the 4300 series for casting purposes are given in recent research publications on the subject.[21,32]

SCOPE OF WORK

Work reported herein has concentrated primarily on a study of the effect of solidification variables on 1) microsegregation, 2) microshrinkage and 3) inclusion count and distribution. The solidification variables which were measured and controlled were cooling rate and thermal gradients. Most of the work reported has been carried out by study of simple ¾-in. plate castings, end risered. High strength steel was employed, and a variety of molding materials (plus metal chills) were used to obtain different solidification conditions.

Castings were poured in 1) end chilled zircon sand mold, 2) end chilled dry sand mold, 3) dry sand mold without chill, 4) ethyl silicate bonded mullite mold without chill at room temperature, 5) ethyl silicate bonded mullite mold without chill heated to 125 F and 6) ethyl silicate bonded mullite mold without chill heated to 1600 F.

In one portion of the investigation, some of the high strength alloy was solidified rapidly by freezing a weld bead; microstructures and mechanical properties were examined.

The alloy analysis was modified 4330 with a nominal analysis as shown in Table 1. All specimens were quenched and tempered, and the tempering treatment was designed to produce a strength level of approximately 200,000 to 210,000 psi. Thermal analyses of the solidifying castings were made to determine cooling rates and thermal gradients during solidification. Sections from the plates were examined by metallography and microradiography, and tensile and impact test bars were taken from different locations of the plates. The experimental program was designed to determine the inter-relationship between thermal variables, structural variables and mechanical properties.

The plan of research follows that of a similar undertaking at M.I.T. (under sponsorship of U. S. Army Ordnance, Pitman-Dunn Laboratories, Frankford Arsenal) to determine why maximum mechanical properties were not being obtained in aluminum alloy castings; by isolating the effects of such variables as gas content, soundness, thermal gradients and solidification rates, and by developing practical means for insuring optimum conditions of solidification the tensile strengths of aluminum alloy castings were doubled and their ductility increased several hundred per cent.[22] It is expected similar practical improvements in the properties of steel castings will ultimately accrue from this investigation.

PROCEDURE

Plate Pattern Studies

Molding and Melting. The test pattern in this study was a 7 x 5 x 1 in. plate with a 3 in. diameter riser positioned over one end (Fig. 1). The dimensions of the plate and location of the riser were such that all plates were expected to be sound by x-ray to at least 2 per cent definition.[23] Exothermic sleeves, of dimensions shown in Fig. 1, were used around the riser to aid feeding, every effort being made to insure a sound casting. A circular downsprue was used with a 2:1:1½ gating ratio. In the chilled plates, a water cooled steel chill 6½ x 1½ x 1¼ in. was positioned, as shown in Fig. 1.

Molding materials used were zircon sand, silica sand and ethyl silicate bonded mullite refractory, the ethyl silicate mix being that of a proprietary process. These molding materials were chosen to gain a range of mold chilling (or insulating) properties. Heats poured are listed in Table 2.

Thermal analysis was performed (during solidification) on five castings to determine the effects of the different mold materials on the progress of solidifica-

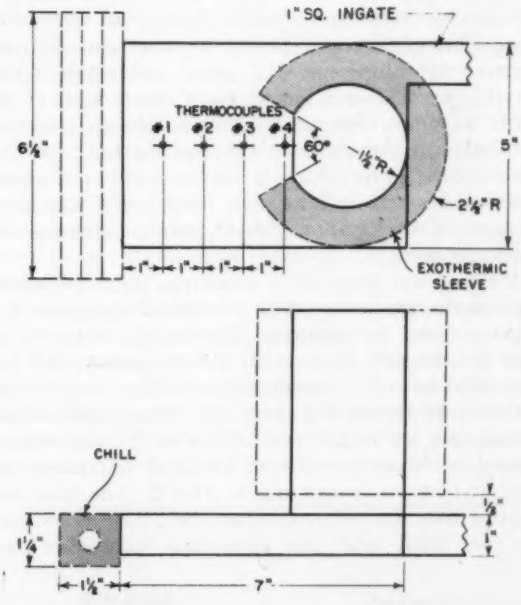


Fig. 1 — Sketch of test pattern with locations of thermocouples.

TABLE 2 — SUMMARY OF PLATE HEATS POURED

Heat	Mold Material	Mold Preheat Temp.	Rigging
A	Dry Zircon Sand	Room	Chilled
B	Dry Silica Sand	Room	Chilled
C	Dry Silica Sand	Room	Unchilled
D	Ethyl-silicate-mullite	Room	Unchilled
E	Ethyl-silicate-mullite	1250 F.	Unchilled
F	Ethyl-silicate-mullite	1600 F	Unchilled
G	Dry Silica Sand	Room	Chilled

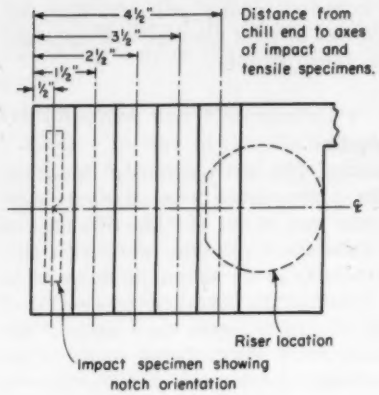


Fig. 2 — Location of tensile and impact test bars in plate pattern.

tion. Thermal data were obtained by placing thermocouples in the mold cavity, in locations shown in Fig. 1. Four plate castings were poured in each heat (except *E* and *F*, in which only two castings were poured). All castings in each heat were rigged identically; one casting was used for thermal analysis and the remainder were used for physical or mechanical testing.

A modified 4330 analysis was chosen as a typical high strength steel for purposes of this work, the 4330 analysis being modified principally by raising the silicon and manganese levels. Table 1 lists the aim analysis employed, and the analyses of each heat.

High Purity Melting Stock. All heats were made from high purity virgin melting stock and alloys. Electolytic iron was used for all heats except heat *G* (in order to reduce the sulfur and phosphorus levels to a minimum). Armco iron was used for heat *G*. The carbon was supplied by a carbon alloy previously made from armco iron and cabot coke. Calcium-manganese-silicon alloy plus aluminum were the deoxidizers used.

Melting was done in a magnesia lined induction furnace, and a ladle with a rammed magnesia lining was used for pouring. The charge consisted of iron (electrolytic or armco), nickel squares and ferromolybdenum. At meltdown ferrosilicon was added followed by the carbon alloy and ferrochrome. Ferromanganese was added at 3020 F after the slag was removed. Another ferrosilicon addition was made immediately prior to tapping at 3100 F. The heat was tapped into the calcium-manganese-silicon deoxidizer in the ladle, and the aluminum deoxidizer was plunged into the metal in the ladle at two-thirds tap. Molds were poured immediately after tapping.

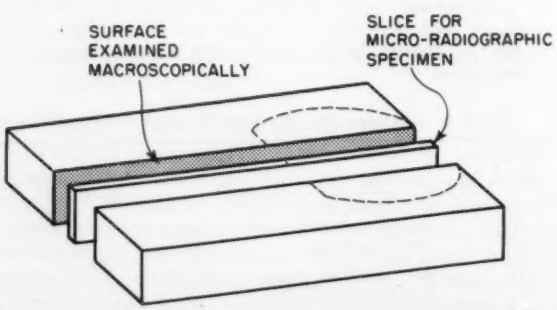


Fig. 3 — Location and macro and microradiography specimens.

Heat Treatment and Testing. The plates used to provide tensile and impact specimens were cut into 1 in. square bars before heat treating (Fig. 2). Since the purpose of this work was to determine the effects of varying solidification rate, no effort was made to maximize the strength levels obtained through heat treating; the following heat treating schedule was used for all test bars:

Solution 1800 F, 2 hr, air cool.
Harden 1650 F, 2 hr, oil quench.
Temper 700 F, 2 hr, air cool.

The 700 F tempering treatment was chosen to be just above the region expected to produce low impact properties.

All test castings were x-rayed before any sectioning was done, and proved to be sound at 2 per cent definition. Test bars for tensile and impact specimens were then cut parallel to the end furthest from the riser, in locations shown in Fig. 2. Standard 0.505 in. diameter tensile test specimens with a 2 in. gage length were machined from one plate in each heat. Standard ⅜-in. square notched impact specimens were made from a second plate casting in each heat. The location of the impact specimens corresponded to those of the tensile specimens, and the notch of the impact bars was made perpendicular to the plane of the plate (Fig. 2).

Tensile strength, yield strength, per cent elongation, per cent reduction in area and hardness were measured from the tensile specimen, the yield strength being obtained by 0.2 per cent offset method with a recording extensometer. The impact tests were made at −20 F.

Microscopic Examination. Microstructures were examined in the as-cast and heat treated condition. The specimens used to obtain the heat treated microstructures were prepared from the impact bars after they were broken; the location of the specimens being approximately ⅛-in. from the notch. The as-cast microstructures were taken in locations corresponding to the centers of the tensile and impact bars. These specimens were also used to show the inclusions present in the as-cast condition. Nital (1 per cent) was used to etch the heat-treated and as-cast specimens.

Macroscopic Examination. Cross-sections from the chilled end to the risered end were examined macroscopically to show the as-cast grain structure resulting from different rates of solidification. Figure 3 shows the orientation of the section examined. These specimens were etched with a 30 per cent (by weight) solution of ammonium persulfate applied at 90 F for 1½ min.

Microradiographic examination. Thin slices cut from cross-sections of plates in the as-cast condition were examined by microradiographic techniques. The orientation of these slices is shown in Fig. 3. They were approximately ⅛-in. thick as originally cut, and were then ground by taking equal amounts from both sides, until their thickness was under 0.020 in.

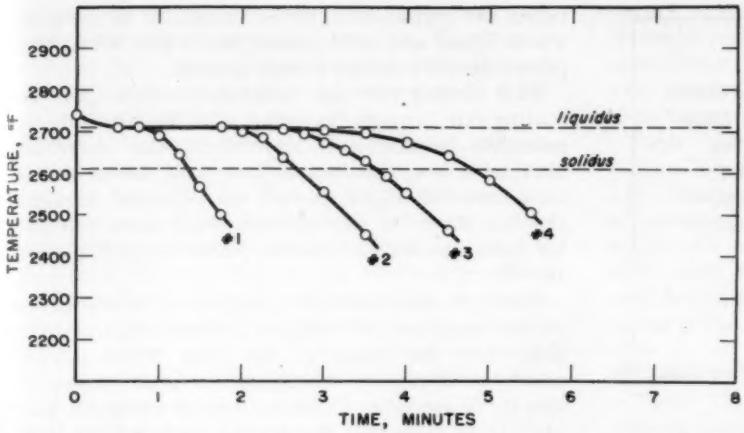


Fig. 4 — Cooling curves for plate cast in zircon sand, end chilled (heat *A*). Location of thermocouples shown in Fig. 1.

Grinding was done using a magnetic chuck (care was taken to keep the slice flat and of uniform thickness).

The slices were polished by hand with fine emery paper to remove the grinding scratches, and then x-rayed. With a thickness under 0.020 in., microscopic discontinuities such as interdendritic shrinkage are of sufficient size relative to the total thickness that a radiograph will show their presence.

Chemical analyses were obtained for each heat; these are given in Table. 1.

Ultra-High Solidification Rate Studies

Examination was made of the effect of extremely high solidification rates on the dendritic structure obtained in cast 4330 steel. The ultrahigh solidification rates were obtained by the melting and subsequent resolidification of a weld zone in a cast steel plate. Cooling rates (during solidification) obtained by this procedure were calculated to be approximately 700 F/min, as compared with maximum cooling rates in the chilled plates of about 300 F/min. Homogenization treatments of the structures so obtained have not been examined, but the effect of tempering temperature on hardness and bend test properties of the weld zone have been studied.

RESULTS AND DISCUSSION

Thermal Analysis

Cooling curves were obtained for each type of test plate casting made (zircon sand, end chilled; silica sand, end chilled; etc.). Figures 4 and 5 are representative of the curves obtained. Figure 4 presents thermal data for a plate cast in zircon sand, end chilled; and Fig. 5 presents data for a plate cast in a preheated ethyl-silicate mold. Cooling curves for the remaining plate castings may be found in the original report on which this paper is based;[33] however, Figs. 6, 7 and 8 summarize the important thermal data from the original curves.

Thermal analysis of early heats indicated that the liquidus of the alloy was at 2710 F and the solidus at about 2610 F. The liquidus temperature of some of the castings poured in the investigation showed minor deviations from 2710 F; when this occurred, the cooling curves were raised or lowered slightly so that the liquidus "hold" was moved to 2710 F. This was done to make it feasible to correlate thermal data from the various heats. In any event, even quite large deviations from the liquidus and solidus temperatures would not change the qualitative conclusions to be drawn from Figs. 6, 7 and 8 discussed hereafter.

Figure 6 compares the average cooling rate during solidification for various locations in the cast plate (at the centerline). In all castings the cooling rate through the solidification range is greater near the end farthest from the riser and decreases as the riser is approached. At a given location in the cast plates, the casting poured in zircon sand with an end chill

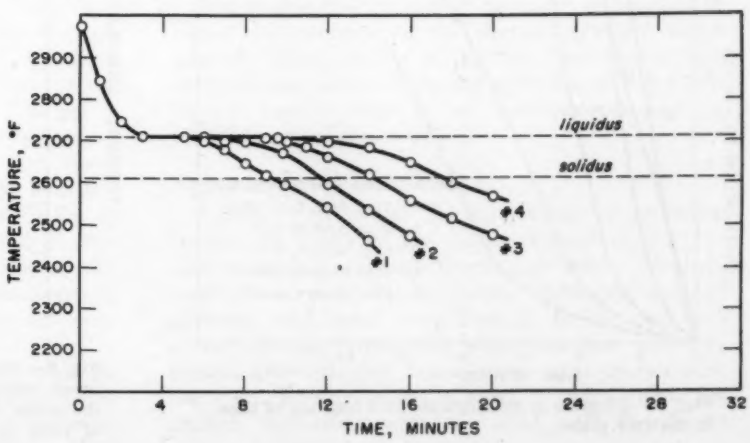


Fig. 5 — Cooling curves for plate cast in ethyl-silicate, no chill. Mold at 1250 F (heat *E*). Location of thermocouples shown in Fig. 1.

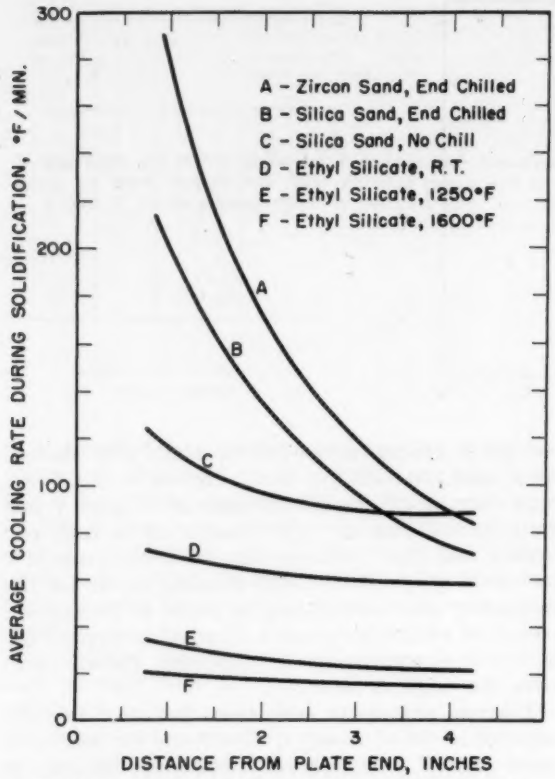


Fig. 6 — Average cooling rate during solidification (from 2680 F to 2610 F) as a function of location in the cast plates.

shows the greatest cooling rate during solidification; the casting poured in the hot ethyl-silicate mold shows the slowest cooling rate.

Average Cooling Rates

The cooling rates, shown in Fig. 6, are the average cooling rates between 2680 F and 2610 F. They were obtained from the original cooling curves by dividing the temperature difference 2680 F-2610 F by the time required for the metal to cool from the higher to the lower temperature (at a particular thermocouple location). The temperature 2680 F was arbitrarily chosen between the liquidus and solidus temperatures; the region from 2680 F to 2610 F is a region where liquid and solid coexist, but where substantial percentages of solid are already present.

This cooling rate was calculated, rather than the cooling rate through the entire solidification interval, primarily because it is more easily and accurately obtainable from the thermal data. Also, earlier studies on aluminum alloys showed that thermal data for the last stages of solidification were more valuable for interpretation of casting structure and properties.[22]

Figure 7 illustrates the progress of solidification in the cast plates in a slightly different manner. The data show the length of the plate which is completely solid (at the centerline) vs. time; points for this curve are taken directly from the original cooling curves. Note that the interface between the fully solid material and the liquid-solid mixture ("mushy" zone) moves most rapidly in the case of the zircon sand mold, end chilled, and least rapidly for the unchilled ethyl-silicate mold (1600 F). In all cases, however, solidification is directional; that is, the completely solid interface moves progressively from the plate extremity towards the riser.

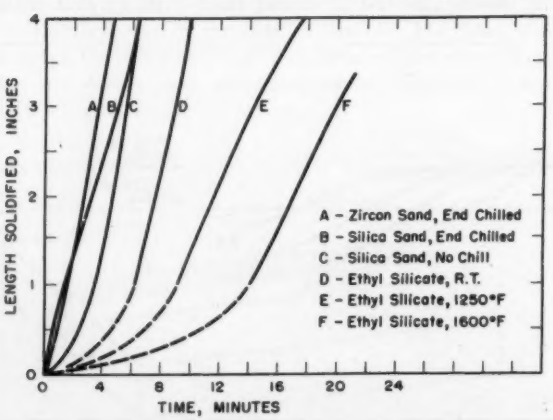


Fig. 7 — Progress of solidification as a function of time in the cast plates.

Thermal Gradients

It has been found that for some alloys[22] directional solidification, in the sense defined above, does not always insure a casting free of shrinkage defects. Thermal gradients must also be present in the liquid-solid

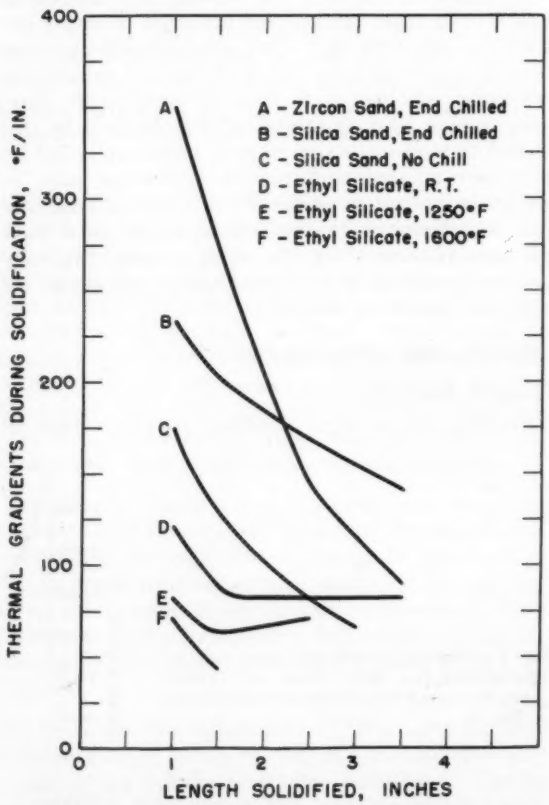


Fig. 8— Thermal gradients during solidification (between 2680 F and 2610 F) as a function of location in the plates. Length solidified is distance from the end of plate to the beginning of liquid-solid zone.

region of the solidifying casting. The gradients must be steep enough to maintain open feed channels so the liquid metal from the riser can flow easily through the interstices of the dendrites to accommodate solidification shrinkage.

Figure 8 illustrates these thermal gradients in the six different plate castings. The gradients are those existing at a particular location in the solidifying plate when that point is just completing solidification; they are the average gradients between 2680 F and 2610 F.

In Fig. 8, it may be seen that thermal gradients (near the plate extremity) are greatest for the zircon sand-end chilled casting and least for the ethyl-silicate (1600 F) casting. Near the riser, thermal gradients are steeper in the case of the end-chilled casting poured in silica sand than in the case of the end-chilled casting poured in zircon sand. They are also greater in the case of the casting poured in ethyl-silicate at room temperature than in the case of the unchilled casting poured in silica sand.

This rather unexpected result is undoubtedly due to the effect of the riser on thermal gradients. The heat effect of the riser should be felt to a greater extent when the molding material is more insulating.

Figure 8 was obtained by first plotting the progress of the solidus isotherm vs. time (Fig. 7). The progress of the 2680 F isotherm for each casting was then plotted on the same coordinate. Thermal gradients in the liquid-solid zone at any point in the casting were then readily calculated from the relationship

$$(\text{T.G.})_{L_1} = \frac{2680 - 2610}{L_2 - L_1}$$

where:

$(\text{T.G.})_{L_1}$ = thermal gradients at point 1 (F/in.)

L_2 = distance from plate extremity to 2680 F isotherm (in.)

L_1 = distance from plate extremity to 2610 F isotherm (in.)

As-Cast and Heat Treated Structures of Plate Castings

One casting each of Heats *A* through *D* were sectioned, as shown in Fig. 3, the cut surface polished and etched and the macrostructure examined. All castings were composed largely of columnar grains, the only major differences between the castings being that those that were end chilled exhibited a marked columnar zone which extended a maximum of about 1 in. from the chill face of the casting towards the riser.

In each of the four castings, microstructures in the as-cast condition were examined at intervals along the length of the plate. Microstructures at locations 1, 3 and 5 in. from the plate end are shown in Fig. 9. In the two end chilled castings, a large difference may be seen in the dendritic structure between the end near the chill and the end farthest from the chill, the finer dendritic structure being obtained near the chill.

Comparing the structures of the different castings, it is evident that as the molding material becomes more insulating, the dendritic structure becomes coarser (spacing between individual dendrite arms becomes greater). In general, a qualitative correlation is evident between the coarseness of the dendritic structure at any given location and the cooling rate of that location, as shown in Fig. 6.

Figure 10 illustrates the microstructures at the various locations in the plate after heat treatment. The homogenization treatment employed is observed to have been insufficient to completely homogenize any of the structures, including those quite near a chill. Hard, interdendritic areas of segregate remain, and these trace the same dendritic structure apparent in the as-cast condition.

Microporosity in the Plate Castings

Figure 11 shows sketches of microradiographs of sections approximately 0.020 in. thick, taken from four of the plate castings in the manner shown in Fig. 3. Microvoids are visible in all four of the castings shown. In the end chilled casting poured in a zircon mold few voids are apparent; these are in the region between about 3-4.5 in. from the plate end. In the casting poured in a silica sand mold, end chilled, voids are present between about 2.5 - 4.5 in., and these voids are somewhat more pronounced than those in the previous casting.

In the unchilled casting poured in silica sand, the voids are seen to be quite marked as compared with the previous two castings. They are present between about 2 and 4.5 in., and are present as elongated cavities between the columnar grains (extending almost to the plate surface in certain locations). In the case of the casting poured in the ethyl-silicate mold (room temperature), microvoids are again present between about 3-4.5 in., but these voids are fine and not well defined.

It has been concluded that the voids apparent in the microradiograph are microporosity. It is to be expected that the relative amounts of microporosity in the various castings should be related to the thermal gradients measured during solidification (Fig. 8).[22] Microporosity increases from the casting of Heat *A* to the casting of Heat *C* as would be expected since thermal gradients decrease in this order. However, the thermal gradients in Heat *D* are substantially lower then those of the preceding heats, but microporosity is generally less evident.

One reason for the lack of complete correlation between the steepness of the thermal gradient curves and the microporosity is the fact that thermal gradients could be obtained only up to about 3 to 3.5 in. In this region, little or no microporosity was observed in any of the castings.

Appearance of Inclusions

Figure 12 illustrates the appearance of inclusions in the cast plates of low sulfur content metal. The figure presents microstructures (at 300×) of three locations in each of four plates. Inclusions were all globular, and none were Type 2. In the two end-chilled castings inclusions were fine and well dispersed, although they became somewhat coarser near the riser. In the casting poured in silica sand, unchilled, inclusions were fine and well distributed

near the plate extremity, but remaining locations of the plate showed larger agglomerated inclusions.

Inclusions in the ethyl-silicate mold were also coarse as compared with inclusions in chilled sand plates. In general, it appears that rapid cooling rates result in finer, more evenly distributed inclusions; slower cooling rates result in coarser, agglomerated inclusions.

Figure 13 illustrates inclusions obtained in a heat of somewhat higher sulfur content than that employed in other heats (0.017 per cent sulfur, Heat *G*). Here, the deoxidation practice used yielded in-

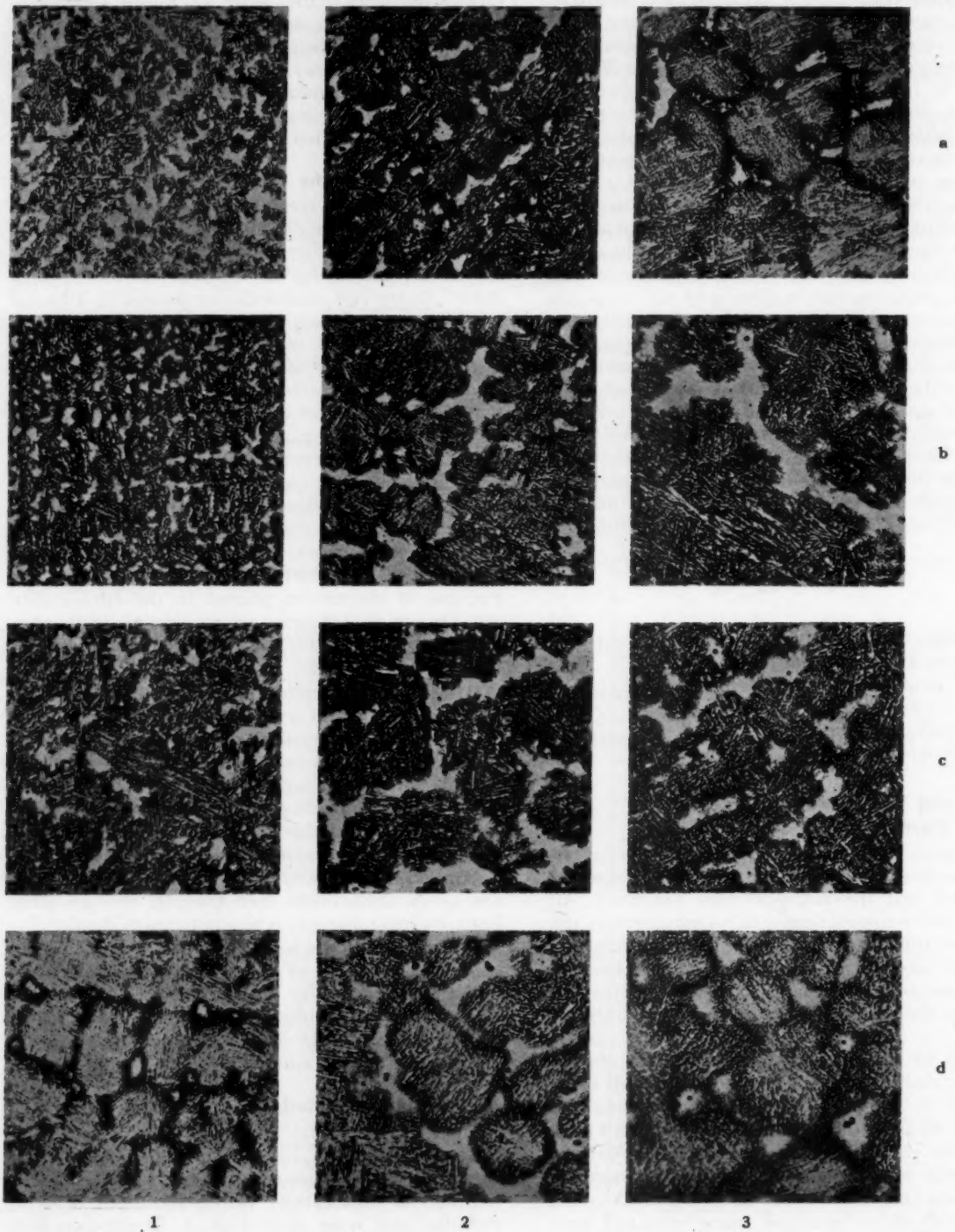


Fig. 9 — As-cast microstructures from test plates. Etched. 50 ×. Distance from plate end — *row 1*, 1 in.; *row 2*, 3 in.; *row 3*, 5 in. Horizontally — *row a*, zircon sand mold, end chilled; *row b*, silica sand mold, end chilled; *row c*, silica sand mold, no chill; *row d*, ethyl-silicate mold, no chill.

clusions of the unfavorable Type 2; the Type 2 films became somewhat coarser and stringier towards the riser end of the plate.

Mechanical Properties

Table 3 lists the mechanical properties obtained in the plate castings after heat treatment according to the schedule given earlier. The heat treatment employed was designed to produce a yield strength of approximately 180,000 psi, and an ultimate tensile strength of approximately 210,000 psi. Nearly all bars tested were 175,000-180,000 psi yield strength and 205,000-210,000 psi tensile strength. Elongation, reduction in area and impact strength varied markedly

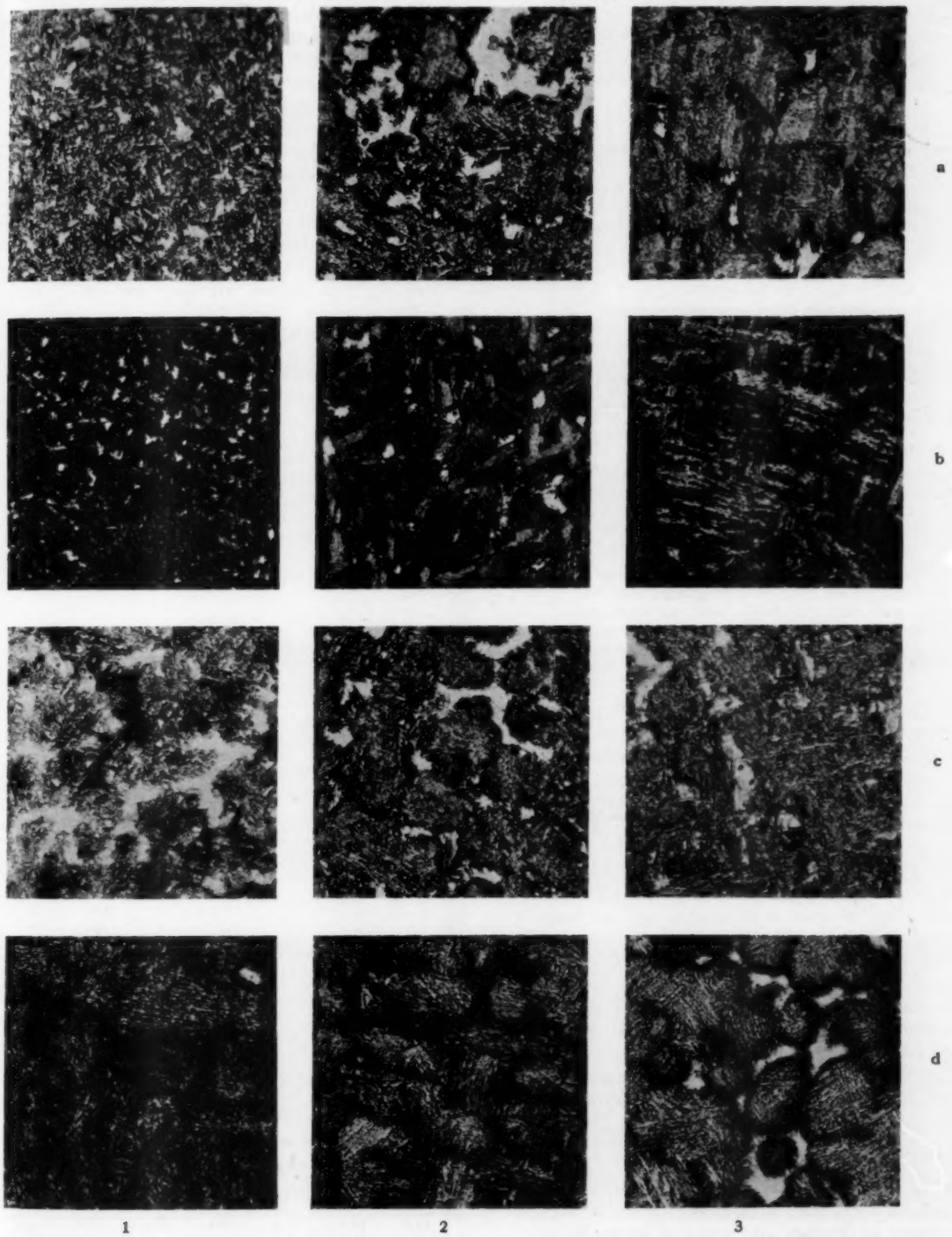


Fig. 10 — Microstructures from test plate after heat treatment. Etched. 50 ×. Distance from plate end — *row 1*, 1 in.; *row 2*, 3 in.; *row 3*, 5 in. Horizontally — *row a*, zircon sand mold, end chilled; *row b*, silica sand mold, end chilled; *row c*, silica sand mold, no chill; *row d*, ethyl-silicate mold, no chill.

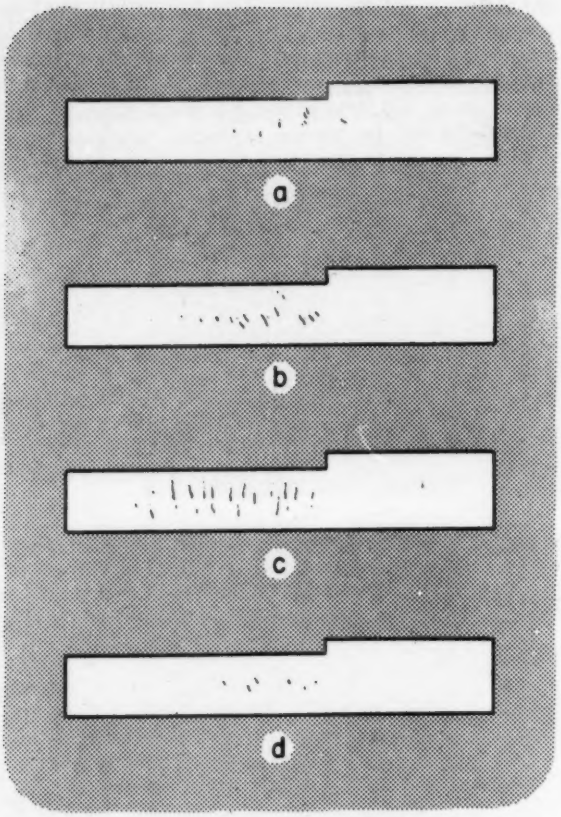


Fig. 11 — Sketches of microradiographs of sections cut from cast 1 in. plates. (a) zircon sand, end chill; (b) silica sand, end chill; (c) silica sand, no chill; (d) ethyl-silicate, room temperature. (The microradiogaphs are sketched for clarity as the voids are too fine to be reproduced through halftone screen).

depending on the type of mold employed for the individual plate, and on the location within the plate.

Figure 14 shows the elongation and reduction of area at various locations in the plates of the sand cast, low sulfur heats (Heats A, B and C). All plates show optimum properties at the location farthest from the riser, the best being obtained in the plates which were chilled. Lowest properties were obtained about 2-4 in. from the end of the plate, with improvement being obtained at a distance of 5 in. from the plate end (under the riser).

Figure 15 illustrates the same mechanical properties for the three castings of low sulfur content poured in ethyl-silicate molds. Except for the anomalous behavior of elongation values of Heat *F*, the elongation and reduction of area properties for the ethyl-silicate mold are of the same order of magnitude or slightly lower than the comparable properties for the casting poured in silica sand with no chill.

Microporosity Effect

Comparing the mechanical property data of Fig. 14 with the microradiographs of Fig. 11, it may be seen that the lowest properties are obtained generally in the areas of greatest microporosity. In the case of the ethyl-silicate mold (Heat *D*, Fig. 15), minimum properties are also obtained in the locations where microporosity is greatest. However, if microporosity were the only variable acting, it would be expected that the properties of the casting poured in ethyl-silicate would be somewhat higher than the properties of the casting poured in silica sand (because mircroporosity is less).

That the properties of the casting in the ethyl-silicate mold are not higher than those in the sand mold is taken as an indication that other factors are important in determining properties, including microsegregation and inclusion count and distribution.

Elongation and reduction of area properties of the high sulfur casting poured in silica sand with no chill are listed in Table 3. These properties are generally lower than those of the comparable casting poured with low sulfur metal (Heat *C*), indicating the deleterious effect of the higher sulfur and the consequent unfavorable type inclusions.

Impact strengths of all heats poured are also listed in Table 3. The values are low for the tensile

TABLE 3 — MECHANICAL PROPERTIES OF PLATE CASTINGS

Yield Strength (psi, 0.2% offset)

Bar Position	Heat A	B	C	D	E	F	G
1	175,500	177,500	180,000	177,500	172,500	172,500	176,250
2	177,500	176,500	175,500	177,500	173,750	165,000	176,250
3	176,250	178,500	177,000	177,500	173,750	182,500	177,500
4	177,500	176,000	175,000	177,500	173,750	182,500	175,000
5	180,000	175,000	—	177,500	172,500	182,500	175,000

Ultimate Tensile Strength (psi)

Bar Position	Heat A	B	C	D	E	F	G
1	209,250	210,500	211,750	211,000	206,250	207,000	209,000
2	210,000	219,750	208,000	210,500	200,250	207,500	208,750
3	209,500	211,750	209,250	211,000	207,500	207,500	207,500
4	211,750	209,250	207,750	208,750	205,000	206,000	207,500
5	212,000	208,000	207,500	209,250	203,750	207,000	204,500

Elongation (% in 2 in.)

Bar Position	Heat A	B	C	D	E	F	G
1	12	13	10	7.5	9	11	8
2	8	9	6	7	1.5	11	4.5
3	5.5	7	6	5	5	10	4
4	5	8	5	5.5	5	10	6
5	7	10	8	10	8	10	8

Reduction in Area (%)

Bar Position	Heat A	B	C	D	E	F	G
1	39.3	38.5	26.4	17.9	25.0	—	18.6
2	22.4	22.6	8.2	17.6	3.6	—	7.4
3	10.2	18.9	11.9	7.1	6.4	—	5.0
4	7.4	18.9	10.6	7.8	10.2	—	10.0
5	19.5	24.2	20.2	25.4	22.0	—	21.4

Impact Strength at −20 F (ft lb)

Bar Position	Heat A	B	C	D	E	F	G
1	14.9	14.5	11.8	10.0	14.2	11.2	8.9
2	13.6	12.7	6.2	9.2	14.2	7.5	9.5
3	8.6	10.6	8.9	10.9	12.4	12.4	7.3
4	5.4	10.3	10.0	9.7	10.6	12.7	10.9
5	10.6	8.9	12.4	13.0	13.6	12.4	7.8

strengths of the metals used in this investigation, and show no particular correlation with location in the plate. This lack of correlation is felt to be partly due to the fact that duplicate tests were not made; only one test was made for each location in the plate. However, it is also felt that the tempering temperature employed may have been too close to the 500 F embrittlement range.

Ultra-High Solidification Rate Studies

To further examine the effect of cooling rate on structure and properties, the modified 4330 alloy used in this investigation was solidified under extremely fast cooling rates obtained by melting and resolidification of a large plate section by arc welding. Cooling rates obtained were calculated to be approximately 700 F/min, as compared with maximum cooling rates in the chill plates of the order of about 300 F/min. Figure 16 illustrates a typical structure obtained in the weld deposits. An extremely fine dendritic structure results; note the photomicrograph is magnified to 500×.

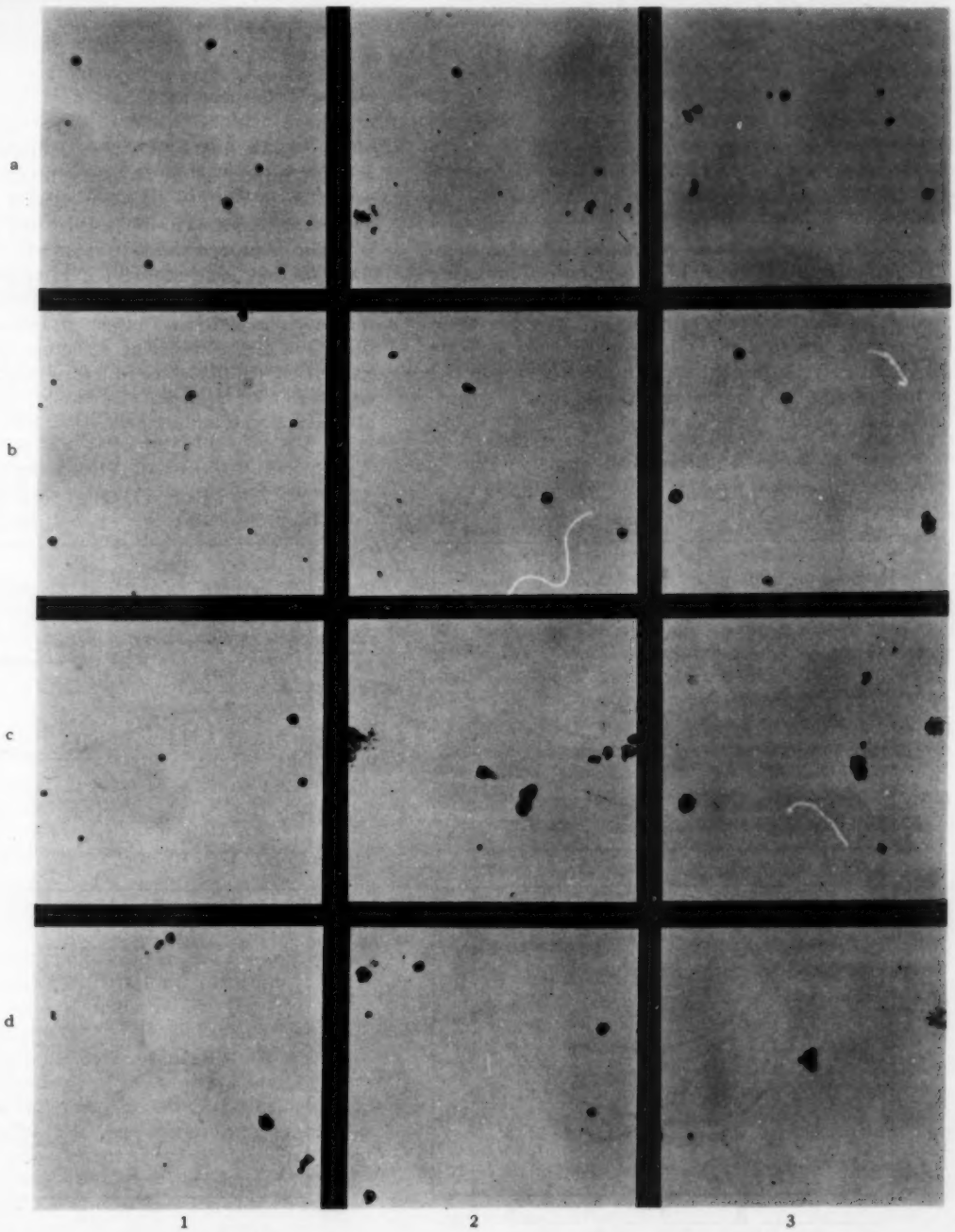


Fig. 12 — Appearance of inclusions in test plates. Unetched, 300 ×. Distance from plate end — *row 1*, 1 in.; *row 2*, 3 in.; *row 3*, 5 in. Horizontally — *row a*, zircon sand mold, end chilled; *row b*, silica sand mold, end chilled; *row c*, silica sand mold, no chill; *row d*, ethylsilicate mold, no chill.



Fig. 13 — Appearance of inclusions in plate of high sulfur (0.017 per cent) metal cast in silica sand mold, without chill. Left to right — 1 in., 3 in. and 4 in. from plate end. Unetched. 300 ×.

Cooling in the weld deposit after solidification is so rapid that martensite and fine carbides are formed.[34] In fact, excellent mechanical properties can be obtained in a structure of this type by employing only tempering treatment after the casting (no austenitizing treatment). Figure 17 illustrates a tempering curve for a series of welds. Hardness (Rc) was measured at tempering temperatures varying from room temperature up to 1000 F; these hardnesses indicate that tensile strengths result which are the same order of magnitude as are obtained in homogenized, quenched and tempered castings.

Also, bend tests performed on weld deposits indicated a minimum of 10 per cent elongation in the weld, which compares favorably with the mechanical properties obtained in the heat treated cast plates.

While homogenization treatments of weld deposits have not yet been studied, it is apparent that these weld deposits are an excellent tool for studying the possible beneficial effect of intense homogenization. The extremely fine structure obtained should lend itself to almost complete elimination of microsegregation of alloying elements at ordinary temperatures in a reasonable time.

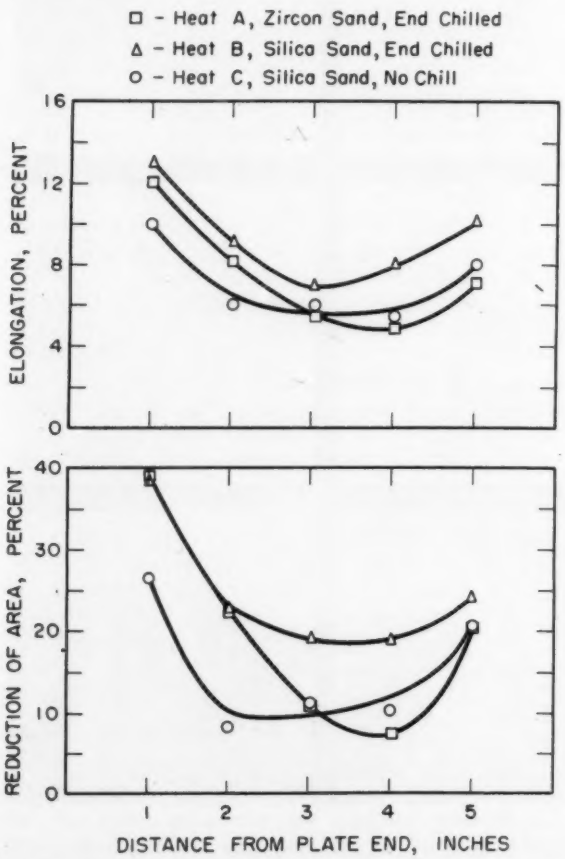


Fig. 14 — Elongation and reduction of area in zircon and silica sand plates.

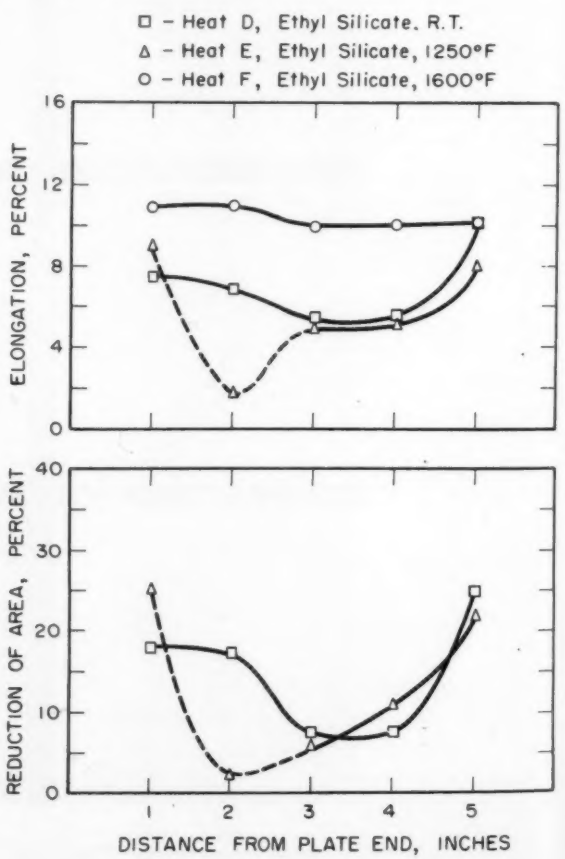


Fig. 15 — Elongation and reduction of area in ethlysilicate plates.

SUMMARY AND CONCLUSIONS

Plate castings of modified 4330 alloy were poured in a variety of different molding materials to obtain different solidification conditions. Molds employed were 1) zircon sand mold with a metal chill at the end of the plate, 2) dry sand mold chilled similarly, 3) dry sand mold without chill, 4) ethyl-silicate bonded mullite mold without chill, room temperature, 5) ethyl-silicate bonded mullite mold without chill, preheated to 1250 F and 6) ethyl-silicate bonded mullite mold preheated to 1600 F.

A wide range of solidification conditions were obtained in the above molds. Cooling rates as high as 300 F/min, and thermal gradients as high as 350 F/in. were obtained 1 in. from the plate end (in the end chilled, zircon sand mold). Cooling rates and thermal gradients decreased along the length of the plate. In the most insulating mold (ethyl-silicate bonded mullite, heated to 1600 F), cooling rate was only 20 F/min and thermal gradients were 70 F/in. at 1 in. from the plate end. The cooling rate and thermal gradients measured were those existing at a given location in the plate near the end of solidification at that location.

Macrostructures of the plates were not appreciably affected by varying thermal conditions. All castings showed predominantly columnar (original austenite) grains. Microstructure was, however, affected markedly; fast cooling rates produced an extremely fine structure within the austenite dendrites. Slower cooling produced a substantially coarser structure. The "fineness" of the structure (spacing between dendrite arms) is apparently dependent only on cooling rate.

Microsegregation in the as-cast structure was not eliminated by the heat treatment employed. Even the finest chilled structures showed some segregation after heat treatment although these should be most amenable to elimination of such concentration gradients.

All castings were examined radiographically and found to be free of macroscopic centerline shrinkage. Sections 1/8-in. thick cut from the plates were also radiographically sound. However, microradiography of sections 0.020 in. thick disclosed voids at or near the centerline in all plates examined. The amount and distribution of these voids were dependent on the mold material employed and it was concluded that the voids were microshrinkage. The unchilled plate cast in silica sand exhibited the most severe porosity (between 2.5 and 4.5 in. from the plate end).

Inclusion size and distribution in the cast plates were also dependent on mold material and on location within the plate. Fast cooling rates resulted in fine inclusions, evenly distributed; slower cooling rates produced coarser, more agglomerated inclusions.

Mechanical properties were determined at various locations in the plate castings; properties determined were ultimate tensile strength, yield strength, elongation, reduction of area and impact strength. The castings were heat treated to obtain strength levels of 210,000 psi ultimate and 180,000 psi yield strengths; all bars tested were of these strengths ±2000-3000 psi.

Elongation and reduction of areas were dependent on the molding material employed and on the location in the casting. It was found that:

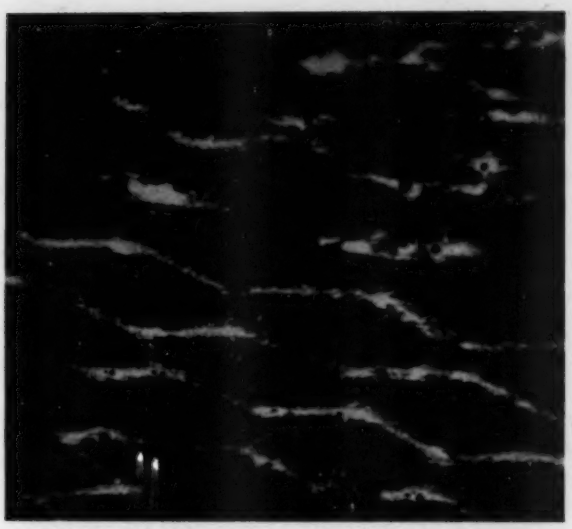


Fig. 16 — Microstructure of a solidified weld deposit of 4330 alloy (modified). Etched. 500 ×.

a) Optimum properties were observed in the end chilled plates, near the chill.
b) Properties of the end chilled plates decreased to a minimum at between about 3 and 4 in. from the chill and then increased again at 5 in. from the chill (under the riser).
c) Properties of the unchilled plates varied along the length of the plate in similar fashion to the above, but were generally lower.

The lowest elongation and reduction of area properties in each plate were found in the region where microporosity was observed to be greatest, and this microporosity was concluded to be a major factor in contributing to the low properties. It is apparent,

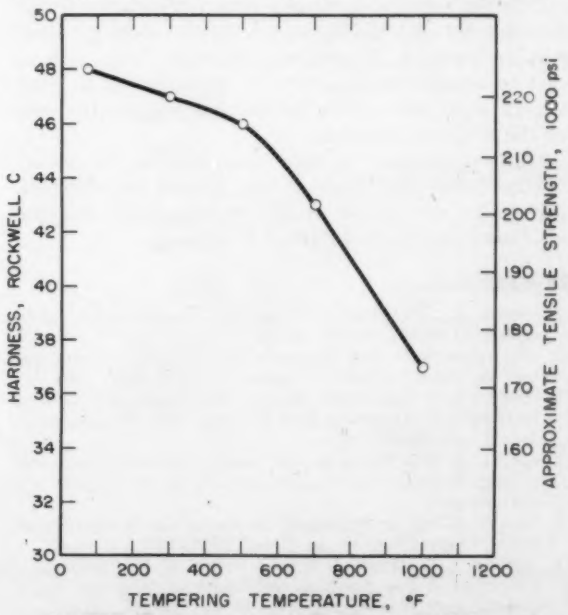


Fig. 17 — Tempering curve for 4330 (modified) weld bead.

however, that other factors, particularly microsegregation and inclusion distribution, may be important in determining properties in the plates. For example, elongation and reduction of area at the location only 1 in. from the plate end are appreciably higher in the chilled plates than in the unchilled plates.

At this location, no microporosity was observed in any of the plates, and the improvement in the chilled plates is apparently due to reduced microsegregation or to improved distribution of inclusions resulting from the fast cooling.

Conclusions above are based on study of high purity material (less than 0.01 per cent sulfur and phosphorus). One plate cast with higher sulfur (0.017 per cent S) exhibited substantially lower elongation and reduction of area properties. Inclusions were larger than with the low sulfur metal and were also of unfavorable elongated (Type 2) distribution.

Impact properties of bars taken from the cast plates showed no correlation with location in the plate or with solidification variables. One reason for this is thought to be the few tests performed.

The effect of ultra-fast cooling on the structure of cast low alloy steel was examined briefly, by arc melting on a solid, cold plate. A dendritic structure many times finer than that produced in sand castings resulted. At high strength levels the structure so produced possessed good ductility after only a tempering heat treatment. Homogenization should completely eliminate microsegregation in the fine dendritic structure without use of excessively high temperature.

Thus, material cast by arc welding should prove useful in determining what beneficial effects might result from complete elimination of microsegregation in cast low alloy steels.

ACKNOWLEDGMENTS

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REFERENCES

1. Walker, J., "Solidification of Ingots," *Symposium on Solidification of Metals*, A.S.M., 1957.
2. Alexander, B. M. and Rhines, F. N., "Dendritic Crystallization of Alloys," *A.I.M.E. Transactions, 188*, 1267, 1950.
3. Harper, J. F. and Stein, H. J., "Microscope as a Control Instrument in Annealing Steel Castings," AFA TRANSACTIONS, *vol. 32*, pp. 679-689, 1924.
4. Lorens, A. W., "Problems and Practices in Heat Treatment of Steel Castings," *A.S.T.M. Procedures, vol. 32*, Part II, pp. 253-268, 1932.
5. Sims, C. E., "Heat Treatment for Grain Size in Cast Steel," *Metal Progress, vol. 26*, no. 3, pp. 22-27, 1934.
6. Wallace, J. F., Savage, J. H. and Taylor, H. F., "Mechanical Properties of Cast Steels as Influenced by Mass and Segregation," AFS TRANSACTIONS, *vol. 59*, pp. 223-243, 1951.
7. Aktutay, M. T. and Rosenthal, P. C., "Homogenization of Cast Steels," *Foundry, vol. 79*, No. 3, p. 100, March, 1951.
8. Briggs, C. W. and Gezelius, R. A., "The Effect of Mass Upon the Mechanical Properties of Cast Steel," *A.S.M. Transactions, vol. 26*, pp. 367-385, 1938.
9. Bishop, H. F. and Fritz, R. E., "Segregation in Small Steel Castings," AFA TRANSACTIONS, *vol. 55*, pp. 412-426, 1947.
10. Heine, R. W. and Rosenthal, P. C., *Principles of Metal Casting*, McGraw-Hill, New York, 1955.
11. Briggs, C. W., *The Metallurgy of Steel Castings*, McGraw-Hill, New York, 1946.
12. S.F.S.A. Research Report No. 15, "The Influence of Microstructure on the Mechanical Properties of Cast Steels."
13. Holloman, J. H. and Jaffee, L. D., *Ferrous Metallurgical Design*, John Wiley and Sons, 1947.
14. Passmore, E. M., Flemings, M. C. and Taylor, H. F., "Fundamental Studies on Effects of Solution Treatment, Iron Content, and Chilling of Sand Cast Aluminum-Copper Alloys," AFS TRANSACTIONS, *vol. 66*, p. 96, 1958.
15. Merten, W. J., "High Temperature Treatments of Castings and Forgings from Heavy Sections," *A.S.S.T. Transactions, vol. 13*, pp. 1-21, 1928.
16. Bull, R. A., "Purchase Requirements for Steel Castings with Notes on Physical Properties in Test Bars and in Commercial Castings," *A.S.T.M. Procedures, vol. 32*, Part II, pp. 77-105, 1932.
17. AFA Committee, Report of Steel Div. Committee on Test Coupons, AFA TRANSACTIONS, *vol. 45*, pp. 97-105, 1937.
18. Kura, J. G. and Rosenthal, P. C., "Cast Steel Homogenization Heat Treatments," AFA TRANSACTIONS, *vol. 54*, p. 154, 1946.
19. Marcotte, R. J. and Eddy, C. T., "The Effect of Homogenization on Cast Steels," *A.S.M. Transactions, vol. 40*, 1948.
20. S.F.S.A. Research Report No. 42, "The Effect of Heat Treatment Variables on the Toughness of Cast Steels," Ordnance Dept. Contract DA 33-019-ORD-1510 through Watertown Arsenal.
21. American Brake Shoe Co., "Development of Low Alloy Steel Compositions Suitable for High Strength Steel Castings," Progress Reports 1, 2, 3, 4, W.A.D.C. Materials Laboratory Contract No. AF33(616)-5299; Project No. 7-8-7 351, 1958.
22. Flemings, M. C. and Taylor, H. F., "Adapting Theory to Practice in the Manufacture of Light Metal Castings," Exchange Paper, 25th International Foundry Congress, Brussels. 1958.
23. Pellini, "Factors Which Determine Riser Design and Feeding Range," AFS TRANSACTIONS, *vol. 61*, p. 61, 1953.
24. Walther, W. P., Adams, C. M. and Taylor, H. F., "Mechanism for Pore Formation in Solidifying Metals," AFS TRANSACTIONS, *vol. 64*, pp. 658-664, 1956.
25. Jackson, W. J., "Some Features of the Metallurgical Properties of Castings," *British Foundryman, vol. 50*, p. 211, April 1957.
26. Sims, C. E. and Dahl, F. B., "The Effect of Aluminum on the Properties of Medium Carbon Cast Steel," AFA TRANSACTIONS, *vol. 46*, pp. 65, 1938.
27. Sims, C. E. and Briggs, C. W., "The Effect of Various Deoxidizers on Cast Steel," to be presented at the 25th International Foundry Congress, Brussels.
28. Sims, C. E., "A Review of Metallurgical Practices and Their Effects," *A.I.M.E. Electric Furnace Steel Proceedings, 10*, 152-171, 1952.
29. "Survey of Low-Alloy Aircraft Steel Heat Treated to High Strength Levels," Parts 1-6, W.A.D.C. Technical Report 53-24, 1954.
30. Nekervis, R. J., Lund, C. H. and Hall, A. M., "Status of High Strength Steels for the Aircraft Industry," Battelle Memorial Institute, TML Report No. 91, p. 24, Jan. 3, 1958.
31. Hodgson, S. K., "Chance Vought Uses More Steel Castings," MODERN CASTINGS, Jan. 1958, p. 26.
32. Tyler, J. W. A. and Ahearn, P. J., "Investigation of Heat Treatments and Mechanical Properties of High Strength Steel Castings," Watertown Arsenal, Rodman Laboratory, Report No. RPL 11/5, Aug., 1957.
33. Foundry Section, Metals Processing Division, Department of Metallurgy, M.I.T., "Investigation of Solidification of High Strength Steel Castings Under Simulated Production Conditions," Final Report. Dept. of the Army, Ordnance Corps, Contract No. DA-19-020-505-ORD-4511, July, 1958.
34. D'Andrea, M. M., "Fusion Zone Properties of Arc-Welded Modified 4330 Steel," S.B. Thesis, Dept. of Metallurgy, M.I.T., 1958.

CORE BOX RIGGING FOR HIGH PRODUCTION

By W. H. Miller

ABSTRACT

Most foundries have their own formula for determining the size of core boxes and driers. This formula is dictated by available facilities such as core blowers, core ovens, core racks, etc. With this in mind this paper includes a discussion of those items of rigging that have proved successful at the author's company in helping to develop a relatively high production standard with minimum core box maintenance required. The majority of the items to be discussed may be unique at this foundry, but can be used by many foundries regardless of their facilities.

INTRODUCTION

The definition of high production would vary depending upon individual experience. The writer feels that high production exists when one man can operate an interchangeable blow machine at the rate of at least 160 blows/hr while making an average size core. Of course much higher production can be obtained from a special one job machine.

Because the term "core box rigging" can be construed to include the entire design and manufacture of a set of core boxes, the writer feels it necessary to pre-determine the limits of this discussion.

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Only oil sand, blow type, steel faced, magnesium core boxes will be considered, although some of the items to follow are applicable in other situations. Upper half core box will mean the top core box or the core box half which contains the blow holes. The lower half core box will be the bottom.

The type of lower half core box to be discussed is not unusual, but the upper half is of special design. Figure 1 is a cut away sketch of an upper half core box. Note that the box is made in two pieces—1) the main body, which contains the core cavity, and 2) the top magnesium plate, which is used to hold the blow tubes. The top plate also contains the upper half core box mounting holes.

PLAN FACILITIES

Probably the most important requirement in a high production core room is interchangeability, especially where different kinds of cores are required in large volumes. This type of core room has to be well planned particularly when obtaining blow and draw machines. The machines need not be of the same make, but they should have the same dimensional characteristics such as daylight and core box capacity sizes, minimum and maximum. The machines

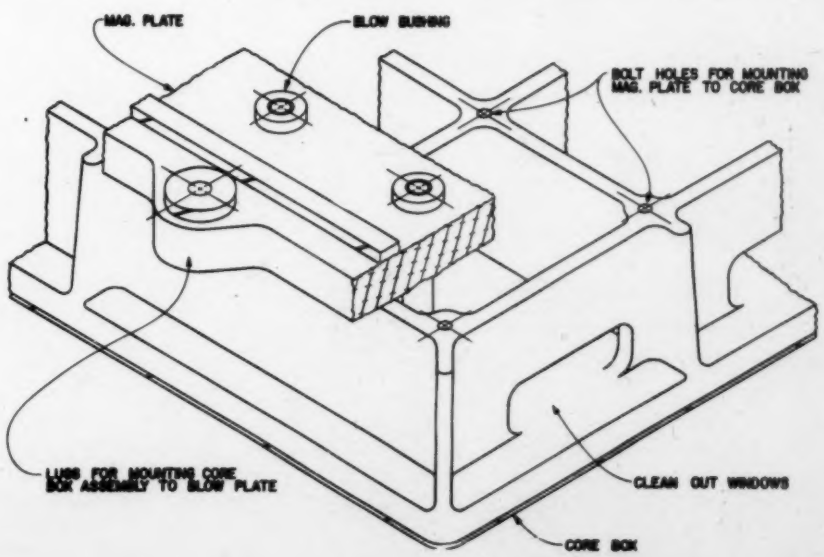


Fig. 1 — Cut away section of upper half core box.

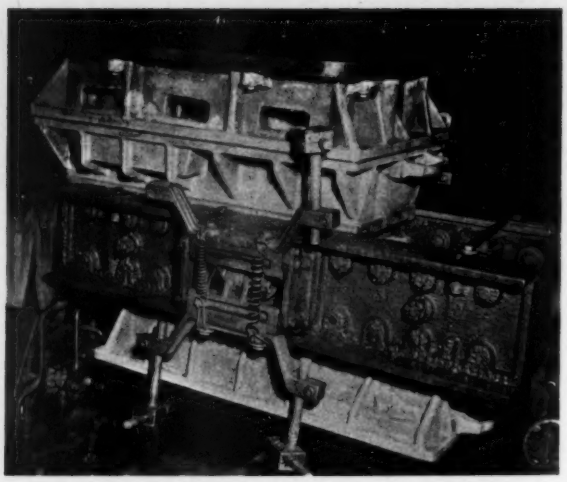


Fig. 2 — Typical set of core boxes being put on a machine setup.

should also have identical core box guiding, locating and holding systems.

With a core room thus equipped, one set of core boxes can be removed from a machine setup, and another set, used to make an entirely different core, can be installed during a half-hour lunch period. Figure 2 shows a typical set of core boxes being put on a machine setup. The machine and core boxes have been standardized to a point that the only installation work required is to fasten the upper half core box blow plate assembly to the sand reservoir and adjust the core box end stops on the blow and draw machines. The core boxes are made with guide rails and center locator standard so as to fit many machines.

The reservoir is a standard size, 10.00 by 26.00, and is designed to be used with a great number of core boxes. Of course the draw mechanism must be adjusted in most cases because of the variation of core heights. In many cases the physical dimensions of core boxes are identical and no machine adjustments are necessary. With the above setup over 600 core boxes, at the author's company, can be used on the same machines with no major adjustments required.

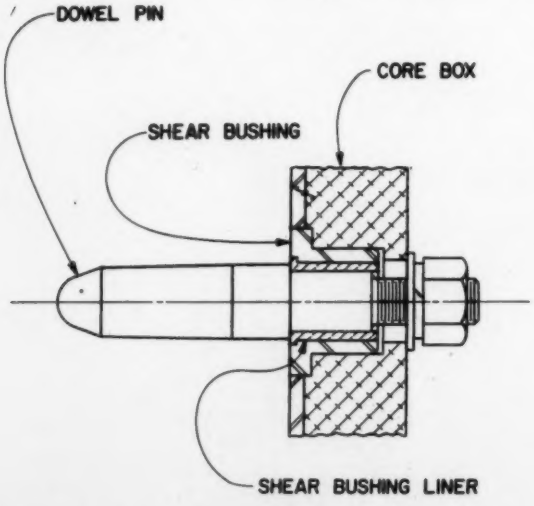


Fig. 3 — Shear bushing assembly.

CORE BOX CLEANING WINDOWS

It is extremely important to furnish high production core room supervision core boxes that can be used hour after hour with a minimum of trouble in making quality cores. A properly screened core box with a liberal number of well placed blow holes will generally suffice. However, such a core box must include certain other features to sustain a high rate of production. For example, sand fines will escape through vent screens while a core is being blown. Some of these fines will lodge in the screen passages, and eventually accumulate to such an extent that the passages become plugged and not usable.

To overcome this undesirable condition, windows have been cast in the sides of the upper half core boxes. (Fig. 1). An air hose nozzle can be inserted through the window and used to blow the screen passages clean. Core boxes without cast windows must be removed from the machine for cleaning after 4 hr of use, while the boxes with windows can be used steadily for 16 hr.

CORE BOX SHEAR BUSHING

The core box pin can create a problem. If a machine functions out of cycle and starts to close the lower half core box with the upper half core box before proper centering has taken place, the core box pin ear will be broken. The prevent core box damage in such an instance a shear bushing assembly has been developed and installed in all core boxes. Figure 3 shows this assembly. A steel flanged type bushing is installed in the core box first. Next a brass bushing is installed. You will note that this bushing has a small flange and shoulder. Finally, the core box pin is installed.

In the case of a machine operating out of cycle, as noted above, the only damage, in most cases, is to the brass bushing. This brass bushing can be replaced in a short time while the core box is still on the machine. This greatly reduces machine downtime and core box repair costs.

BLOW TUBES

Many types of blow tubes have been used in core rooms throughout the foundry industry, and many opinions have been expressed for and against these various types. In the writer's opinion, the most trouble free and long-lasting type is shown in Fig. 4. This blow tube is an assembly consisting of three parts, 1) the blow bushing, 2) blow bushing liner and 3) blow bushing insert. The blow bushing is made of cold rolled steel and is turned with a 0.25 in. thick flange. This flange is used as a spacer between the upper half core box and blow plate, and also as a seal at the blow plate. The inside diameter of the blow bushing is made to accept the liner, slip fit and the insert, press fit.

A plain piece of standard stainless steel tubing, ground on the outside diameter, is used for the blow bushing liner. The liner is held in place by

melting a small amount of solder between the liner and the blow bushing. Stainless steel is used because of its durability. The blow bushing insert is pressed into the bottom of the blow bushing. You will note the back angle of the inside of the blow bushing insert (Fig. 4). This back angle is used to separate the sand within the blow tube from the blown core during the draw operation.

As is the case with all surfaces in a core box, the blow tube is subject to wear and must be maintained. The liner should be replaced before it gets worn through, and the blow bushing insert should be replaced at the first sign of a sand stool on the drawn core.

MAINTAINING CORE BOXES

A core box rigged for use in high production must be durable. Continuous blowing of sand under pressure can cut a core box in a short time. The areas most susceptible to this cutting action are generally directly below the blow hole, vertical side walls adjacent to the blow hole tip and restricted passages within the closed core box. With this in mind, finite attention should be given to these areas when preparing a core box for production use.

The writer does not know of a "miracle material" that will withstand sand abrasion and still meet all of the other core box material requirements such as strength, machineability, satisfactory initial cost, etc. Consequently, steps should be taken to insert steel or copper change pieces in a new core box. With this type of core box construction, close dimensional control can be maintained by changing inserts as often as necessary. The insert principle increases the efficiency of a core box since the amount of time necessary to keep the core box out of production for repairs is greatly reduced.

Figure 5 shows a new type wear pad that is installed whenever possible directly below the blow hole in the lower half core box. A hardened steel washer is inserted in the core box and a large flat head screw is threaded in place. This simple assembly forms a backdraft hole. When the first core is blown the hole is filled with sand, and since it is packed in a backdraft area the sand will not draw out when the core is removed. The sand that remains in the core box becomes an effective wear pad as subsequent cores are made.

Of course, the washer and screw will wear in time and must be replaced. This wear pad assembly should not be used indiscriminately since it will mark the core slightly.

VENTING

Core box venting is generally accomplished by installing vent screens. The size and mesh of the screens, as well as their location, is controversial and therefore will not be discussed. Much more can be accomplished by pointing out other methods of venting that can be used in core boxes along with vent screens. Figure 6 shows a good example of the use of sheet screening. In this particular example the core is too small to install the smallest standard vent screen and still have room for the required vent pin. The sheet screening is backed by a steel plate in which small holes spaced around the periphery have been drilled.

Care must be exercised when choosing the hole size since the sheet screen will buckle into too large a hole while in use. This type of venting is extremely effective and can be used extensively.

Another method of obtaining core box venting is by use of clearance around core box insert pieces. For example, when inserting a barrel in a cylinder block water jacket core box, the insert can be made about

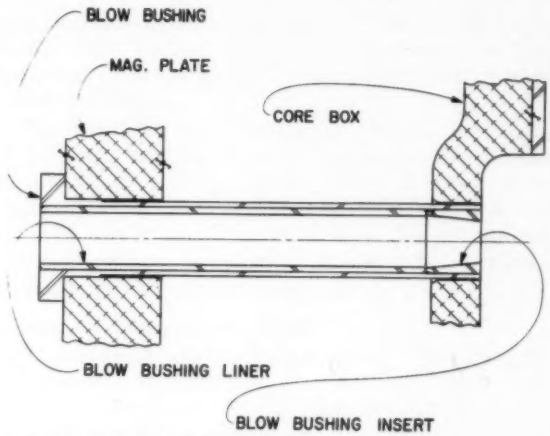


Fig. 4 — Blow bushing assembly.

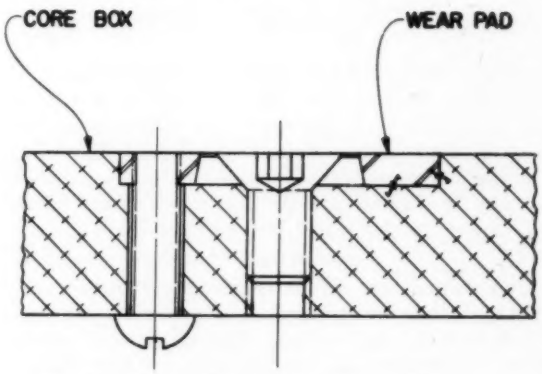


Fig. 5 — Wear pad installation.

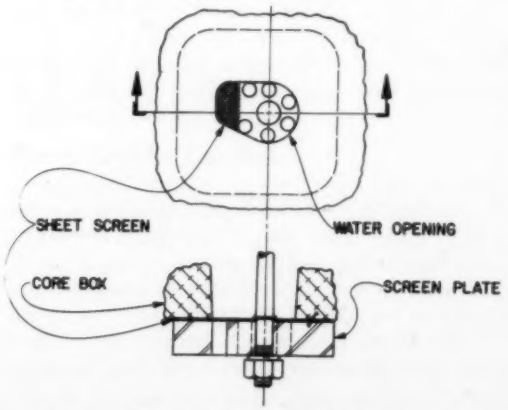


Fig. 6 — Screen plate installation.

0.015 in. smaller in diameter than the core box. By drilling a series of small holes in the core box beneath the insert piece, a type of vent is completed.

If a small hole appears in a core on a surface that is adjacent to an internal parting, the probable cause is trapped air. By cutting a 0.25 in. wide x 0.007 in. deep channel into the parting and terminating this channel at a screen, also installed in the parting, this problem will generally be solved.

All of the items included in this discussion are now being used at the author's company, and possibly have been put into use in other foundries. In each instance a definite cost savings or quality improvement has been established. The foundry industry has made rapid strides with the advent of high production core blowing and certainly within a few years the methods of rigging as described above will probably be obsolete.

FOUNDRY DESIGNING FOR STEEL CASTINGS

By Alfred B. Steck

ABSTRACT

The foundryman is forever blaming castings defects on design. He is at best a third guesser in eliminating defects after the casting has been made. More effort should be placed on foundry designing for steel casting. With this in mind the latest techniques in controlling hot tear defects are explained.

INTRODUCTION

Why is it that in a production run of castings there are found good castings and bad ones? If design is blamed for bad ones, what is blamed for the good ones? Certainly, if there are good ones among many bad ones, foundry techniques are not consistently good and, therefore, design should not be blamed.

It remains to determine what variables are present in foundry processes; to design processes with a minimum amount of variables; and to keep these variables under control.

The theme of this paper is aimed at developing a systematic approach towards foundry designing for steel casting. The nature of the defect should be determined, the cause found and removed and the process redesigned so that the defect will not reoccur. The systematic approach can not be over emphasized because too often the foundryman pushes the panic button and changes all variables at once.

Perhaps he gets out of trouble and maybe not, but the fact remains that if he does get out of trouble, he does not know why. Actually he is still in trouble because when the defect reappears, he does not know which variables to change to correct the defects.

A casting and its manufacture must be looked upon as being made up of materials and the reaction of these materials to high temperatures. With this in mind let us look at the so-called design defects encountered in the casting of steel, and how the modern foundryman designs his process to eliminate these defects. The design defects of main concern are those called hot tears.

HOT TEARS

For all practical purposes it may be said that hot tears occur in the vicinity of the solidus of metal systems. This means that if the steel casting is not allowed to contract freely during the liquid-film stage of solidification, a hot tear will occur. The first step is to identify the defect. The hot tear is recognized as a jagged crack which is discolored due to oxidation of the crack surface. On close examination the surface of the crack appears to have parted when liquid metal was present.

The next step is to determine what is preventing the casting from freely contracting during the liquid-film stage by looking at that material which is at right angles to the plane of the hot tear. If the material is core or molding sand, then the strength of the sand mixture must be weak when the steel goes through its liquid-film stage.

Chemical Composition Effect

It must be remembered that as the chemical composition of the steel changes, the liquid-film stage changes. A sand mixture designed specifically to prevent hot tearing in an 0.60 per cent carbon steel may not be satisfactory for an 0.20 per cent carbon steel. The 0.60 carbon steel will tear at about 2550 F whereas the 0.20 carbon steel will tear at about 2675 F.

Also the liquid-film stage of a basic steel is different from that of an acid steel so that some casting configurations will tear more readily with acid steels than with basic steels, and vice versa. This change in film stage can be related to the greater amounts of sulfur, phosphorus and oxygen present in the acid steel. These elements lower the solidus temperature and this changes the liquid-film stage.

It should be remembered also that hot strength of sand mixtures is a time dependent reaction. That is to say, the hot strength of dry sand mixtures heated rapidly will be equal to the baked strength as tested in the laboratory. This is assuming that the laboratory specimen is rammed to the same density as the sand in the mold or core. This means that on thin sections of castings subjected to hot tears, sand mixtures must be designed to have low baked strength rather than low hot strength.

Heavy Section Castings

On heavy sections prone to hot tearing, sand mixtures must be designed for low hot strength. For example, in designing a core for a $\frac{3}{8}$-in. wall casting which is prone to hot tearing, one should design the sand mixture so that it will have to have extremely low baked strength—of about 200 psi compression to prevent hot tearing. Large cores surrounded by thin metal sections of necessity will be too weak to handle.

It is well to coat these cores with a thin layer of wash to give the cores more physical handling

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strength. Another scheme which helps is that of making the drag portion of the core stronger than the cope portion, and in this way the entire core is easier to handle.

If it is found that hot tearing still occurs when the core and mold material is as weak as practical at hot tearing temperatures, some means should be found for changing the time of hot tearing. Thus, hot tearing will be a function of the hot strength of sands rather than a function of the baked strength. It is much simpler to alter hot strength properties than baked properties because cores must have a minimum baked strength for handling, whereas hot strength requires no minimum or maximum for handling.

High Temperature Pouring

One way of changing the time of hot tearing on a thin wall casting is to pour the casting at a higher temperature. If the core has been designed for low hot strength at the tearing temperatures, no hot tears should result. This is because the high pouring temperature will have burned out the bond which created the baked strength. If the core had been underbaked, the heat of the molten metal will continue to bake the core, the baked strength will increase rather than decrease and more hot tears will result.

Many foundrymen believe that if a sand mixture has good shakeout or low retained strength properties, few hot tears will result. This is false thinking as exemplified by the cold setting binders. Even though good shakeout properties result, hot tears in thin sections may occur, because the baked strength of the cores made of cold setting binders can be high. Also there is a tendency to underbake these cold setting binders with the result that strength increases from the heat of the poured metal.

Thin Section Addition

Another way of changing the time of hot tearing so that baked strength is not a factor is that of adding false padding to the thin section making it appear like a thicker section. This involves casting extra metal parallel and close to the section that is tearing. The sand between the casting and the false padding will be heated to high temperatures, and thus make the thin section appear like a thick section. Care should be used to see that the section which is made thicker does not require additional feeding.

If the transition from one section to another on the casting is drastic and hot tearing occurs, it is well to strengthen the section junction with ribbing which should be removed prior to shipment.

CRACKING

Another type of defect commonly called a hot tear but really is not, is cracking which emanates from shrinkage cavities. The cracking or tearing normally does not break through the casting surface, but can be picked up readily by magnetic particle inspection. The tearing is a result of the stresses occurring during solidification. The obvious solution here is to feed the casting properly or chill the area to strengthen the metal against tearing when the solidification stresses are high.

The so-called bore cracking on cast valves is a result of solidification stresses. In an unfed square section, tears will emanate at right angles to all four sides. As the section changes from a square to a rectangle, those tears perpendicular to the long sides gradually disappear.

CONCLUSION

If hot tears come and go in particular castings, the mold and core materials causing the hindered contraction should be designed to be weak when the molten steel passes through its liquid-film stage. Every effort should be made to control the process variables causing the come-and-go type of hot tearing before blame is placed on casting design.

FLUORESCENT MAGNETIC PARTICLE METHOD FOR CASTING INSPECTION

proper interpretation and inspection

By Arthur Lindgren

ABSTRACT

The use of fluorescent magnetic particle in casting inspection is considered as a method of locating defects. The principles of this type of inspection are given, as well as examples of its use. The need for close cooperation between the foundry and the casting buyer is also emphasized.

INTRODUCTION

Two wrongs do not make a right, but under certain circumstances two rights actually make a wrong. Recently, it happened, and since it involves castings it was thought you would be interested in some of the details.

Standing behind an inspector in a production malleable foundry using fluorescent magnetic particle, it was observed that one out of five castings were being set aside as being rejectable. The specifications agreed upon by both buyer and seller called for 100 per cent fluorescent magnetic particle inspection with no indications. By indications we mean a small accumulation of powder particles which form on the surface of the casting directly over a crack-like imperfection. A close examination showed that the majority of these indications were little more than surface irregularities. Only about one in 20 of the castings being set aside had a depth exceeding 0.002 in. In effect, if the specifications had been followed to the letter, quite a few castings would have been slated for remelt.

As a practical matter, many of these "rejected" castings were actually salvaged by wire brush or grinding, and after reinspection were shipped to the customer. This meant additional manufacturing cost. The point is, the foundry had done its part. Since they had accepted this specification along with the order, they did follow through despite the fact that it would eventually cost them and the customer additional dollars.

How about the casting buyer? Was he right? He was. This was not a brand new part to him, it had formerly been a forging. He knew the stresses involved. In overall design, it was a critical part. However, he knew he could get the strength he needed in a casting and that the cost to him would be less, but it had to be crack-free. What better way out for him than to insist upon a 100 per cent inspection with fluorescent magnetic particle, allowing no indications.

Here is a situation where two groups are both in the right, and yet both are destined to lose in the long run. Even though the buyer saved money by going to a casting, actually he could have saved more had he said defect rather than indication in his specification. The foundryman knew he could produce this part for less money and pass the savings on, but under these conditions he is unable to do so. Who is to determine what is a defect and what is not? This is the problem. How can it be solved?

WHAT FLUORESCENT MAGNETIC PARTICLE IS

Suppose we start by first seeing just what the fluorescent particle used is, and second determining how it is used. At one time or another, you have all made use of a small horseshoe magnet for picking up nails, paper clips or metallic dust. If you dip the two poles of the magnet down into iron filings, you know that the particles will readily adhere to it. The magnetic (flux) field travels through the magnet and passes from the North to the South Pole. These little flux lines would a thousand times rather travel through steel than they would through air, and once started they must continue to flow.

As they reach the North Pole, they jump out of the magnet and return to the South Pole by the easiest possible route. Remember, they would a thousand times rather travel through steel than they would through air. The nail you see hanging to the magnet in Fig. 1 furnishes the easiest possible route of return.

If the poles of the magnet are bent around until they almost touch the magnetic flux line jumps from one pole to the other by the easiest possible route. Going a step further the two poles can be fused together. Now the magnetic lines are free to travel completely within the magnet, and there is no external evidence that they even exist. If we were to define a magnetic pole on a part, it would be "that area on the part where lines of force are either entering or leaving the part." Figure 1 also shows that the magnet now no longer has poles (lower left view).

One final step can be taken and the magnet can be

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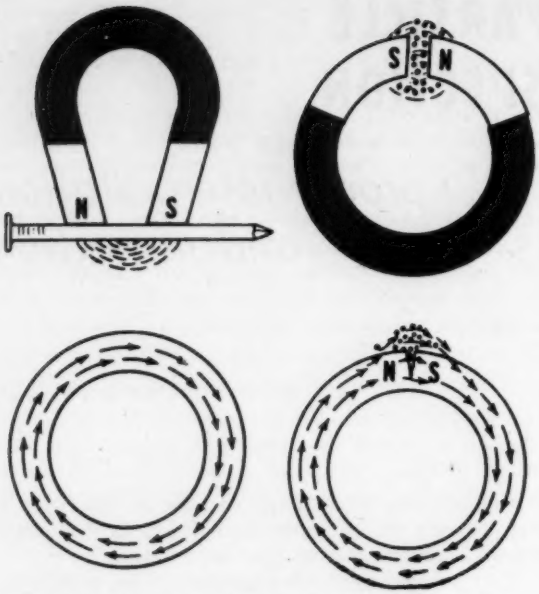


Fig. 1 — Upper left — The nail offers the flux lines the easiest possible return line between the North and South Poles of the magnet. Upper right — Bending the magnet poles so they almost touch the flux lines jump from one pole to the other by the easiest route available. Lower left — Fusing the two poles together allows the magnetic lines to travel completely within the magnet. Lower right — Cracking the magnet so it is not completely severed forces some magnet lines to jump the gap, as North and South Poles have again been created. Some flux lines successfully continue through the unsevered part of the magnet.

broken again. Since it is not completely severed (Fig. 1, lower right), in effect we have a crack in the magnet. Remember, the lines of force much prefer traveling through iron or steel, so as they approach the crack they attempt to by-pass it. Some of the lines successfully pass through the unsevered portion of the magnet, while others are forced out of the part and jump across causing a small leakage magnetic field.

This occurs as long as the lines of force are crossing the crack at a right, or nearly a right, angle. If fine metallic particles are applied to the vicinity of the crack, they will adhere rigidly to it and build a bridge for the flux lines just as the nail did. This bridge of magnetic particles is called an indication, but we should add an indication of what—in this case, it is an indication of a crack.

Basically, this is magnetic particle inspection. To make the method more practical on small size castings, the magnetic particles are manufactured very small and each is coated with a fluorescent dye. They are then suspended in either water or light oil distillate and flowed onto the part at the time it is being magnetized. Should the part be able to hold a magnetic field for a long while, like the magnet Fig. 1, we may magnetize first and apply the particles afterward. A pearlitic malleable casting, for example, will hold a residual field for a long time. On the other hand, a regular malleable casting loses a good portion of its magnetism almost immediately, and, therefore, the particles must be applied at the time the part is being magnetized.

MAGNETIC FIELD

Creating this magnetic field within a part is simple. The current can either be passed through the part, through a central conductor within the part, or it can be passed around it using a coil. The part cannot be magnetized in both directions at the same time. This brings us to an important point. Some castings must be given two inspections in order to find all possible defects. The important rule to remember is to pass the current in the same direction as the defect you wish to find. For example, to find a seam in a bolt, you pass the current from one end to the other. If we were looking for transverse defects up underneath the head of the bolt, we would place it in a coil, thus passing the current in the same direction as the crack.

This is the theory of fluorescent magnetic particle inspection. One of the most typical castings inspected this past year has been the automotive yoke. Suppose there are three places where cracks might appear as shown in Fig. 2. The defects at *A* and *C* are located easily by sending current directly through the part. If we were looking for the defect at *B*, the yoke would be processed in the coil. In each case the current is passing in the same direction as the defect.

The foundryman knows where defects are likely to occur. The user knows where the critical areas are located. Having all this information available, the trained inspector can set up the proper test procedure. A crack at *C* is unlikely and would not be dangerous if it did occur. However, even a small check at *B* could lead to failure under repeated stress, so a coil shot on this yoke is a must. This coil shot will also show any crack of any consequence which breaks over the edge at *A*. By using only a coil shot we will also avoid showing small surface blemishes at *A* which are normally of no consequence, and which would take up valuable inspection time on the evaluation.

For another example, consider a chain drive sprocket casting, as shown in Fig. 3. Any defect in the area of the teeth could be important, particularly if the part is to be heat treated later. Hot checking or shrink around the hub is also important, but only if it is gross in nature. Radial checking is also possible. We, therefore, conclude that a central conductor shot (pass

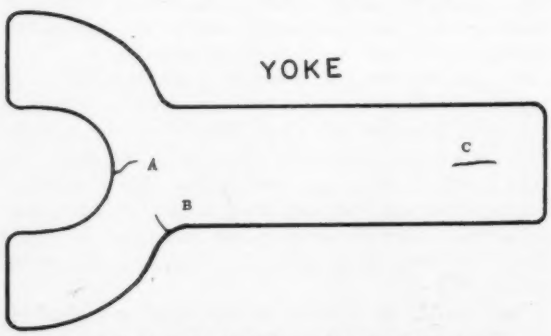


Fig. 2 — One of the typical castings inspected recently is the automotive yoke. *A*, *B* and *C* are places where defects are likely to occur. A defect at *B* is dangerous, but not too likely to occur in the casting process. Defect *C* is not dangerous, and its occurrence also would be rare. Defect *A* is dangerous if it breaks over the edge as shown, and can occur in the casting process. This defect should be watched for closely.

current through a copper bar which has been passed through the center of the hub) will find any possible radial cracks or defects in the teeth area and would also show gross hot checking in the hub area. A pilot run inspection only should suffice. Maintaining 100% inspection here would be wasted dollars.

One more example is a trailer hitch as shown in Fig. 4. If this were made as a fabrication, we may find defects occurring at points *A*, *B* and *C*, and perhaps a few more places. At point *B*, the stress is low and the occurrence of a casting defect unlikely. Should one occur at point *A*, it could be highly critical. At point *C*, if the radius were not made smooth, a hot check could occur. From a use standpoint, this area is always in compression, and by proper contouring of the part the possibility of a crack going over the edge is almost impossible. The answer is a coil shot during the pilot run to watch areas *A* and *C* closely. Once the first run has been successfully made, a percentage inspection should give excellent assurance of quality.

CASTING EXAMPLES

The complicated part shown in Fig. 5 is a good example why many firms have gone to the foundry to find the most economical approach to their requirements. With a casting you can get not only the high strength required for a part, but can have it at much less original cost. Fabricating this part would be difficult. To obtain the most desirable shape using a minimum of material, eliminating all unnecessary machining time, the answer adds up to a casting. Toward the center of the part there is a small surface imperfection.

Normally, this is of little consequence, but at pilot run it suggests slight modification. With one slight change in the contour, however, even this imperfection was eliminated during the pilot run. Future trouble was thereby eliminated. One coil shot during pilot run found this, and would have found anything else of consequence.

Figure 6 shows a similar casting having a defect more major in nature. Here again, the difficulty was caught at once by the foundry's thorough quality control program. A simple coil shot during pilot run was the answer. While this crack might not be detrimental in service, as it is only an aligning bracket, it is the aim of the foundry to eliminate any possibility right at the beginning. In reality, it is now an engineered product.

Figures 7 and 8 represent a similar condition in another casting. This part replaces a fabrication at a considerable savings to the casting buyer. A complex shape was resolved into an attractive product. Nothing was given up in either strength or performance. Even a minor flaw, as seen in Fig. 7, found by a coil shot, was eliminated by a small change in contour. In Fig. 8, another check was eliminated simply by adding the smooth radius you see at approximately the center of the casting.

With all the advantages to be gained, not the least of which is economic, by converting some parts to castings, there should be some way for the buyer and seller to get together on common ground. It is to their mutual advantage to produce the part at a minimum cost. The link between them must be the specification.

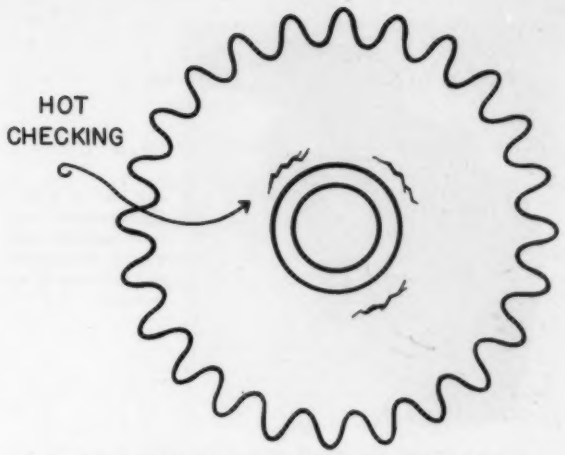


Fig. 3 — A chain drive sprocket casting showing possible hot checking around the hub. If good control is not maintained, discontinuities such as these are likely to occur.

CASTING SPECIFICATION

Logically, the only one who best knows the ultimate function of the part is the casting buyer. He is the only one who knows the location of all critical areas and also those not so critical. Unless he has an experienced foundryman on his staff, however, he should make as much use of the foundry's engineers as possible.

Given the necessary information, the foundryman

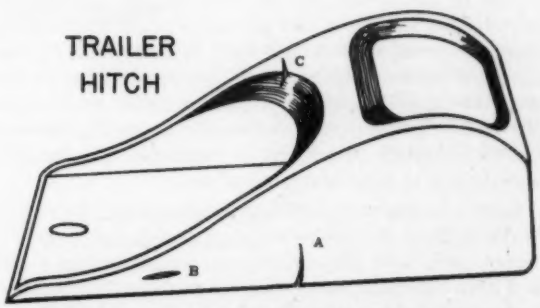


Fig. 4 — Defects can be seen in this trailer hitch as points *A*, *B* and *C*. At point *B* the stress is low and the occurrence of a casting defect unlikely. A defect at point *A*, if it should occur, could be highly critical. If the radius at point *C* were not smooth a hot check could occur.



Fig. 5 — Minor imperfection on casting shown with fluorescent magnetic particle inspection.



Fig. 6 — Large fluorescent magnetic particle indication on casting, but this one is not serious.

should be able to translate it into an easily castable product. Only the foundryman knows best where the natural parting line should be, how to eliminate as many cores as possible, how to avoid the use of cracking strips and, finally, how to reduce weight. No product is perfect. The more we try to approach perfection without good reason, the more costly that part becomes.

Attempts at perfection adds to the foundryman's cost, the manufacturer's cost and the user's cost. It detracts from our nation's ability to produce. Standards of quality that are satisfactory for the job to be done will win the economic war with competition from other countries.

Part of the answer, at least, is working together on the original design of the product. Until the original pilot run is made and approved, is there really a reason for a rigid specification? It is true, a tentative cost must be agreed upon and critical areas established, but until that first run has been thoroughly checked by both parties concerned, a tentative specification may do just as well as a detailed one.

Such a procedure would give the foundry an opportunity to find out just what they can do for their customer, and how inspection can actually reduce cost and help maintain this required quality. The foundry customer is interested only in a satisfactory product at the lowest possible cost. The foundry must fill these requirements or it will not have an order. If the foundry must maintain 100 per cent inspection, this is a cost consideration. If, on the other hand, it can use inspection to reduce cost and monitor production, then inspection is a worth-while tool.

Given a free hand, the foundryman may later find that a good quality control, rather than a 100 per cent inspection control on a particular part, will yield an equally good product, and the savings thereby earned can now be shared with his customer. Without this cooperation in the beginning, real trouble can be experienced in trying to pour a part not designed to be cast. At a later time, the foundry might even find that a good sampling inspection of each production run is all that is really necessary to maintain a high quality product.

CASE HISTORIES

Can such a system work? Let us find out by looking first at a small malleable foundry, and second, at a large pearlitic foundry. At the first one, small to large runs of critical castings have become a specialty. After each day's run, the foremen get together and select 20 individual castings that have been poured during that day. One might be a casting in real trouble, where the rejection rate has been high. Another might be the pilot run of a new product. Still another may be a long distance run where they have corrected a contour, eliminating a cracking strip or otherwise made a change they feel will be to the customer's benefit. The night shift fluorescent magnetic particle inspector is instructed to select a random group of about 25 samples of each casting and give them a thorough inspection.

For their morning meeting, the foremen have a complete report on the findings. Nothing was rejected by the inspection that would not have been rejected by a close visual inspection. The sole purpose of this nightly inspection is to help the foremen come up with answers.

With the facts at hand, corrective measures are taken. Except in rare instances, all castings considered are still quite usable. This procedure is to assure that no casting ever goes out of control, by catching minor imperfections before they become serious and doing something about them. Each foreman is called upon



Fig. 7 — Minor check on malleable casting corrected with slight contour change.



Fig. 8 — Smooth radius eliminates possible hot checking on malleable casting.

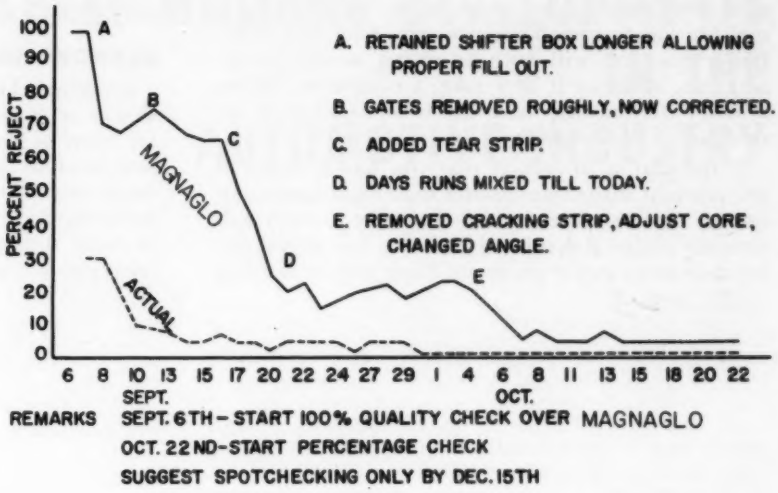


Fig. 9 — Progress report using fluorescent magnetic particle on high volume run of malleable castings.

for his ideas which will eventually mean more profit for the foundry, better designed and lower cost castings for their customers.

Production Foundry Procedure

The production foundry, on the other hand, follows a slightly different fluorescent magnetic particle procedure since they have high volume rather than large variety. More emphasis is given to the best possible design from the beginning. On any new casting that warrants it, a quality control chart is made up for the beginning production run, as shown in Fig. 9.

One particular arm was critical enough to warrant a 100 per cent quality control check over their production fluorescent magnetic particle machine. The first two day's results were not good. Almost every casting had some slight imperfection. Note, however, that after salvage, 70 per cent were still found perfectly usable. After determining the type of defects encountered, the inspection group got together and ran down the major contributing factor. A shifter box was being yanked away sideways too quickly, removing sand from around the gate. This caused the casting to cool too quickly resulting in poor fill out. It was suggested to production that the shifter box be retained a bit longer. This caused a sharp drop in inspection rejects. After salvage, the final rejectable group dropped to 10 per cent.

Over the next few weeks, two more corrections were made. Figure 10 shows their method of providing feed-back information. At point *B* it was found that the gates were being removed too roughly. This situation was corrected. At point *C*, they added a tear strip to eliminate a minor, though possibly important, discontinuity, should it ever get out of control. Note at this point that although the inspection rejects were still at 65 per cent, the actual rejects were now down to almost 5 per cent. Finally, at point *E* the cracking strip added formerly was removed by adjusting the core and changing one angle slightly.

With production now under control, for a full two week period, the 100 per cent fluorescent magnetic particle inspection was replaced by a standard quality control sampling check. Even though the fluorescent magnetic particle with conveyor processing was almost automatic, cost could still be reduced. As shown on this particular chart, if all went well they intended to reduce inspection hours even further by later going to a sampling inspection. Control, rather than a rigid inspection specification is the answer to keeping costs low.

CONCLUSIONS

Next time you are considering making a change to reduce costs, look around for a casting application. It might be a bracket, steering arm, gear, crankshaft, camshaft or rocker arm. Talk it over with the foundry. If it can not be poured economically, they will be the



Fig. 10 — Daily feed-back information in a malleable foundry.

first ones to tell you. Have your engineers work closely with them on the pilot run to make sure your final specification is something you can both live with. Doing this, you will both be saving money in the long run. Whether it be a yoke, a connecting rod or a motor lift bracket, leave the final decision on inspection up to the foundry.

If the part is so critical that you feel it demands 100 per cent inspection despite their rigid inspection system, and there are some of these, the foundry will certainly oblige. But it is only fair that you should pay for these extra inspection hours. Once you are satisfied they are doing the job promised, relax this added requirement.

ACKNOWLEDGMENT

Grateful acknowledgment is given to the Cleveland Works of National Malleable and Steel Casting Co. for many of the photographs and charts used in this presentation. Additional data, help and encouragement were also furnished by Auto Specialties Manufacturing Co., St. Joseph, Mich., and Dalton Foundry, Warsaw, Ind., as well as several other members of the American Foundrymen's Society.

SHELL MOLDING DEVELOPMENTS IN THE AUTOMOTIVE INDUSTRY

By H. C. Grant

ABSTRACT

The details of the shell molding process at the author's company are pictured and described, covering details at the company's three iron foundries. The benefits which they receive from using the shell molding process in the foundries, in the machine shop and to the product engineer are given.

INTRODUCTION

To convert the foundry art into a science has been the long-range goal of learned professors, government agencies, private industry and the American Foundrymen's Society. The advent of shell molding onto the scene will not result in that long dreamed of "book" containing empirical formulas that are panaceas for all casting ills.

At the author's company it is no longer debated whether or not to further expand the shell molding facilities. Instead, the only question asked is which specific shell molding procedure should be adopted to any particular part (from the many adaptations now in production use).

The author's company has three iron foundries and one aluminum foundry in its Engine and Foundry Div. Two iron foundries have partially converted to shell molding, while the specialty foundry has shell molded exclusively since its origin in 1949. The aluminum foundry uses only die casting and permanent mold techniques.

To assure a uniform perspective, the author will first briefly review the shell mold process and broadly survey the general methods of application before specifically describing the different techniques used at the three plants.

SHELL MOLDING

The shell mold process consists of using resin as a binder instead of the conventional oil or cereal binders. The mixture of sand and thermosetting resin is placed on a preheated pattern. The residual heat melts the resin to form a gummy plastic which cements the sand grains together to a depth of $\frac{3}{16}$ to $\frac{1}{4}$ in. This hard crust formed against the pattern is a shell. When stripped from the pattern and assembled with another similar crust, a complete shell mold is formed. Some molds can be poured in this state. Others would rupture or distort from a more severe internal pressure created by the molten metal, and, therefore, require support from a relatively fluid backing material such as shot or sand confined by a flask.

INVESTMENT TECHNIQUES

There are three basic ways to apply this sand resin mixture to the pattern or core box:

1) The dump box method.
2) The strike-off method.
3) The blow method.

With the dump box method a sand resin mixture is dropped onto the heated pattern from an attached dump box. The dwell time (or investment period) is governed by the shell thickness desired, after which the pattern is rotated 180 degrees to drop the excess mix back into the dump box before it unclamps from the pattern. This method is used at the author's company for making our crankshafts, camshafts, gear cases and for the valve job deep-draw mold components.

The strike-off method is an excellent way to make a shallow-draw mold. Again, an excess of sand-resin mix is dropped onto a hot pattern. However, the sand comes from a stationary overhead hopper and the patterns are open faced and do not rotate about their axis. The excess mix is removed by a strike-off bar to produce a flat backed shell or mold section. Two of these sections, when subassembled with integral dowels, form what is called a "book." Several such books are piled up vertically to form a stack mold. Such molds require no backup and are merely clamped or weighted before pouring. Valve rocker arms are cast by this technique at the author's company.

With the blow method, a sand precoated with resin is blown into a pattern or core box cavity. Part or all of the metal surfaces in contact with the sand are preheated. Several variations of this method are used, but all have one common characteristic, i.e., the mold or core is finish formed and precisely dimensioned on all surfaces. Such molds obviously lend themselves to fixturized support. Thus, the company is able to permanently mold large intricate ferrous castings by using the shells as expendable, refractory liners. The company is approaching the semi-die casting of iron and steel.

In addition, there is a wide latitude in selecting from a variety of available sand mixes, namely; dry mixes, cold mixes and hot mixes.

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Dry mixes are simple mixtures of sand with about 5¾ per cent of phenolic resin in dry, powder form. Such mixes are limited to use with the dump box method. They also tend to segregate and create a dust nuisance. The trend now is toward precoated sand.

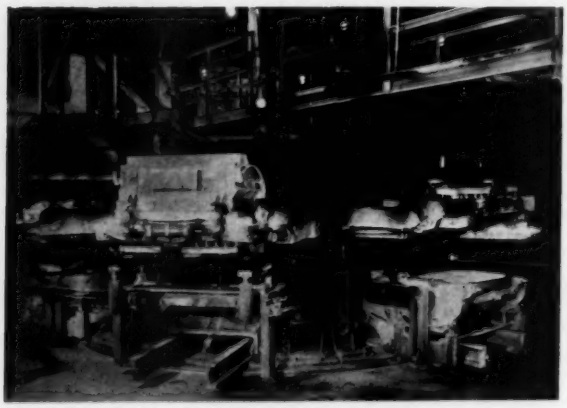
With cold mixes, the resin and a solvent (generally alcohol) are added to the sand at room temperature in a mixer. The alcohol dissolves the resin, causing it to flow freely and uniformly precoat the sand grains.

With hot mixes, the phenolic resin is brought into contact with heated sand in a mixer, where it melts and flow coats the sand grains. Subsequent exposure to the higher pattern temperature melts the resin and it cements the sand grains into a cohesive mass.

Further latitude is obtained by using resin for precoating in the powder, lump or liquid forms.

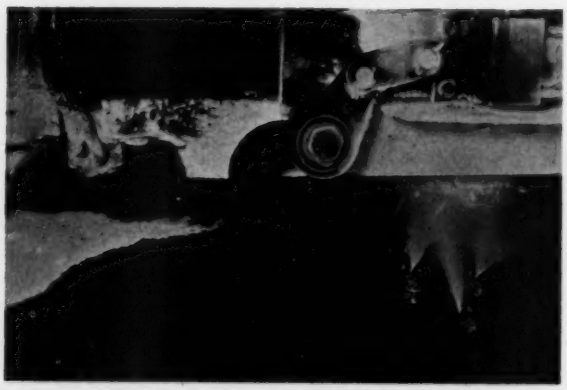
SPECIFIC SHELL MOLD DEVELOPMENTS

This section describes and illustrates the diversified techniques used to shell mold specific castings in the three iron foundries.

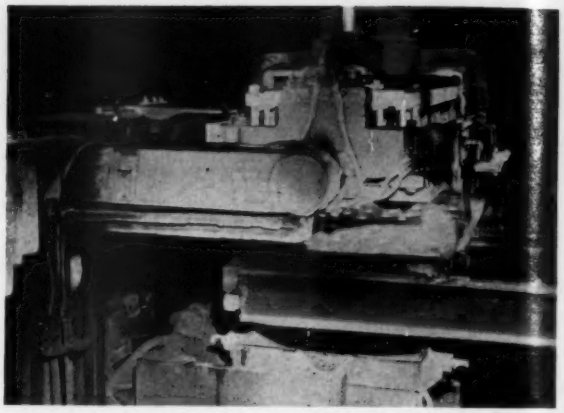


For the ductile iron crankshafts made in the specialty foundry the shells are made on this multi-station machine (shown in overall view), which has six spokes emanating from the hub. A pattern is mounted at the outer end of each spoke. The hub contains the "brain" to actuate the cams which control the series of operations.

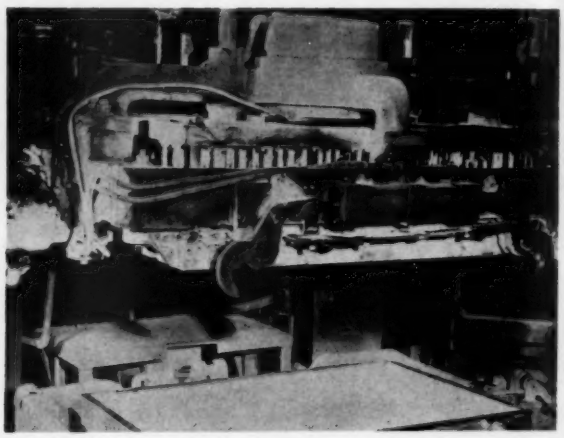
The patterns at the outer rim circle continuously; a complete cycle is made every 100 sec to produce six shells. Each of the patterns is simultaneously subjected to a different operation at the six stations. All operations are automated except the final shell transfer to the delivery conveyor.



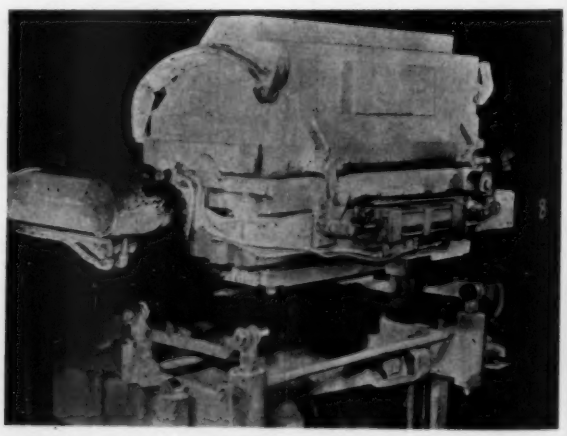
First, the patterns are sprayed with a liquid parting agent. This lubricates the pattern and lessens the shell's resistance to stripping. A 5¾ per cent water emulsion of silicone is used, constantly replenished and agitated to maintain a uniform concentration.



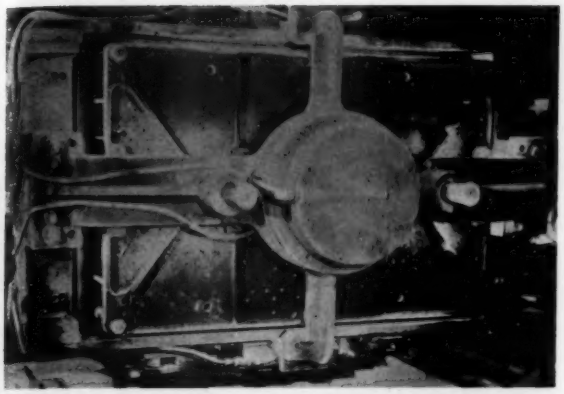
Second, the pattern is preheated over a multitude of low-pressure, open flame gas jets. The heat input is adjusted to exactly balance the combined heat loss at the other five stations.



At the third or investment station, the dump box clamps to the face down pattern, which then inverts on its trunion mounts. The pattern is shown in position for clamping.



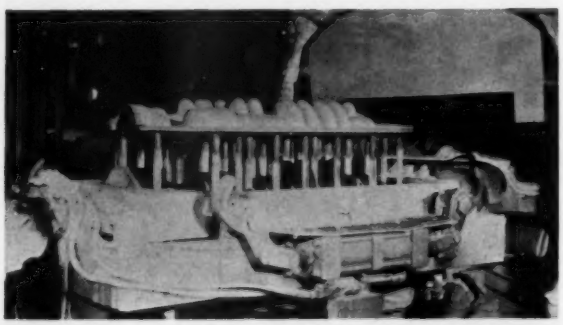
Gravity causes the sand-resin mix to dump on the pattern face. During the subsequent 8 to 10 sec dwell time (or investment period), the residual heat in the pattern melts the resin causing it to flow and smear the sand grains to a depth of 3/16-in. The gummy cement imparts sufficient adhesiveness to assure retention of the soft shell to the pattern during the subsequent rollover.



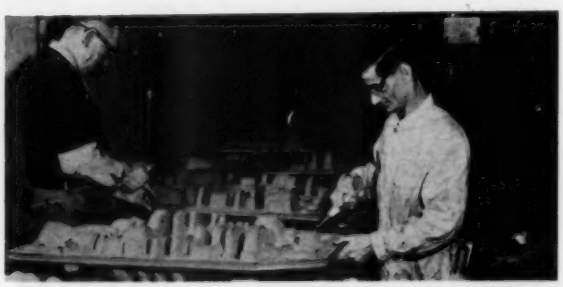
The rollover at station 4 returns the uncemented excess of sand-resin mix to the dump box. Also, the box is unclamped and the pattern inverted again for entry into the curing oven.



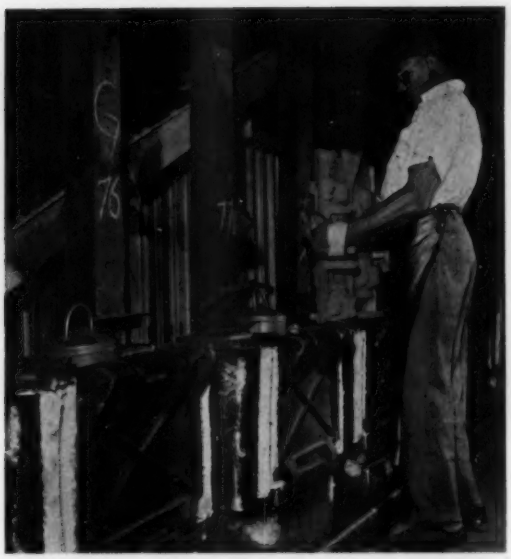
At the fifth station, the shell (while still in intimate contact with the pattern) is finish-cured into a hard crust within a gas-fired, refractory-lined radiant heater.



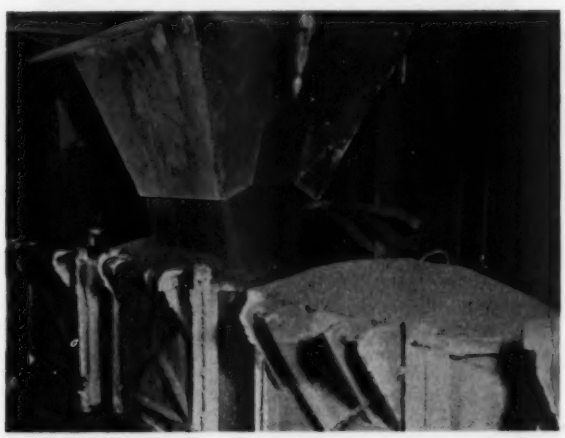
At the sixth and last station, the mechanical stripping mechanism integral within each pattern ejects the shell and positions it for manual grasp and transfer to the delivery conveyor. The ejected shell is shown on stripper pins.



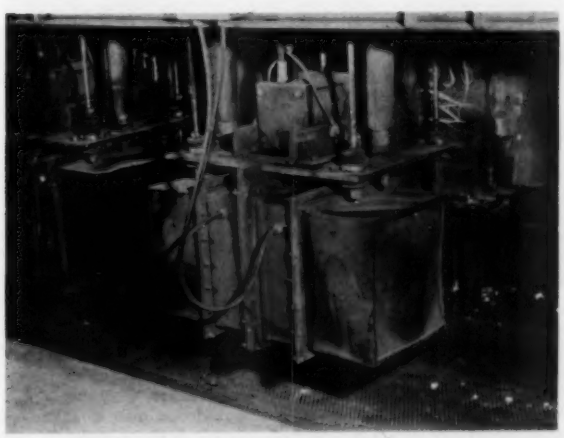
Each shell contains 2 complementary rather than identical half-crankshaft cavities. Two duplicate shells are brought face-to-face for assembly into a complete mold. Thus, output and storage of copes and drags are perpetually balanced.



After two shells are fastened together with ten spring wire clips, the mold is manually inserted into a cylindrical, boiler plate flask attached to a pouring reel.



The open top, hinge bottom flask serves as a container for backing shot dropped from an overhead, gravity hopper. The shot fills all cavities within the flask to support the thin shell mold against the internal pressure of the liquid metal when poured.



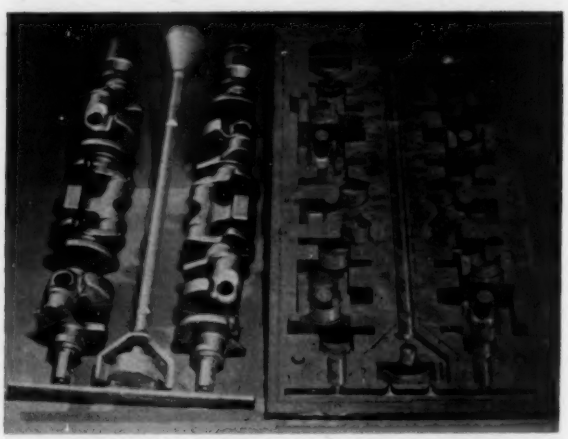
The flasks are automatically grasped by high frequency vibrators for 15 sec to thoroughly pack the backing shot.



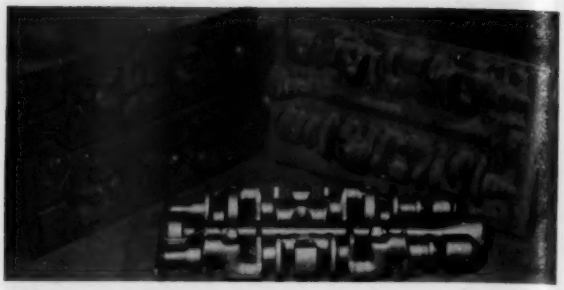
The molds are poured with 1-ton bull ladles.



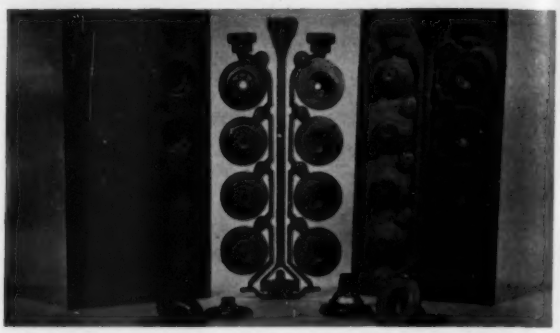
At the automated shakeout, the synchronized cooling line hangers grasp the clusters and convey them to the breakoff station. Thereafter, shot blasting and inspection suffice to ready most crankshafts for transfer to the machine shop.



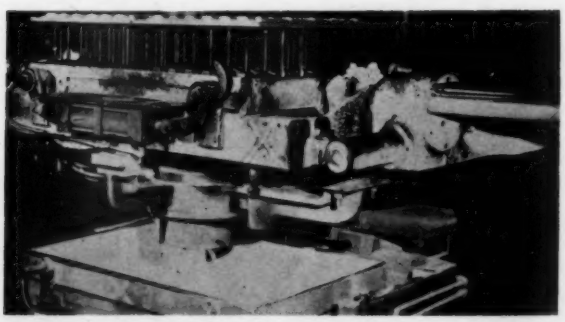
The preceding figure illustrates the shell details and the mating casting details reproduced. Note how the metal is channeled from the pouring cup into the sprue, then through a tangential gate to a slag trap wherein it acquires radial momentum to induce a swirling action that centrifuges out any slag or dross. The cleaned metal drops into the runner, then flows upward into the two casting cavities through the bottom gates with a quiet, non-turbulent flow. Note the interlocking locaters and the anti-dust baffles.



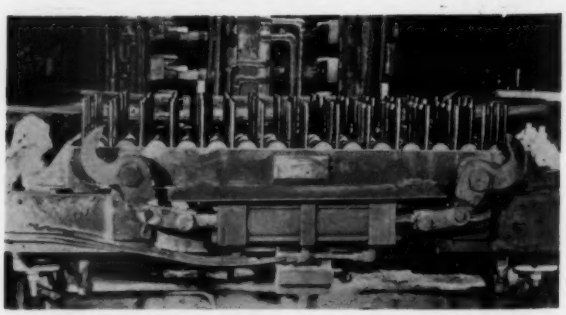
The preceding figure compares the internal and external shell details for a six cylinder crankshaft cluster. Note the integral mold sections used to core out the lightener holes.



The gear cases and cover used to house the rear axle differential gears are molded with a method almost duplicating crankshaft molding. The foregoing figure shows a composite view of gear case shell, shell assembly, cluster and castings.



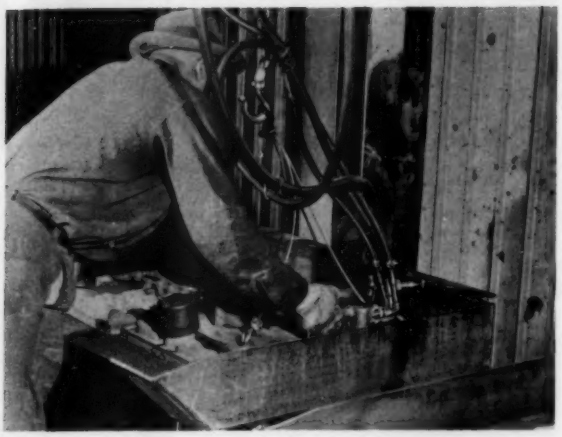
The camshaft mold making operation in the specialty foundry is identical to the crankshaft molding, except that 14-gang patterns are used, with the cope and drag patterns being separate and the molds are poured horizontally. The drag shell is shown on the stripper pins. A second machine simultaneously produces cope shells.



The shell has been removed in the preceding photograph to permit a study of the compact pattern layout which maximizes the castings per mold, and minimizes the material cost per casting.



Two camshaft mold assemblies are laid on a metered cushion of bedding shot positioned in the flask. The open top flasks hang from a pouring reel.



The molds are then vibrated to shape the bedding shot to the mold contours and assure overall support.



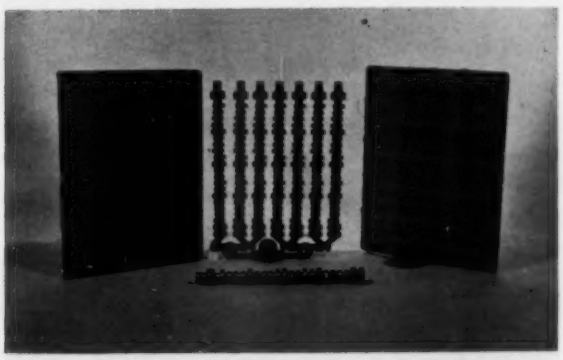
Shot is then automatically metered from a gravity hopper to weight the mold against the internal pressure of the molten metal.



The molds are poured with a shank ladle.

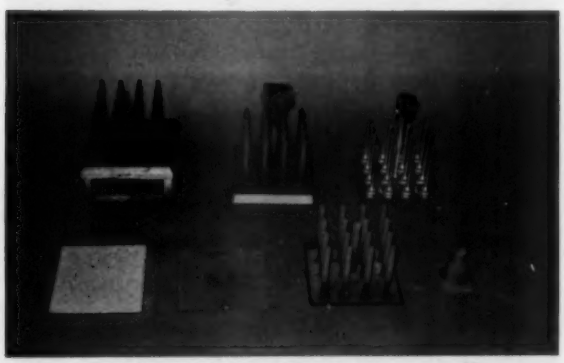


At shakeout, an automatic device pneumatically tilts the flask to eject the clusters onto a grating. The shot drops through the grating for recycling.



This composite view emphasizes the compact shell layout which utilizes every sq in. of pattern surface to form casting and gating cavities, interlocking locaters or stripper pin contact.

Note particularly the excellent definition and the absence of parting line flash.



The exhaust valves were the first part shell molded in the author's company. In the figure you see on the assembly in the background, the four mold components. The light colored runner core in the lower left corner is made by blowing; the ⅛-in. thick gate core is made by the strike-off method; the valve core and runner cup cores are made by the dump box method. In the upper left corner is a valve core pattern; at top center is a complete mold assembly and on the far right the 16-gang casting cluster which it produced.



The runner core is blown in a two-gang core box with plain uncoated sharp sand mixed with a powder resin binder. It is cured in a dielectric oven.



The gate core is made by the strike-off method, which has been found an economical way to make shallow-draw, flat-backed mold sections. The gate core shown, incidentally, is only ⅛-in. thick.

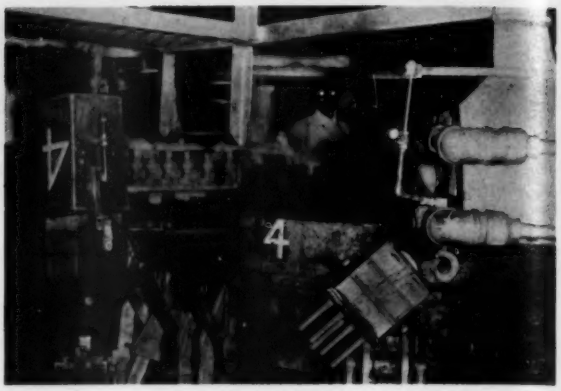


This view shows the gate core after finish-cure in the oven, ready for manual transfer to the delivery conveyor. This is the only manual operation on the 12-station machine.

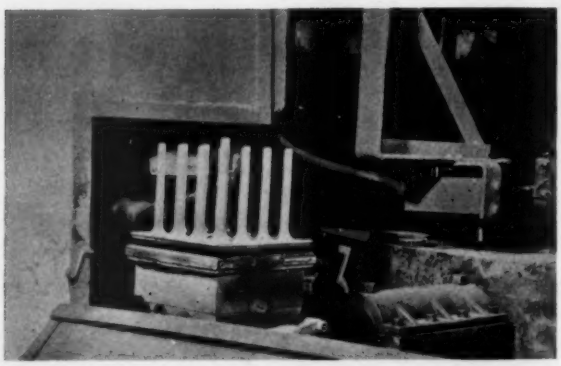


This figure shows a valve pattern cycling through a dip tank containing a 1.5 per cent silicone emulsion. An excellent pattern lubricant is necessary to facilitate draw, since the pattern stems have only 0.003 in. total draft in a 5½ in. length.

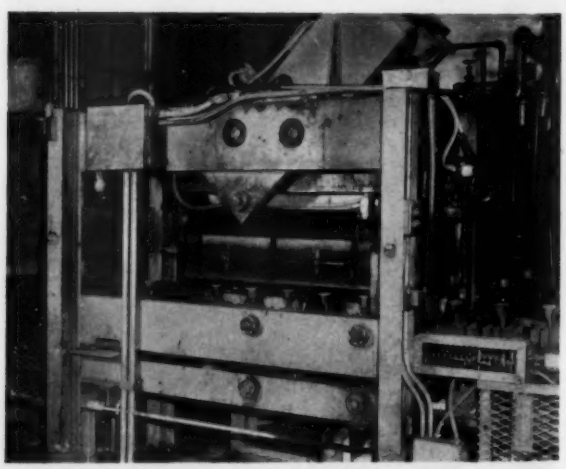
For this mold component, a precoated sand is used, prepared by hot coating with a novolac lump resin. A stearate ingredient added to the mix imparts a self lubricating property to the shell which enables the silicone concentration to be reduced from 5 per cent to 1.5 per cent.



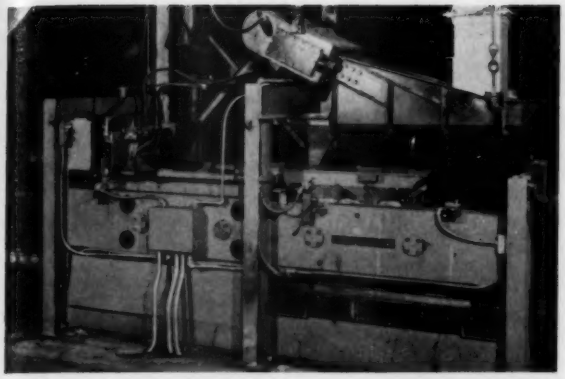
The valve pattern at the left (or no. 4) is being invested on this 15-station shell making machine. At right, the dump box (no. 3) has just unclamped from the pattern after completing investment. The invested pattern is rotating into position for finish cure of the shell in the radiant heat oven at the far right.



The valve shell, on emerging from the oven, has been lifted from the pattern by the integral stripper pin mechanism and is positioned for manual transfer to the delivery conveyor. This is the only manual operation on this 15-station machine.



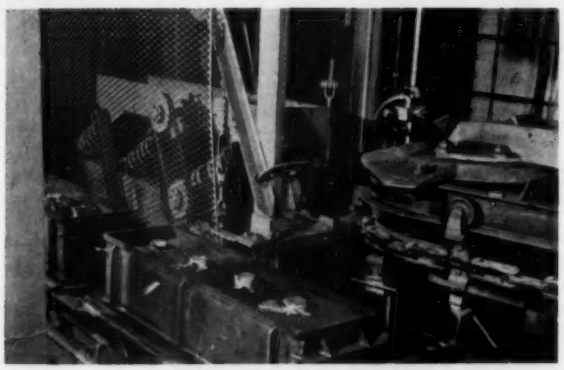
A flask is being automatically positioned around five complete mold assemblies that will produce 80 valves. The forward travel of the flask positioning device is synchronized with the constant mold conveyor travel.



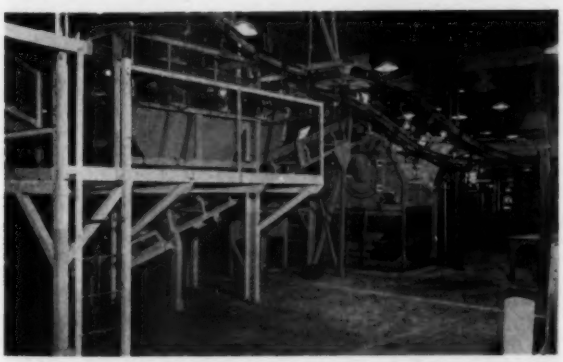
The four mold components in each assembly are self-located with interlocking locators; no fasteners are used. Shot is automatically metered from a gravity hopper to prevent separation of the mold components by the internal pressure of the molten metal.



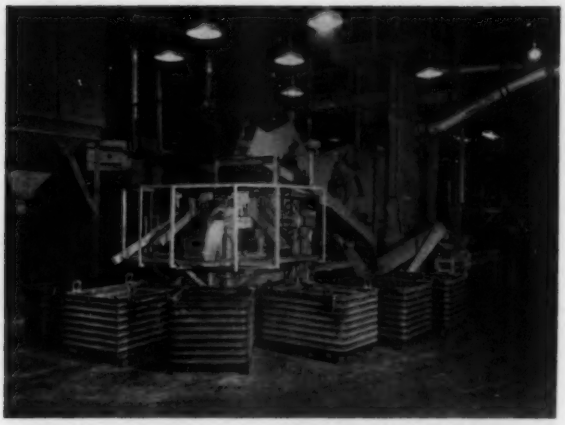
Metal is poured into the valve molds at a temperature of 2940 to 2960 F.



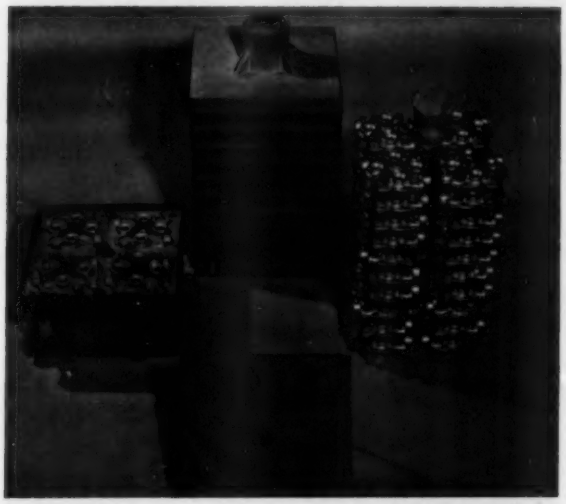
Shakeout of the valve clusters is automatic.



After cooling, the clusters then enter an automatic break-off service which breaks the castings from the gates.



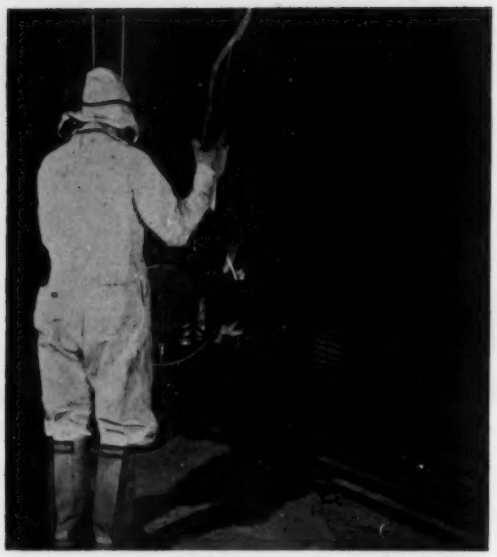
The valves then enter an automatic sorting and inspection machine which routes the various types into a shipping container.



This figure will permit a perspective study of why the strike-off method was chosen to make the rocker arm castings. The shallow-draw, flat-backed mold section in the foreground is made in the pattern at the left. Two such sections are paired to form a sub-assembly called a "book." Ten such books are piled vertically and topped with a pouring cup core to form a stack mold. The complete mold produces the 160 casting cluster seen at the right.

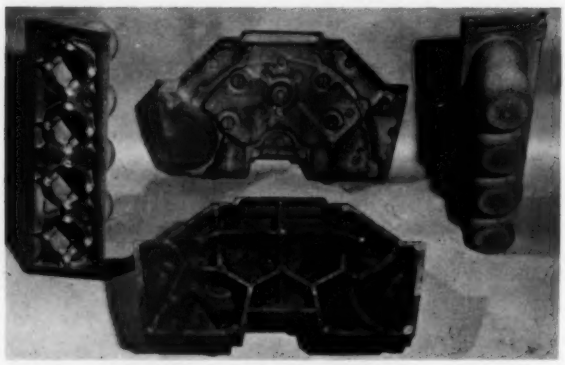


All rocker arm mold sections are made on this 22-station machine. The only manual operation is the transfer of the shell to the delivery conveyor.



The rocker arm molds are poured with a bull ladle synchronized to keep pace with the mold conveyor. Note how the entire shell mold has become red hot before pouring is completed. However, no cracks or mold distortion develop from the thermal shock or the ferrostatic pressure exerted.

The cluster continues on, and after cooling is dumped automatically off the pouring line into a tumbler which performs the dual function of separating the castings from the gates and also cleaning them. They exit from the tumbler onto an inspection and separation belt and then drop into the shipping containers.



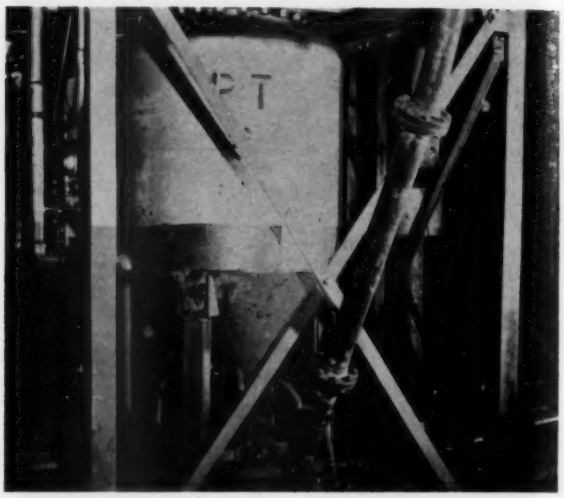
One iron foundry (the largest foundry) is converted to blowing the following hollow shell cores for the cylinder block and the rear axle carrier case:

Barrel cores.
Flywheel end cores.
Gear end cores.
Valve chamber cores.

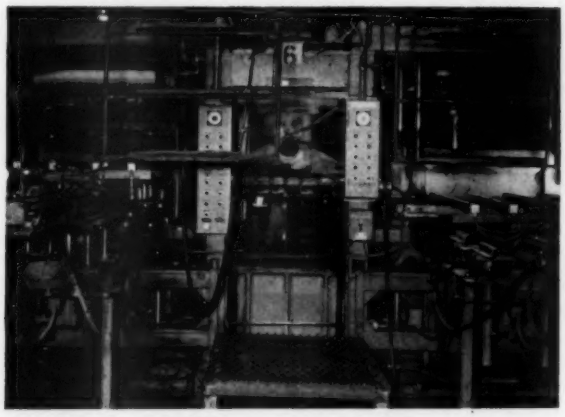


In the foreground of the figure is the hollow shell core used for the rear axle carrier case. Other conversion than shown is in the development stage.

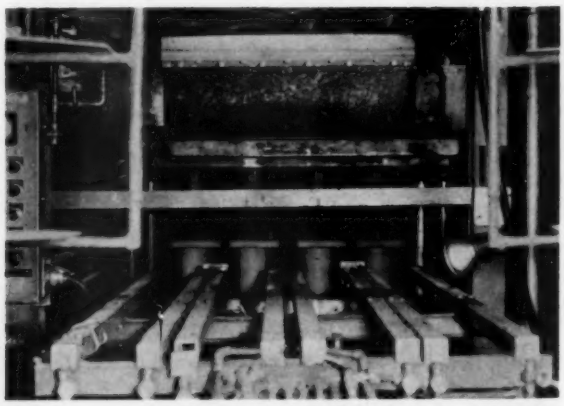
Sharp sand is used exclusively in this foundry. It is precoated by the warm mix method at 200 to 250 F using a liquid resin with an alcohol vehicle.



The precoated sand is pneumatically conveyed in tubes to the hoppers above the top-blow shell making machine.

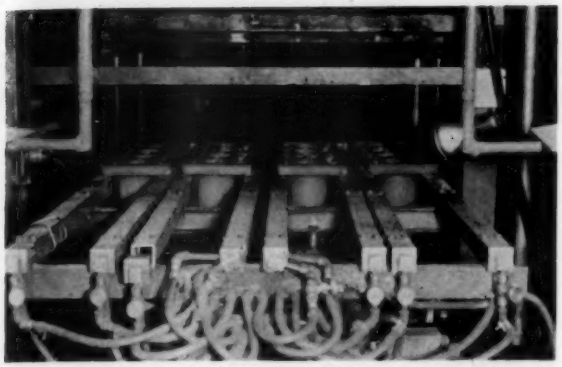


This frontal view shows a two-station blow machine used to make cylinder block shell cores. While an operator on the center platform transfers to the delivery conveyor the four cores ejected from the left-hand station, four additional cores will be blown, cured, stripped, trimmed and ejected at the right-hand station. The cycle is automatic.

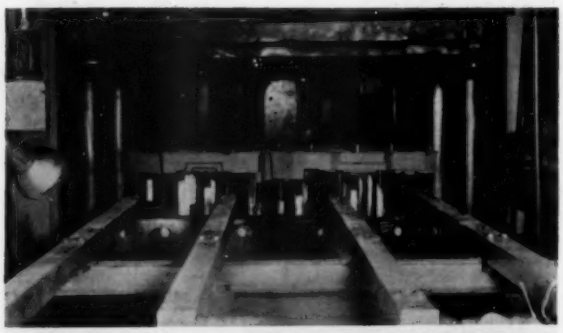


Here you see four barrel cores after blowing, curing and

stripping from the four-gang core box. Note the strike-off bar to automatically knock off the blow tips.

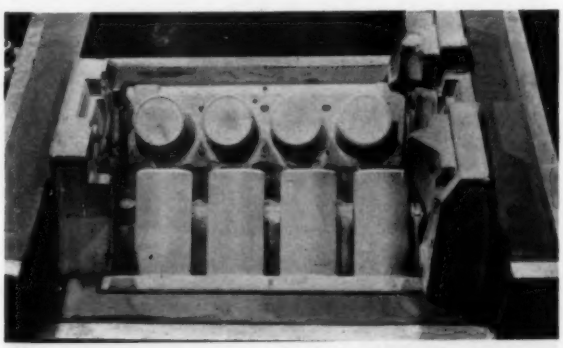


The same barrel cores are shown in this figure after the blow tips have been struck off during automatic transfer onto the ejection ways.



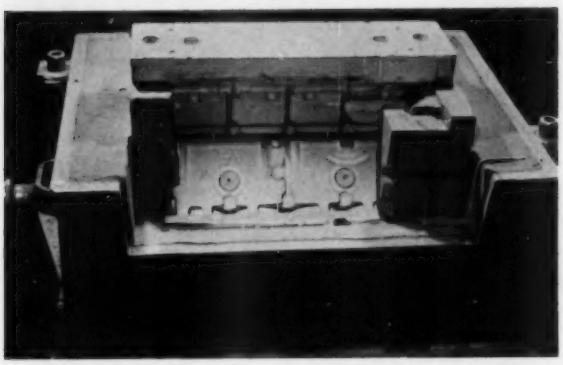


These two repeat views of the previous two figures show the gear end cores being made. They illustrate the versatility of the machine for any core with a change of patterns and a simple adjustment of the ways.

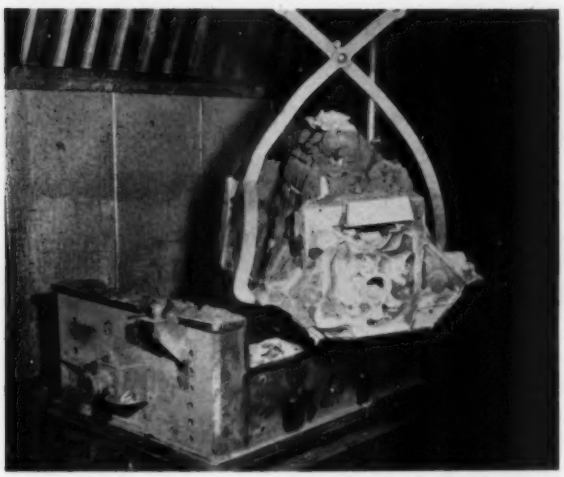


In this figure are a valve chamber core, two barrel cores, the flywheel end core and the gear end core assembled into a green sand drag. Any extended delay in finish assembly or pouring off would not result in any moisture pickup by the cores from the moist green sand.

The shell cores require no wires, arbors or extensive venting; They are hollowed to reduce weight, conserve material, evolve less gas and to facilitate collapse during contraction of the casting. They are ribbed for rigidity and strength. Also, they resist erosion during handling, can be precisely assembled and give good definition. The conversion from oil-cereal cores reduced the barrel core weight from 22 lb to 13 lb, and the flywheel end core from 31 lb to 14 lb.



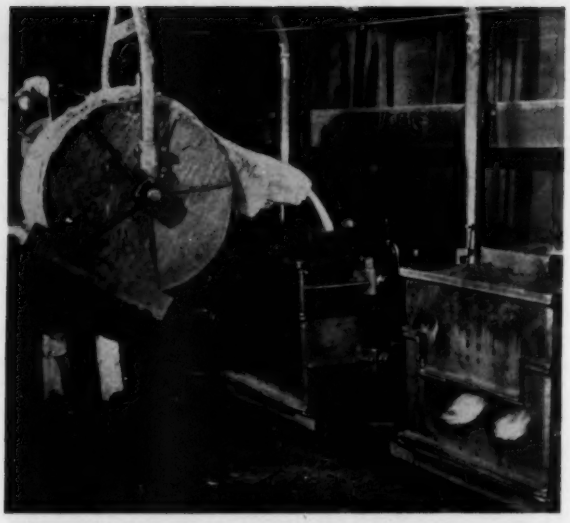
The remaining baked sand cores not yet adapted to the shell technique have been assembled and the cope will be lowered to complete the mold.



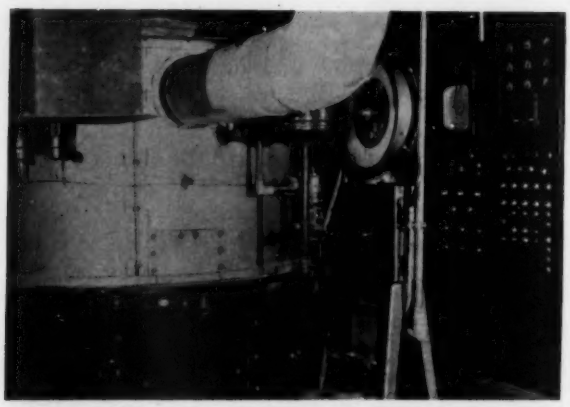
Note on this block at shakeout how the shell cores have disintegrated, and the sand has run out freely from the cored cavities. A core knockout operation is no longer required in the shell core areas.



At the other iron foundry, a substantially similar technique is used to make the cylinder block cores with equally gratifying results.

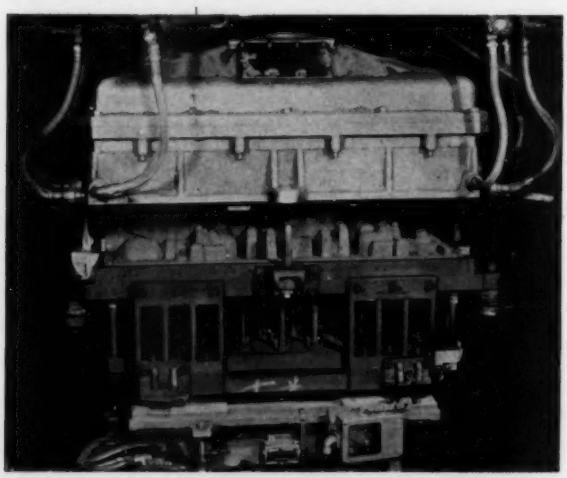


Note the reduced gas evolution resulting from even a partial use of shell cores.



All sand used at this foundry is precoated. The centralized control panel permits instant and precise adjustment of the pre-coating process in the mixer at the left.

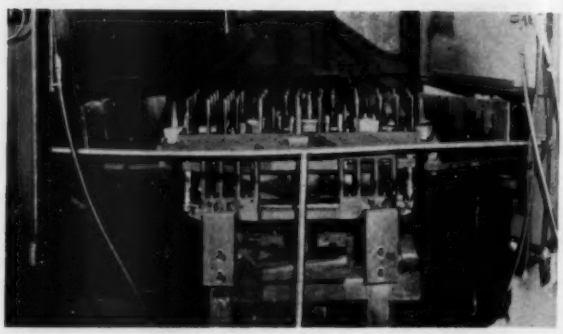
The crankshaft molding technique at this foundry is perhaps the company's most advanced shell mold development, and is at present undergoing even further development. The shells serve as expendable, refractory liners inserted into permanent molds.



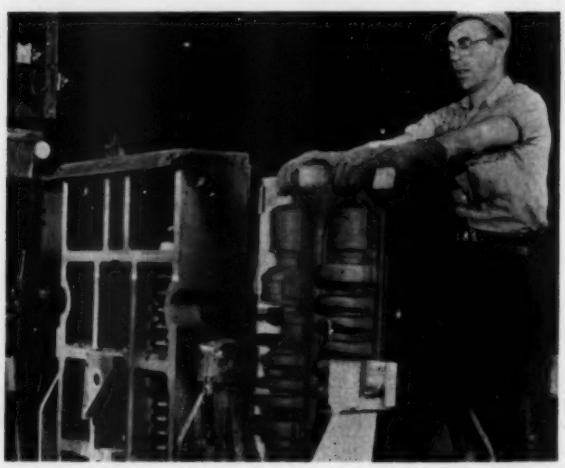
This view of the six-station shell blowing machine shows the blown and partially cured shell on the pattern.



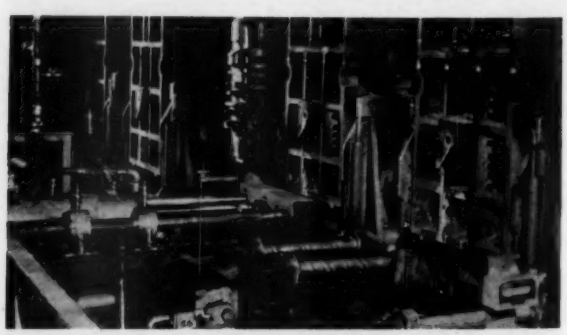
Here the cured shell has been automatically stripped from the pattern by the integral stripper mechanism and will be manually transferred to the delivery conveyor by the lone operator.



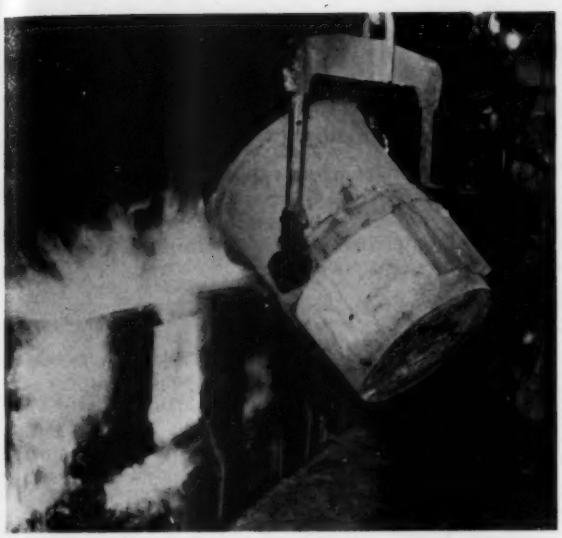
The same station with the shell removed gives a clear view of the stripper pins which are actuated by a cam-controlled stripper plate.



The shell assembly without any fastening is manually inserted into the permanent mold to act as a refractory liner.



This closeup view shows the hydraulic apparatus which automatically closes the dowel-aligned molds.



A bull ladle is used to pour the mold.

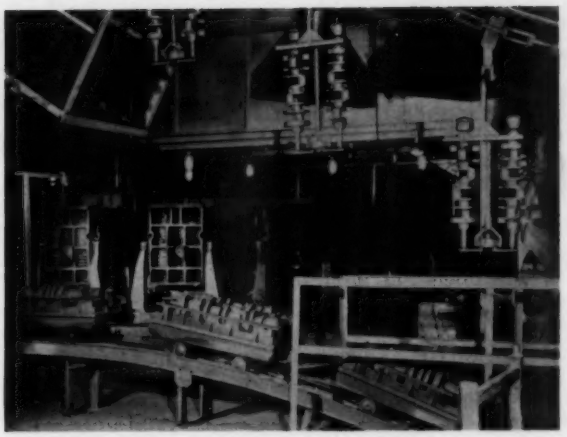


A hydraulic mechanism automatically opens the permanent molds at shakeout.

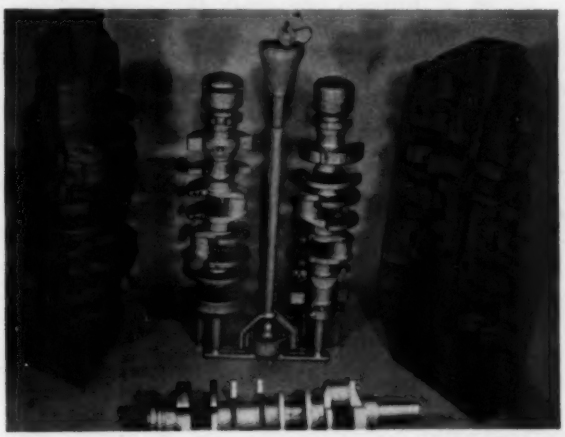


This close-up view shows the opened mold with the crankshaft cluster ready for automatic knockout by a ram. This action also transfers it to the synchronized cooling line which is descending into alignment.

No vibration is required at shakeout thereby reducing the noise level. Also, there is little fuming. Fatigue has been eliminated from what has traditionally been the foundry's most fatiguing operation. The reduction in noise, smoke and dust at shakeout makes the entire foundry healthier and cleaner.



This figure shows three related conveyors — the shell conveyor in the foreground which delivers the liner assemblies to the pouring line, the pouring line itself in the background and the cooling line overhead which transfers the casting clusters to the cleaning room. Note particularly the as-cast surface finish of the castings and the absence of parting line flash. The castings will require no warp check, being uniformly straight. They remain on the hanger for shot blasting, visual inspection and loading into shipping containers.



This composite picture permits a detail study of internal and external shell design, the as-cast cluster and the flogged crank before cleaning. The gate and riser design generally adhere to the principles discussed previously.

BENEFITS FROM SHELL MOLDING

Specific Experience from Conversion of One Part to Shell Molding

The company's specific experience with the crankshaft can best illustrate the benefits gained by the foundry, the machine shop and the product engineer from the conversion to shell molding.

In the Foundry

1) A wider range of pouring temperatures can be tolerated.
2) Stress-relief heat treatment is eliminated.
3) Only 30 per cent of the metal poured need be remelted—a gratifying yield. Also, all the back scrap is so sand-free it needs no cleaning to prevent contamination of the melt.
4) Shell molding permits the use of the latest and most advanced principles of gating and risering. These molding features can be incorporated in-

tegrally in the shells on a cost-free basis, rather than by use of extra mold components.
5) Foundry scrap has historically averaged only a fraction over 1 per cent and foundry machined scrap less than 0.5 per cent.
6) The operations are less fatiguing, less noisy, less smokey, cleaner, safer and healthier. To climax this statement, it can be cited, with excusable pride, the company's world record set recently for accident-free operation by an iron foundry.
7) Much less floor space and equipment is required.
8) The operations are readily automated.

In the Machine Shop

1) Finish stock removed during machining and balancing has been reduced from 9.5 lb on the dry sand molded crankshaft to 5.5 lb on the shell molded casting.
2) The precise counterweight spacing enables elimination of entire batteries of cheeking lathes.
3) Cheek spacing is controlled within $\frac{1}{32}$-in. Also, the counter-weight O.D.'s are cast with such close tolerances that running clearance in the engine is assured without prior machining. To cite a striking example, in one recent model year over a million crankshafts with calculated clearance of 0.0285 in. between one cheek O.D. and the camshaft distributor gear were assembled into engines without a single incident of interference.
4) Tool advance in rapid traverse is maximized because of less finish stock and precisely positioned diameters. Also, the counterweights do not interfere with the lathe tools and grinding wheels.
5) Chip disposal has been reduced 42 per cent.
6) Tool life is extended by the absence of chilled spots, chilled fins and sandy surfaces.

To the Product Engineer:

Functionally, or from the product engineering standpoint, the benefits have been that the designer can provide close running clearance in the engine without injecting extra machining expense and can minimize the weight.

SUMMARY

In summary, shell molding has not only increased the mold material cost but also requires more expensive patterns and molding machines. However, the increased productivity per man, and the reduction in secondary operations, has more than offset these increased costs, especially in view of steadily rising labor rates.

Better and safer working conditions attract and conserve a higher grade of personnel. A more compact layout and less critical process controls further reduce total cost.

Less tangible perhaps, but undeniably potent, are the long-range benefits accruing from consumer satisfaction and design flexibility.

The competition inherent in the free enterprise system will continue to stimulate a series of technological breakthroughs that must inevitably culminate in reaching this goal.

INDIRECT ARC ELECTRIC FURNACE PRODUCTION OF DUCTILE IRON

By L. Miller

ABSTRACT

The general aspects of a method of utilizing the indirect arc rocking type electric furnace for producing high quality ductile iron castings is given. A description of the furnace, its use in ductile iron production, chemistry control and the plunger ladle is presented. A detailed tabulation of the furnace heats over a period of time, the composition and properties of the metal, are brought forth by the author.

INTRODUCTION

The types of melting furnaces that can be adapted to produce ductile iron base metal are, in a broad sense, limited only in their ability to impart sufficient temperature to the melt. However, composition and temperature of the base metal prior to the addition of magnesium are of fundamental importance in the process of producing ductile iron. Therefore, some types of furnaces under certain conditions of operations are more suited to ductile iron production than others.

The probability of two or more foundries operating under identical conditions in every respect is remote, and only by careful study and understanding of the factors that determine the overall manufacturing costs can the individual foundry arrive at a simple, economical method which best suits its particular requirements.

In addition to the ingenuity of management, factors which influence or determine the overall manufacturing cost of ductile iron are the manufacturing facilities, the quality and type of castings produced, the production requirements and, of course, the method of introducing magnesium into the base metal.

In general, a method of utilizing the indirect arc rocking type electric furnace for producing high quality ductile iron castings is covered, and any attempt to review the pros and cons of the various types of melting furnaces, together with the variables that exist from plant to plant, is beyond the scope of this presentation.

QUALITY CONTROL

Ductile iron, in this application, is produced under license agreement with The International Nickel Co. For quality control, a chill test bar and a chemical analysis coupon are cast with each heat. Determinations are made for silicon, carbon, manganese, nickel and magnesium for every heat. Microscopic examination generally is at random, but specific, for heats low in magnesium or otherwise off-color.

The chill test bar is fractured, and by daily comparison gives the first indication of any trend from the normal or the possibility of an off-colored heat. Originally, mechanical properties were determined from keel bars cast with each heat. However, after sufficient data had been accumulated, and because of the high cost of keel bars, these were reduced to one keel bar daily for each type of ductile iron produced unless keel bars are specifically required for certifying a particular heat.

Each heat of castings is piled separately and held until it has been approved chemically and structurally by the laboratory. The gates and risers are then removed and the castings heat treated, after which they are spot checked for Brinell hardness. Any heat of castings suspected of being off-color is heat treated separately. This is followed by a thorough Brinell check on the castings after heat treatment.

APPLICATION

In this application, castings are produced from over 600 production and semi-production patterns. Castings range in section size from $\frac{3}{32}$-in. up to 3 in., with the bulk falling between $\frac{1}{4}$-in. and $\frac{5}{16}$-in. Average weight is about $4\frac{1}{2}$ lb. With few exceptions, castings receive some form of heat treatment.

The yield of good castings for the year 1958 averaged 48.2 per cent, with foundry scrap averaging 5 per cent and machine shop scrap running less than 1 per cent. The yield and scrap figures of themselves are not impressive, and no doubt similar castings of less stringent engineering requirements could be produced more economically with other types of melting units. However, these are small, light section castings the main products of which are safety devices, all of which require high engineering standards of workmanship, inspection and quality of materials.

These high standards insure that the parts function properly in protecting the lives of multitudes of people. Because of this, the significance of the methods and controls employed is apparent.

INDIRECT ARC ROCKING FURNACE

Two 1000AA conical shell furnaces and one 750A cylindrical shell furnace are employed in this application. Combined in a 16 hr operation, they have a

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total cold charge melt capacity of 20 tons. The 1000AA furnaces are equipped with 530 kva transformers and have a 400 kw rate of input. The cold charge melting rate for ductile iron will average 20 lb/min or 1200 lb/hr. Power consumption in a 16 hr operation will average 670 kw/ton. Electrode consumption will run 15 to 17 lb/ton.

The furnaces are lined with an outer lining of second quality fire brick and an inner lining of 92 per cent aluminum oxide special refractory shapes. These linings are neutral or slightly basic and fairly impervious to slag attack.

Linings will average 240 tons melt each. However, prior to the use of return scrap in the charge, lining life in the first of this type lining gave 320 tons melt. The sand is not cleaned from the return scrap, and while there is little slag in the furnace, it is highly acid. In time, at the high working temperatures employed, slag takes its toll on the lining. Silicon carbide door bricks are used and they will average 40 to 50 tons melt per brick.

Five conical furnace shells are rotated between the two 1000AA furnaces for refractory maintenance. A shell is removed from the furnace once a week in an 8 hr operation, or twice a week in a 16 hr operation for major patching with a high alumina ramming mix. Daily minor repairs (mostly around the door opening) are done with a hot patch refractory cement. Refractory repair material will run 15 to 20 lb/ton melt.

FURNACE OPERATION

At the start of a day's operation, after minor refractory repairs have been made, the furnace is preheated by applying 200 kw before the metal is charged into the furnace. These furnaces are equipped with automatic electrode feed and rocking controls. Since no refining or slagging operations are carried out, the operation is simply one of bringing the temperature of the metal up to between 2930 and 2950 F before tapping.

These furnaces perform best and most efficiently when operated at their full rated capacity. Carbon control is generally uniform. However, carbon content of the metal can vary considerably at times, depending on the arc characteristics of the furnace. The arc conditions occurring within the furnace are best described as a normal or neutral arc, midway between a smokey reducing arc and a sharp oxidizing arc.

A normal arc prevails after the heat is under way, the kilowatt input is at full rated capacity, and the electrode feed control maintains a steady arc gap that gives a slight carbon haze or protective reducing atmosphere in the furnace.

High Temperature Oxidizing

While a sharp arc of itself is not oxidizing in a true sense, the atmosphere in the furnace is clear and free of the protective carbon haze. It is while under this condition at high temperature that silicon, manganese and carbon are oxidized from the metal. The degree of refinement in breaking up and dispersing graphitic carbon being carried to the point where the metal is in an oxidized condition. Primary carbide formation is increased, with the result that the subsequent inoculation of ferrosilicon (after the magnesium treatment) cannot break down the carbide completely.

A smokey or reducing arc prevails when the rate of electrode feed provides only a narrow gap between the electrodes. This may be caused by a drop in the line voltage. In extreme cases, the metal is saturated with carbon, and at times silicon is reduced from the slag making it difficult to have all the carbon dispersed and not in the graphitic form. The carbon equivalent moves too far to the right lessening the carbide forming propensity of the magnesium.

Carbon Saturation

In mild cases of carbon saturation, more magnesium will be required to produce spheroidal structures free of flake graphite. However, in heavy, isolated sections black patches of dendritic or internal shrinkage cavities will be evident.

In extreme cases of carbon saturation where the metal, on solidifying, tries to free itself of the excess carbon as in kishing, the fracture of the castings will show a black banded structure in the cope surface. Microscopic examination of this structure will show a much greater proportion of spheroids than the metal beneath the banded area as evidence of the kishing action. On closer examination, these spheroids appear as if they had exploded. Actually, they are breaking down to graphitic carbon.

The chill test bar fractures of the base metal, the treated metal and the final inoculated metal will give indications of variation in carbon and the conditions under which the metal is melted.

CONTROL OF CHEMISTRY

All the metal is cold-charge melted, and since no slagging or refining operations are carried out control of chemistry is comparatively simple. It begins with the selection of materials used in the base metal charge.

Selected intermediate low phosphorus pig iron is used, and every car of pig iron received is analyzed by the laboratory and piled separately. One or two test heats are run to qualify the pig iron and to check the analysis of the charge before regular consumption of the car of pig iron begins. The tentative purchase specification for pig iron which must be low in subversive elements is as follows:

Silicon, %	0.50 to 0.75
Manganese, % max	0.30
Phosphorus, % max	0.04
Sulfur, % max	0.03
Titanium, % max	0.05

The working range in carbon is between 3.70 per cent and 3.90 per cent in the ferritic type, and between 3.60 per cent and 3.80 per cent for the pearlitic-type ductile irons. Variation in carbon occurs by gradual increase or decrease of carbon in the return scrap, and by variation in carbon from car to car of pig iron. These variations are controlled by adjusting the base metal charge.

Tables 1 and 2 show the breakdown of metal charges for ferritic and pearlitic ductile iron.

The third variation in carbon occurs and is controlled within the furnace itself. Accurate carbon determinations are important, and they are only reliable when using a chilled sample.

PLUNGER LADLE

When the metal in the furnace is ready to tap the treating ladle is brought into position in front of the furnace, and a small quantity of metal is tapped in order to check the temperature with an optical pyrometer. If the temperature is 2930 F or over, the entire

charge is tapped into the treating ladle, at which time the lanthanum-cerium alloy is added, and the ladle is skimmed of slag.

The magnesium alloy plunger is brought into position over the treating ladle, and when the temperature of the metal in the treating ladle is 2820 F the plunger containing the magnesium alloy is lowered into the treating ladle (for light sections, 3/32-in. to 3/16-in., the magnesium alloy is plunged at temperatures up to 2880 F). After 45 sec have elapsed, the plunger is withdrawn, the ladle is again skimmed of slag, and the treated metal is split into two teapot pouring ladles with each getting an inoculation of the 85 per cent ferrosilicon.

Factors of importance to the ferrosilicon inoculation are as follows:

1) The ferrosilicon should contain about 0.75 per cent calcium and about 1.25 per cent aluminum. Metallurgical grades of ferrosilicon not containing these elements in appreciable amounts have little inoculating power.
2) The ferrosilicon should be fed gradually into the stream of metal as it is being transferred from the treating ladle to the pouring ladle.
3) The inoculation should be 0.75 per cent or more silicon.
4) The particle size of the ferrosilicon is also considered to be of importance, and in this application the 3/8-in. by 12 mesh size is used.

Treatment Ladle

The treating ladle is an ordinary foundry ladle to which seven or eight inches have been added to its height. The ladle has no pouring lip and is used only for treating the metal. The ladle has an outer lining of split fire brick and an inner lining of a rammed refractory of the type used in mixing or forehearth ladles. A gage is provided the workmen as it is important to maintain the inner dimensions of the ladle. It is also important that both the treating and pouring ladles be preheated to a cherry red heat or hotter while in use.

The plunger hood assembly weighs 3250 lb, or twice the weight of the metal treated. With proper maintenance of the inside ladle dimensions, including the top edge of the ladle and the refractory surface in the plunger hood, the fumes and flare of the magnesium are practically eliminated. The efficiency, measured in terms of magnesium retained, is good when compared with the open-ladle method. With treating temperatures of 2820 to 2850 F, 50 to 55 per cent of the available magnesium in the plunger is retained in the metal.

The metal has less dross and oxide inclusions and is more fluid than metal treated in the open ladle. The result is that castings from the plunger ladle can be poured at lower temperatures before defects such as misruns, pin holes, and slag inclusions occur.

Plunger Cup

The plunger ladle and hood assembly are simple and economical to construct and will last as long as an ordinary foundry ladle. The weak link is the inverted refractory plunger cup. It costs about 16 dollars, and will last for about 30 immersions. The refractory sleeve costs eight dollars and will last from 90 to 120 immersions. The steel rod, washer, spring and nut cost about three dollars and will give about six months of service.

TABLE 1 — FERRITIC DUCTILE IRON CHARGE

	Pounds		Per Cent
Intermediate low phosphorus pig iron	850	or	53.125
Ductile iron return scrap	640	or	40.000
Boiler plate steel scrap	81¾	or	5.110
Alloy (30% lanthanum - 45% cerium)	¼	or	0.015
Calcium bearing 85% ferrosilicon	14	or	0.875
No. 2 alloy (Ni-Mg-Si)	14	or	0.875
Total charge	1600	or	100.0

TABLE 2 — PEARLITIC DUCTILE IRON CHARGE

	Pounds		Per Cent
Intermediate low phosphorus pig iron	830	or	51.875
Ductile iron return scrap	640	or	40.000
Boiler plate steel scrap	81¾	or	5.110
Spiegeleisen (20% Mn)	20	or	1.250
Alloy (30% lanthanum - 45% cerium)	¼	or	0.015
No. 2 alloy (Ni-Mg-Si)	14	or	0.875
Calcium bearing 85% ferrosilicon	14	or	0.875
Total charge	1600	or	100.0

An expendable metal can containing the alloy is used with each heat. Various types of cans and gage thicknesses have been tried, and it was found that with thinner than 24 gage, the reaction takes place before the plunger hood seats itself. When the can is thicker than 24 gage, the reaction is delayed and spontaneously more violent.

At the present time, it is costing us about 67 cents for the plunger and 53 cents for the can; or a plunger cost of a little more than one dollar for each treatment. The author's company realizes this cost is too high and is trying hard to reduce it. Nevertheless, the cost is tolerable in view of the overall savings affected by this method of treatment.

From a treating temperature of 2820 F, the temperature will drop to 2650-2670 F for the first molds cast from the pouring ladles. The temperature drop in the pouring ladles will average 20 degrees/min, and the last molds cast from these 800 lb ladles will be around 2520 to 2540 F.

RETURN SCRAP

Prior to 1958, with the open-ladle method of treatment (because of the buildup of silicon), practically all the ductile iron return scrap was used in the makeup of cupola gray iron charges. For the entire year of 1958, 40 per cent ductile iron return scrap was used in the ductile iron charges. More than 40 per cent returns could be used if necessary, but 40 per cent returns seems to be a stable amount for this practice.

For better chemical control, two return scrap piles are worked. Current gates, risers, etc., are accumulated on one pile, while return scrap for the charges is used from the older or second pile.

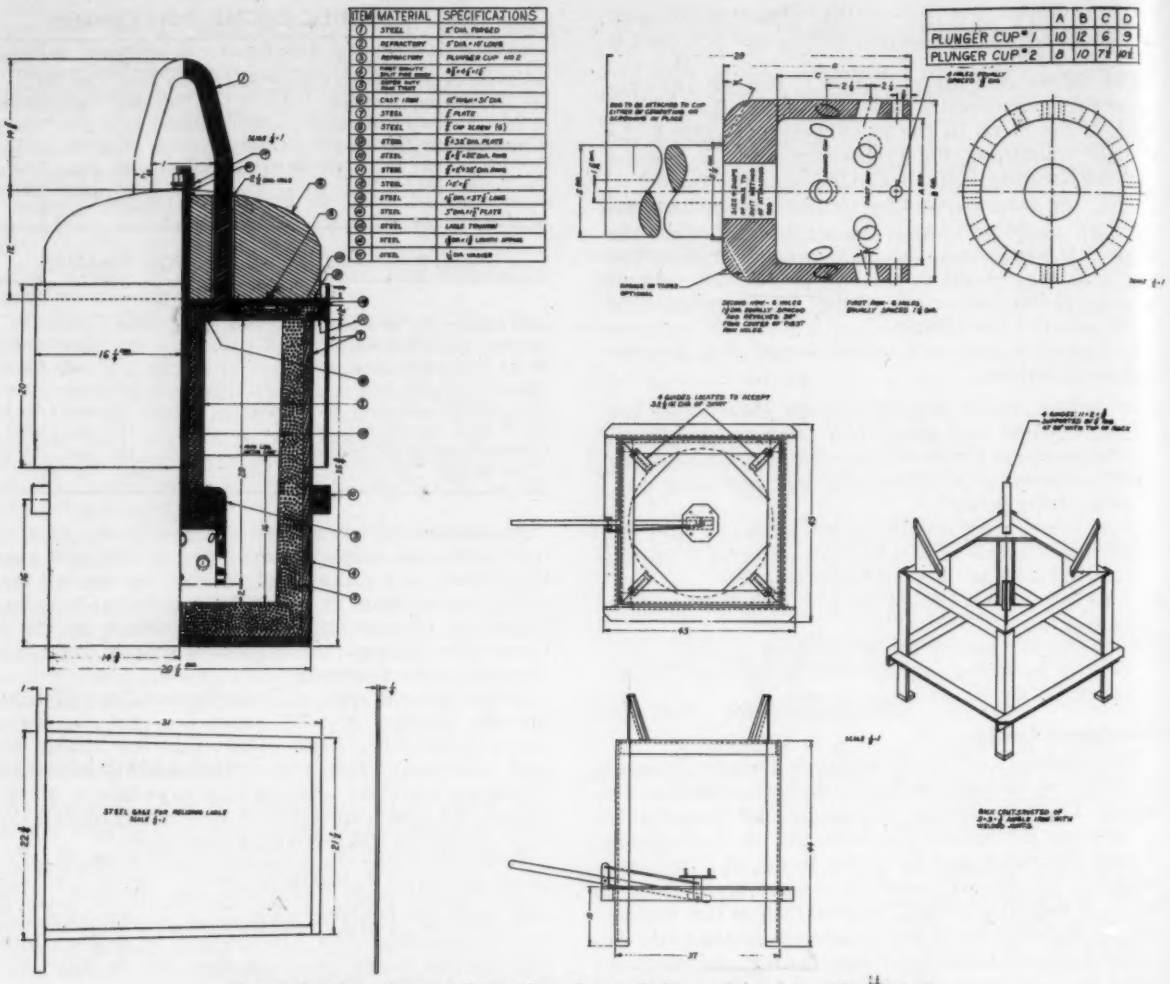
During the year 1958 there were 1751 plunger ladle heats of ductile iron made.

Heats 78 and 745 were rejected due to the presence of flake graphite.

Heat 919 was rejected due to a mixup where the magnesium alloy for a 900 lb heat was used to treat a 1600 lb heat.

Heat 1529, being an experimental one which did not pan out, was rejected.

Table 3 shows the tabulation breakdowns of the 1751 plunger heats in which both chemical and physical tests were taken.



Detail drawing of magnesium alloy plunger ladle used in the production of ductile iron in the indirect arc electric furnace, as described by the author.

TABLE 3 — PLUNGER LADLE HEATS

Date	Heat No.	Composition, % Si	T.C.	Mn	Ni	Mg	Tensile, psi	Yield, psi	Elong., % in 2 in.	Bhn
ANNEALED FERRITIC DUCTILE IRON										
1957										
12-26	20	2.48	3.87	0.26	0.79	0.059	66,000	52,500	21.0	174
12-26	21	2.69	3.75	0.23	0.74	0.054	67,000	53,750	21.0	174
12-26	22	2.46	3.80	0.23	0.76	0.064	65,750	53,125	21.0	170
12-27	23	2.51	3.74	0.23	0.78	0.063	66,250	53,125	21.0	170
12-27	24	2.51	3.77	0.25	0.74	0.059	65,000	51,875	21.0	174
12-30	25	2.48	3.75	0.21	0.73	0.063	63,250	49,375	21.0	163
12-30	26	2.35	3.77	0.22	0.70	0.059	62,750	49,375	22.0	163
12-30	27	2.51	3.70	0.22	0.71	0.063	62,500	48,750	22.0	163
12-31	28	2.69	3.68	0.20	0.75	0.061	66,500	53,750	20.5	170
12-31	29	2.72	—	0.24	0.72	0.051	66,750	53,250	20.5	170
1958										
1-2	30	2.51	3.76	0.22	0.76	0.059	66,750	54,250	21.0	170
1-3	31	2.42	3.74	0.22	0.76	0.064	64,000	45,000	25.0	163
1-3	32	2.45	3.80	0.24	0.73	0.062	64,500	47,000	25.0	163
1-3	33	2.34	3.84	0.22	0.75	0.067	64,750	47,500	25.0	170
1-6	34	2.51	3.80	0.20	0.76	0.055	65,000	50,000	24.0	170
1-6	35	2.45	3.71	0.22	0.72	0.062	64,500	50,000	24.0	166
1-6	36	2.45	3.73	0.20	0.73	0.062	65,000	50,000	25.0	166
1-8	41	2.66	3.83	0.20	0.76	0.061	67,250	51,000	23.0	174
1-8	42	2.54	3.82	0.20	0.77	0.059	65,750	51,000	22.0	170
1-8	43	2.62	3.82	0.20	0.80	0.057	64,250	49,750	23.0	163
1-9	44	2.54	3.84	0.25	0.76	0.065	65,000	49,500	23.0	163

(Continued on next page)

TABLE 3 (continued) — PLUNGER LADLE HEATS

Date	Heat No.	Composition, % Si	T.C.	Mn	Ni	Mg	Tensile, psi	Yield, psi	Elong., % in 2 in.	Bhn
1958										
1-10	48	2.35	3.86	0.20	—	0.051	62,250	47,500	23.0	163
1-15	57	2.46	3.74	0.22	0.82	0.066	68,500	53,000	21.5	179
1-16	64	2.45	3.74	0.22	0.72	0.051	65,250	50,000	23.0	170
1-29	102	2.42	3.74	0.19	0.53	0.039	60,000	48,000	19.0	149
1-29	103	2.48	3.73	0.20	0.76	0.051	64,250	51,000	24.0	163
1-30	110	2.45	3.79	0.19	0.73	0.045	63,500	50,000	25.0	159
1-31	117	2.54	—	0.20	0.75	0.049	64,000	50,000	23.0	166
2-3	128	2.45	3.79	0.20	0.76	0.050	64,500	51,250	24.0	163
2-4	137	2.54	3.77	0.22	0.77	0.053	64,500	50,500	22.5	159
2-5	162	2.58	3.84	0.20	0.73	0.050	63,750	48,500	24.0	166
2-7	178	2.62	3.81	0.22	0.72	0.052	64,000	50,000	23.5	170
2-10	191	2.29	3.74	0.22	0.77	0.056	64,000	49,000	23.5	166
2-17	234	2.23	3.73	0.19	0.73	0.050	61,000	47,000	24.0	159
2-18	243	2.16	3.82	0.19	0.72	0.047	61,000	46,250	24.0	163
2-19	255	2.19	3.83	0.20	0.72	0.054	61,250	45,750	25.5	163
2-24	289	2.35	3.76	0.19	0.76	0.051	62,000	46,250	24.0	156
2-26	319	2.35	3.85	0.20	0.78	0.054	62,500	47,500	25.0	163
2-26	325	2.51	—	0.20	0.79	0.045	63,250	49,500	24.0	166
2-27	340	2.58	3.68	0.20	0.81	0.050	63,500	49,500	23.5	166
3-3	368	2.16	3.52	0.22	0.74	0.045	59,750	46,000	22.0	156
3-3	371	2.51	—	0.19	0.80	0.052	62,750	—	23.5	163
3-4	387	2.62	3.70	0.20	0.80	0.049	65,250	51,500	22.5	170
3-5	405	2.58	—	0.19	0.84	0.059	62,000	47,500	22.5	163
3-6	418	2.48	3.73	0.19	0.81	0.056	64,250	51,500	22.5	170
3-7	434	2.39	—	0.22	0.77	0.060	64,250	52,500	22.5	159
3-13	488	2.32	3.76	0.23	0.77	0.056	66,250	50,625	22.0	170
3-17	528	2.39	—	0.20	0.78	0.054	64,500	50,000	22.5	163
4-23	883	2.42	3.72	0.22	0.82	0.057	62,500	48,500	21.0	166
4-28	914	2.42	—	0.20	0.46	0.050	63,000	48,500	24.0	166
5-26	1011	2.66	3.73	0.20	0.85	0.064	64,000	51,000	23.0	163
5-27	1015	2.42	3.70	0.38	0.83	0.058	64,750	48,125	25.0	156
7-11	1165	2.48	3.70	0.27	0.84	0.059	67,000	54,000	22.5	174
8-19	1210	2.83	—	0.19	0.83	0.043	67,000	52,500	16.0	179
8-21	1229	2.45	—	0.22	0.82	0.056	67,750	51,250	23.5	174
8-22	1238	2.54	3.75	0.22	0.88	0.059	66,000	51,250	24.5	170
8-22	1243	2.51	3.62	0.22	0.89	0.072	66,750	52,000	23.0	170
9-2	1244	2.66	—	0.22	0.86	0.059	67,750	52,500	23.0	174
9-3	1250	2.54	—	0.23	0.86	0.058	67,000	50,000	23.0	166
9-14	1276	2.32	3.57	0.20	0.90	0.069	64,750	46,250	22.5	163
9-15	1277	2.58	3.77	0.23	0.84	0.049	67,250	51,250	22.5	163
9-16	1284	2.45	3.67	0.20	0.83	0.060	66,500	52,500	22.5	179
9-16	1289	2.42	3.69	0.22	0.81	0.052	65,250	51,250	22.0	163
9-17	1294	2.54	3.79	0.22	0.84	0.064	65,250	50,000	23.0	166
9-19	1313	2.62	3.71	0.22	0.86	0.060	72,000	51,875	23.0	174
9-19	1321	2.45	—	0.20	0.84	0.061	66,750	50,625	22.5	174
9-29	1325	2.69	—	0.20	0.84	0.069	65,250	50,000	22.5	170
10-1	1335	2.79	—	0.22	0.81	0.066	66,500	51,250	22.5	166
10-1	1341	2.48	—	0.19	0.82	0.052	65,500	51,250	24.0	170
10-2	1343	2.72	—	0.23	0.82	0.066	67,000	50,000	23.0	163
10-13	1356	2.45	—	0.23	0.86	0.064	65,250	50,000	23.0	163
10-14	1363	2.62	—	0.20	0.84	0.076	65,500	50,625	24.5	163
10-14	1371	2.54	3.75	0.22	0.84	0.068	67,750	52,500	23.0	170
10-15	1374	2.62	3.75	0.19	0.84	0.059	65,250	50,000	23.0	163
10-16	1383	2.82	—	0.20	0.84	0.066	65,000	50,000	24.0	163
10-16	1389	2.48	—	0.23	0.86	0.077	66,000	51,875	24.0	166
10-17	1393	2.48	3.76	0.23	0.82	0.073	65,000	50,000	25.0	159
10-17	1399	2.66	—	0.23	0.84	0.071	66,750	51,875	24.5	163
10-20	1404	2.58	—	0.23	0.84	0.073	66,500	50,625	25.5	163
10-21	1411	2.58	—	0.22	0.83	0.063	65,500	50,000	24.0	159
10-24	1432	2.66	3.72	0.23	0.82	0.062	65,000	48,750	23.0	163
10-27	1434	2.45	—	0.22	0.82	0.074	65,750	50,000	24.0	163
10-28	1446	2.58	—	0.22	0.80	0.051	65,750	52,500	24.0	163
10-30	1457	2.79	—	0.20	0.82	0.065	66,750	52,500	24.0	163
10-31	1464	2.72	—	0.22	0.82	0.070	64,500	50,000	24.0	163
11-3	1472	2.72	—	0.20	0.82	0.056	65,750	50,000	24.0	156
11-4	1480	2.48	—	0.22	0.84	0.071	72,200	50,000	23.0	163
11-5	1483	2.69	—	0.24	0.81	0.070	66,500	50,500	25.0	179
11-6	1491	2.72	—	0.22	0.82	0.066	66,600	50,500	23.0	163
11-7	1500	2.48	—	0.23	0.81	0.064	66,100	50,000	23.0	170
11-7	1502	2.26	3.76	0.25	0.61	0.074	62,000	45,000	25.0	163

(Continued on next page)

TABLE 3 (continued) — PLUNGER LADLE HEATS

Date	Heat No.	Composition, %					Tensile, psi	Yield, psi	Elong., % in 2 in.	Bhn
		Si	T.C.	Mn	Ni	Mg				
1958										
11-10	1509	2.29	3.61	0.20	0.79	0.050	63,100	47,000	22.5	149
11-11	1517	2.58	—	0.23	0.79	0.070	65,600	50,000	23.5	156
11-13	1532	2.39	—	0.20	0.78	0.063	65,000	48,000	24.0	149
11-13	1541	2.32	—	0.20	0.81	0.072	65,000	48,500	24.0	156
11-14	1543	2.62	—	0.22	0.81	0.061	65,300	50,500	23.5	163
11-17	1557	2.58	—	0.20	0.81	0.057	65,250	47,500	25.0	156
11-18	1566	2.51	—	0.22	0.80	0.056	66,500	48,750	22.5	163
11-18	1569	2.39	—	0.20	0.81	0.061	63,500	46,250	23.0	163
11-20	1581	2.69	—	0.22	0.80	0.061	65,750	47,500	24.0	163
11-21	1590	2.42	—	0.23	0.79	0.065	66,000	48,750	23.0	163
11-25	1607	2.58	3.77	0.19	0.82	0.058	68,250	51,250	24.0	170
11-25	1614	2.58	3.70	0.20	0.81	0.059	66,000	47,500	24.0	163
11-26	1626	2.26	—	0.20	0.81	0.070	63,750	45,000	25.0	149
11-28	1634	2.42	—	0.19	0.82	0.070	64,750	47,500	23.5	163
11-28	1641	2.39	—	0.19	0.84	0.075	64,000	46,250	23.0	156
12-3	1651	2.58	—	0.22	0.86	0.071	65,000	46,250	22.5	163
12-4	1654	2.42	—	0.20	0.76	0.060	65,500	47,500	24.0	156
12-5	1660	2.39	—	0.22	0.80	0.059	66,750	50,000	24.0	170
12-8	1667	2.39	—	0.18	0.81	0.066	65,000	47,500	24.0	160
12-9	1678	2.39	—	0.20	0.83	0.063	64,600	45,500	23.5	156
12-10	1683	2.35	3.73	0.20	0.81	0.066	66,000	47,500	22.0	159
12-11	1690	2.51	3.68	0.22	0.82	0.072	65,750	47,500	23.5	163
12-12	1697	2.32	3.70	0.22	0.77	0.064	63,500	46,250	23.5	156
12-15	1706	2.48	3.79	0.20	0.80	0.060	68,000	47,500	21.0	159
12-16	1717	2.32	—	0.20	0.81	0.069	65,500	46,500	22.5	156
12-17	1723	2.42	3.70	0.22	0.80	0.063	66,500	49,375	23.5	159
12-18	1730	2.42	3.70	0.25	0.76	0.063	67,500	51,875	23.5	166
12-19	1742	2.42	3.69	0.20	0.81	0.063	65,000	48,000	24.0	159
12-19	1747	2.35	—	0.22	0.81	0.065	64,300	46,000	23.5	156
12-22	1751	2.39	—	0.22	0.81	0.065	65,200	48,000	25.0	163
12-22	1757	2.42	3.74	0.20	0.84	0.066	64,600	47,000	23.0	156
12-23	1762	2.35	—	0.20	0.81	0.066	61,250	45,000	—	156
12-24	1767	2.35	3.74	0.22	0.76	0.059	61,000	53,250	—	153
AS-CAST FERRITIC DUCTILE IRON										
1958										
2-17	234	2.23	3.73	0.19	0.73	0.050	76,750	56,250	14.5	187
11-7	1502	2.26	3.76	0.25	0.61	0.074	71,500	50,500	15.0	174
NORMALIZED AND DRAWN PEARLITIC DUCTILE IRON										
1958										
2-13	215	2.23	3.75	0.47	0.75	0.053	136,250	93,000	4.5	286
2-20	274	2.23	3.92	0.37	0.73	0.059	123,500	93,500	3.5	286
2-21	282	2.42	3.89	0.40	0.76	0.060	133,500	87,000	6.0	269
2-25	297	2.26	3.62	0.38	0.79	0.047	124,250	83,750	5.0	255
2-27	337	2.29	3.84	0.37	0.81	0.056	132,500	88,500	6.0	269
2-28	351	2.35	3.64	0.41	0.81	0.057	134,000	94,250	5.0	277
3-4	383	2.39	—	0.25	0.78	0.057	133,500	88,000	5.5	286
3-5	400	2.35	3.72	0.38	0.78	0.059	133,250	86,500	6.0	286
3-6	415	2.42	3.79	0.39	0.77	0.058	132,750	86,500	5.0	286
3-7	427	2.48	—	0.44	0.77	0.062	133,250	85,500	6.0	302
3-10	447	2.54	3.86	0.45	0.77	0.059	133,000	89,000	4.5	286
3-11	463	2.39	3.71	0.41	0.77	0.059	126,500	82,500	6.0	286
3-12	472	2.39	3.60	0.40	0.80	0.060	128,500	82,500	5.0	286
3-14	509	2.45	3.78	0.37	0.77	0.059	135,000	88,000	5.0	286
3-17	523	2.45	3.69	0.37	0.77	0.059	132,500	83,500	5.0	286
3-19	555	2.45	—	0.40	0.84	0.066	129,750	91,250	4.0	286
3-20	569	2.45	3.82	0.40	0.81	0.064	136,500	92,500	6.0	286
3-21	588	2.42	3.58	0.37	0.80	0.059	131,750	84,125	5.0	286
3-24	603	2.39	—	0.37	0.84	0.054	135,000	88,750	6.5	286
3-25	617	2.51	—	0.45	0.83	0.060	136,750	92,500	4.5	286
3-26	631	2.54	—	0.43	0.83	0.064	132,000	91,250	4.0	286
3-26	634	2.58	—	0.38	0.93	0.068	132,000	78,500	6.5	286
3-31	661	2.45	3.52	0.31	0.84	0.056	127,500	85,000	5.0	269
4-1	675	2.29	3.50	0.35	0.84	0.054	130,250	78,750	6.0	—
4-2	680	2.51	3.86	0.43	0.78	0.061	135,500	93,000	6.0	286
4-8	721	2.48	3.67	0.38	0.81	0.059	138,500	95,000	4.5	293
4-10	757	2.35	3.62	0.41	0.80	0.062	126,500	88,750	6.0	269
4-11	774	2.39	3.60	0.40	0.83	0.052	132,500	88,000	6.0	286

(Continued on next page)

TABLE 3 (continued) — PLUNGER LADLE HEATS

Date	Heat No.	Composition, % Si	T.C.	Mn	Ni	Mg	Tensile, psi	Yield, psi	Elong., % in 2 in.	Bhn
1958										
4-14	788	2.51	3.68	0.41	0.83	0.048	130,500	85,500	6.0	269
4-15	808	2.51	—	0.40	0.80	0.056	131.000	86,000	6.0	286
4-16	811	2.29	3.45	0.34	0.83	0.061	125,500	81,500	5.0	277
4-17	832	2.32	3.69	0.40	0.81	0.059	126,000	80,500	6.0	269
4-18	848	2.32	3.57	0.18	0.81	0.045	125.650	84,000	5.0	277
4-21	863	2.35	3.61	0.37	0.81	0.050	130.000	90,500	5.0	286
4-25	899	2.35	—	0.19	0.86	0.055	129,000	85,000	6.0	269
4-29	921	2.54	3.53	0.40	0.83	0.054	134,000	90,000	4.0	286
5-1	939	2.32	3.65	0.40	0.82	0.057	123,500	85,000	4.0	286
5-12	960	2.42	—	0.41	0.82	0.066	122,750	81,500	4.5	269
5-13	967	2.64	3.84	0.46	0.86	0.067	130.000	90,600	4.0	286
5-14	979	2.58	3.49	0.41	0.83	0.059	132,000	92,500	4.5	302
5-20	999	2.45	3.48	0.44	0.84	0.058	135,000	88,750	6.0	286
5-21	1001	2.29	3.62	0.43	0.80	0.051	131,000	83,750	5.0	255
5-22	1004	2.45	3.49	0.30	0.82	0.058	125,000	76,875	5.0	269
5-26	1006	2.45	3.53	0.40	0.86	0.068	126,000	83,125	6.0	269
5-29	1032	2.39	3.67	0.38	0.81	0.057	124,000	85,000	4.5	269
6-9	1045	2.54	—	0.44	0.82	0.058	130.000	90,000	6.0	269
6-10	1059	2.72	3.82	0.46	0.88	0.057	126.750	93,125	5.0	269
6-11	1069	2.32	3.53	0.38	0.85	0.070	124,500	80,625	4.0	277
6-12	1078	2.32	—	0.41	0.81	0.063	120.000	85,000	3.5	277
6-24	1108	2.58	—	0.46	0.82	0.060	126,750	84,375	6.0	269
6-25	1118	2.45	—	0.41	0.84	0.071	126.750	85,000	5.0	286
6-30	1127	2.26	3.77	0.41	0.75	0.057	126,750	85,000	6.0	269
7-10	1154	2.45	3.72	0.38	0.80	0.066	121,750	81,250	6.0	269
7-11	1170	2.48	3.74	0.49	0.83	0.059	124,000	77,000	6.5	255
9-18	1302	2.76	—	0.43	0.86	0.060	137,500	86,875	5.5	286
9-30	1329	2.79	—	0.43	0.82	0.061	130,000	76,250	6.0	269
10-2	1344	2.66	3.74	0.46	0.81	0.059	130,750	78,750	6.0	277
10-15	1381	2.39	—	0.37	0.84	0.077	130,000	81,250	8.0	286
10-22	1415	2.76	—	0.43	0.84	0.066	132,000	85,000	7.0	286
10-23	1425	2.58	—	0.50	0.82	0.065	129,500	86,250	5.5	302
10-29	1451	2.32	3.72	0.41	0.81	0.075	119,250	68,750	6.5	262
11-11	1510	2.79	3.69	0.46	0.82	0.061	132,000	81,250	6.5	286
11-14	1552	2.39	—	0.41	0.79	0.061	121,500	70,000	6.5	255
11-19	1572	2.58	—	0.46	0.79	0.058	128,750	77,500	6.5	269
11-24	1598	2.54	3.75	0.41	0.78	0.057	128,000	77,500	7.0	269
11-24	1606	2.39	—	0.40	0.80	0.065	125,500	75,000	6.5	262
11-26	1619	2.45	3.67	0.44	0.82	0.059	125,000	77,500	6.5	269
12-18	1739	2.29	—	0.40	0.79	0.068	127,500	75,625	7.5	255
2-12	194	2.39	3.76	0.43	0.77	0.061	121,000	79,000	8.0	248
3-27	640	2.51	3.73	0.44	0.81	0.066	116,750	67,625	7.0	223
6-23	1096	2.42	3.79	0.41	0.73	0.061	111,250	72,250	9.0	235
6-23	1097	2.32	—	0.38	0.81	0.064	116,000	75,000	8.5	248
6-26	1122	2.42	—	0.40	0.81	0.070	119,000	76,500	7.5	262
7-7	1136	2.42	3.63	0.41	0.81	0.058	112,500	75,000	8.0	269
7-8	1143	2.62	3.58	0.44	0.91	0.061	119,250	79,000	8.0	269
7-9	1152	2.54	3.76	0.51	0.82	0.067	94,250	68,750	12.0	235
7-22	1181	2.58	3.79	0.49	0.88	0.061	98,500	71,000	10.0	235
7-22	1190	2.35	—	0.40	0.84	0.066	101,000	69,000	10.5	228
7-24	1193	2.45	3.62	0.51	0.94	0.050	102,750	75,000	8.0	235
AS-CAST PEARLITIC DUCTILE IRON										
1958										
2-13	215	2.23	3.75	0.47	0.75	0.053	86,500	59,000	11.0	202
2-14	228	2.32	3.79	0.51	0.75	0.058	87,500	59,000	11.0	207
2-27	334	2.29	3.84	0.37	0.81	0.056	81,500	58,500	10.0	179
2-28	351	2.35	3.64	0.41	0.81	0.057	84,500	61,500	12.0	187
5-2	950	2.35	3.60	0.38	0.81	0.063	78,500	58,000	12.0	187

CONCLUSIONS

The indirect arc rocking furnace is of small capacity and rather slow for melting ferrous metals. Because of its design, it does not lend itself to refining or slagging operations.

This type of furnace lacks the electro-dynamic stirring action of the induction and direct arc furnace, therefore, higher initial temperatures must be attained in order to break down or dissolve graphitic carbon in the base iron prior to the magnesium treatment.

Special refractory shapes are required to line the furnace, and only the better grades of refractories can withstand the high temperatures employed. This severe application, together with limited capacity and melting rate, adds up to high refractory costs.

Indirect Arc Furnace

The advantages of the indirect arc rocking furnace lie in its simplicity of operation and its ability to

TABLE 4 — CHEMICAL AND MECHANICAL PROPERTIES

Typical Chemical Composition	Ferritic Ductile, %	Pearlitic Ductile, %
Total carbon	3.80	3.80
Silicon	2.40	2.40
Manganese	0.20	0.40
Phosphorus	0.03	0.03
Magnesium	0.06	0.06
Nickel	0.70	0.70

The mechanical properties after suitable heat treatment are determined on tensile bars from 1 in. keel bars.

	Ferritic Ductile Annealed	Pearlitic Ductile Normalized and Drawn	
		1200 F Draw	1100 F Draw
Tensile strength psi, min.	60,000	80,000	100,000
Yield strength psi, min.	45,000	60,000	70,000
Elong. in 2 in., % min.	15	8	4
Brinell hardness, max.	179	235	277

handle a large variety of ferrous and nonferrous metals. Its performance in consistency and quality is in proportion to the amount of control exercised in its application.

The magnesium alloy plunger ladle was introduced to the ductile iron licensees in the United States by The International Nickel Co. While its use in this country is comparatively new, it will be only a matter of time until refractory or ceramic manufacturers overcome its one apparent drawback of low service life of the plunger cup.

Plunger Ladle Advantages

The advantages of the plunger ladle are not limited to an electric furnace operation, but may be utilized equally as well in cupola production of ductile iron with the added advantage that the need for a forehearth in front of the cupola, with its loss of 30 to 70 or more degrees of temperature, is eliminated.

ACID CUPOLA MELTING FOR DUCTILE IRON

By H. E. Henderson

ABSTRACT

The method of manufacture of ductile iron in the acid cupola used at the author's company is given. The various phases of this procedure including carbon pick-up, ladle addition, composition control and stabilizing elements are given. Tests which were conducted in the basic cupola are also dealt with, giving the company's results of these tests.

INTRODUCTION

The author's company was one of the first foundries to begin production of ductile iron; in 1949 production began. Production was begun as in most other foundries at that time, with an acid cupola, high pig iron charge and desulfurizing with soda ash to 0.05 to 0.06 per cent sulfur. The treatment for converting to ductile iron consisted of about 2.5 per cent of no. 2 nickel magnesium alloy or 0.38 per cent magnesium added, with a late addition of 75 per cent ferrosilicon.

With this procedure there was trouble, either with castings which were nonmachinable or with castings having low physical properties. Also, the castings showed sulfide slag inclusions (one form of cope defect). It was difficult to maintain pouring temperatures high enough to prevent pinholes (another form of cope defect) in the castings. In addition to this, the cost of the charge and the alloy was excessive.

The next approach used was with the basic water-cooled cupola, which was put into operation in Feb. 1951. By using a basic slag, we found it possible to produce a low sulfur base iron with an economical charge containing a high percentage of steel scrap. However, it was found difficult to maintain a constant analysis of silicon, carbon and manganese with the basic operation. In order to maintain the sulfur below 0.02 per cent, which was desired for ductile iron, the carbon content was over 4.0 per cent and carbon segregation in the castings was experienced (still another form of cope defect).

This would show up as a dapple grey appearance on the machined surface of the cope side of the casting. There was also a noted difference in the hardness of cope and drag, and the carbon content would vary from 3.40 per cent in the drag to over 6.0 per cent in the cope. Due to these circumstances, and the unavailability of enough steel scrap sufficiently free of tramp elements in the area, basic melting was discontinued.

At this time work was begun with the Air Reduction Co. on the injection of calcium carbide for desulfurization. It was found possible to make ductile iron with much less magnesium alloy, by treating after desulfurizing with calcium carbide to less than 0.02 per cent sulfur.

Carbide Desulfurizing

This procedure requires an addition of 1.0 to 1.5 per cent calcium carbide, a sizeable cold addition of alloy, which means that the iron must be melted at a high temperature. Further, the efficiency of the carbide desulfurizing reaction increases rapidly with higher temperatures. These tests indicated that the metal must melt at 2900 F min., and preferably 2950 F for optimum results.

In the Fall of 1952, it was decided to build a shell molding foundry to produce both gray and ductile iron castings. The melting unit was selected to produce iron at maximum temperatures, with calcium carbide desulfurization outside an acid lined cupola, for the manufacture of ductile iron. The cupolas, as shown in Fig. 1, were designed to melt from eight to ten tons/hr at tap temperature above 2950 F, and to operate on alternate days for periods up to 24 hr.

In general appearance, they resemble blast furnaces more than conventional cupolas. The steel shells are 72 in. I.D., and are lined with acid refractories from 48 in. to 56 in. depending on the melting rate desired. Four 5 in. I.D. water-cooled copper tuyeres are used per cupola, giving a tuyere ratio of 4.3 per cent at 48 in. I.D. or 3.4 per cent at 54 in. I.D., and air velocities through the tuyeres up to 20,000 ft/min. The tuyeres protrude beyond the lining a distance of from six to nine in. into the coke bed, and the nose to nose distance is 36 in. between opposing tuyeres.

They slope downward by an angle of 12½ degrees to aid in keeping them clear internally of slag or other foreign materials. The tuyeres are located 36 in.

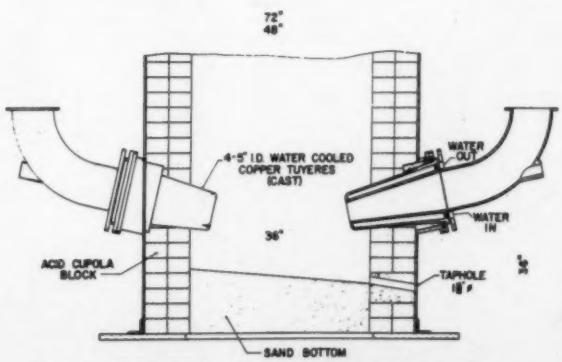


Fig. 1 — Shell mold cupola, details of lower portion.

H. E. HENDERSON is Proj. Engr., Lynchburg Foundry Co., Lynchburg, Va.



Fig. 2 — Excess tuyere cooling water is circulated to a water ring which cools the cupola shell.

above the mandrel plate where they pass through the shell.

The cooling water circuit is automatically controlled by the effluent water temperature and is maintained at 130 F. The excess tuyere cooling water is circulated to a water ring which cools the cupola shell, as shown in Fig. 2.

The blast heater is an independently fired unit using either gas or oil as fuel, and will supply 6700 cfm of blast air at temperatures up to 1000 F. The centrifugal blower is capable of supplying this air at pressures up to 60 oz.

METHOD OF MANUFACTURE

The shell cupolas were placed in operation in July 1954, and have been in continuous operation since that time. It has been found that the company is able to melt for ductile iron continuously at temperatures in the range of 2950 to 3000 F, and reaching a maximum of 3030 F.

Table 1 shows the charge for ductile iron. The 2,000 lb charge consists of 500 lb low carbon, low manganese pig iron; 700 lb selected steel scrap, as free of carbide stabilizing elements and tramp elements as possible; and 800 lb of ductile iron return scrap.

The elements in the charge are 1.02 per cent silicon, 0.32 per cent manganese, 0.054 per cent phosphorus and 2.05 per cent total carbon. The actual average analysis is 1.10 per cent silicon, 0.30 per cent manganese, 0.064 per cent phosphorus and 3.60 per cent total carbon. This shows a silicon gain of 7.3 per cent, a manganese loss of 6.7 per cent, a phosporous gain of 0.01 per cent and a carbon pickup of 1.55 per cent.

TABLE 1 — DUCTILE IRON CHARGE FOR ACID CUPOLA

		Per Cent				Pounds			
Material	Pounds	Si	Mn	P	T.C.	Si	Mn	P	T.C.
Low Mn pig	500	0.10	0.10	0.05	2.00	0.50	0.50	0.25	10.0
Steel	700	0.10	0.50	0.05	0.30	0.70	3.50	0.35	2.1
D.I. return	800	2.40	0.25	0.06	3.60	19.20	2.00	0.48	28.8
Total	2000					20.40	6.00	1.08	40.9
Element in charge						1.02	0.30	0.05	2.05
Ave. actual analysis						1.10	0.30	0.06	3.60
Gain or loss						+0.08	0.0	+0.01	+1.55

Coke, lb/2000 lb	250
Dolomite, lb/2000 lb	30
Air blast, lb-min.	500
Blast temp., F	900
Cupola I.D., in.	56
Bed height, in.	42
Melting rate, ton/hr	14

Carbon Pickup

The carbon pickup as figured by Wally Levi's formula

$$T.C. = 2.40 + \tfrac{1}{2}\,C \text{ in Charge} - \tfrac{1}{4}\,(Si + P)$$

gives an expected carbon of 3.15 per cent. However, if we add 0.20 per cent for the hot blast, 0.15 per cent for the protruding tuyeres and 0.10 per cent from the graphite tubes, we come up with the expected 3.60 per cent total carbon. The coke ratio is 8 to 1 or 250 lb/2000 lb charge of high carbon pickup coke. At a slow melting rate more coke is used, but the company changes to a low carbon pickup coke.

The flux charge is low (only 1.5 per cent) consisting of 30 lb of dolomitic limestone/2000 lb charge. The cupola is charged upon a bed height of coke 42 in. above the tuyeres. The air blast rate is 500 lb of air/min, and the blast temperature is 950 F. The cupola is lined to 56 in. inside diameter and the melting rate is 14 tons/hr.

Figure 3 shows the actual carbide treating basin in operation. The metal flows from the cupola at the left into the front slagging trough where the cupola slag is removed. About 1½ per cent calcium carbide (20 mesh by down) is fed through the two graphite tubes through the cover and injected 12 in. under the surface of the metal at the rate of about four lb of carbide/min through each tube. The dry slag is raked from the surface through the opening in the cover into a water disposal unit.

The graphite tubes are lowered at about 15 min intervals to maintain an immersion depth of 12 in. The calcium carbide is supplied from two 2000 lb capacity feeders located above the treating basin. In these feeders the nitrogen gas fluidizes the carbide, which allows it to be conveyed through a hose to the graphite injection tubes. The sulfur is lowered from 0.085 per cent to 0.015 per cent by this treatment.

Ladle Addition

In Fig. 4 the desulfurized metal is being tapped from the forehearth onto a fixed amount of magnesium alloy (now 0.19 per cent contained magnesium



Fig. 3 — Carbide treating basin in operation.

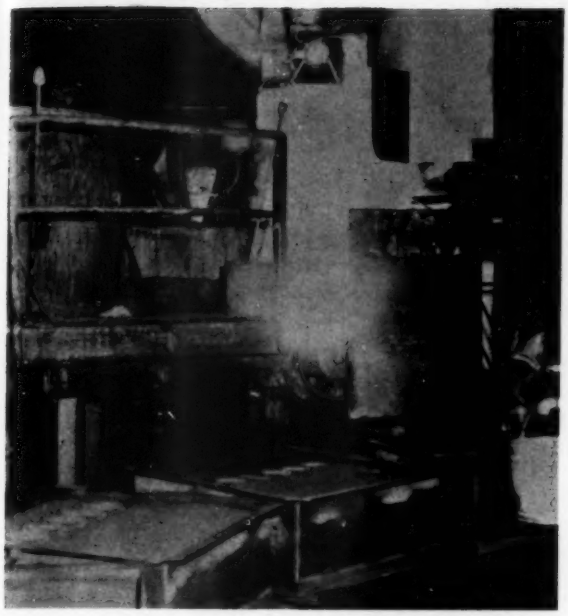


Fig. 4 — The desulfurized metal is tapped from the forehearth onto a fixed amount of magnesium alloy and flux.

in this operation) and flux. After the ladle is one-half full, mischmetal and ferrosilicon plus 0.1 per cent calcium silicon are added and the ladle is filled to the correct weight as shown on an electronic strain gage scale. The metal in the ladle is stirred, slagged, test bars poured and transferred to a pouring ladle to be poured into the shell molds backed with steel shot shown coming to the pour off station.

Figure 5 shows the pouring of the test bars. Two 1-in. keel blocks have been poured and now a 1.2 in. diameter bar, 10 in. long is being poured. In order to be sure the metal will meet specifications a bar is poured from each ladle of iron. By means of controlled cooling, the 1.2 in. bars are tested within 15 min after pouring, and before the castings represented are taken from the mold. The test consists of placing the bars on 8 in. centers and applying a transverse load midway between the centers (Fig. 6).

The load and deflection are recorded, and Brinell hardness tests made. The fracture of the broken bar is examined visually for structure and soundness. If all is in order, and the transverse load and deflection equal or exceed the specified minimum values, the castings represented are released for processing. If the fracture appears to be satisfactory but the breaking load is low, the castings in question are set aside until mechanical properties in the keel block representing these specific castings have been determined. If the fracture of the 1.2 in. bar indicates failure to convert to nodular graphite, all castings represented by it are scrapped. More than 99 per cent of our ductile iron passes this rigid inspection and meets the customer's specification.

CONTROL OF COMPOSITION

It was found, as pointed out before, that in order to make ductile iron with consistent as-cast properties, the composition must be rigidly controlled. Figure 7 shows the variation in physical properties with carbon equivalent. It has been found that for the type of castings the company manufactures the carbon equivalent must be in the range of 4.2 to 4.6. Carbon equivalents less than 4.2 give high Brinell hardness and a tendency to form carbides and excess shrinkage, while those above 4.6 tend to give carbon segregation and low yield strengths.

Figure 8 is a frequency chart for total carbon content. It may be noted that 85 per cent of the carbons fall within the range of 3.50 to 3.70 per cent, and 99 per cent within the range of 3.40 to 3.80 per cent total carbon. The carbon analysis is taken before the alloy additions, therefore, the values are slightly higher than in the castings. The carbon is controlled as in regular acid melting by manipulation of bed height, type of coke, charge materials, blast temperature, etc.

Figure 9 is a frequency chart of the base silicon and final silicon. It may be noted that 72.5 per cent of the values are within the range of 1.00 to 1.20 per cent

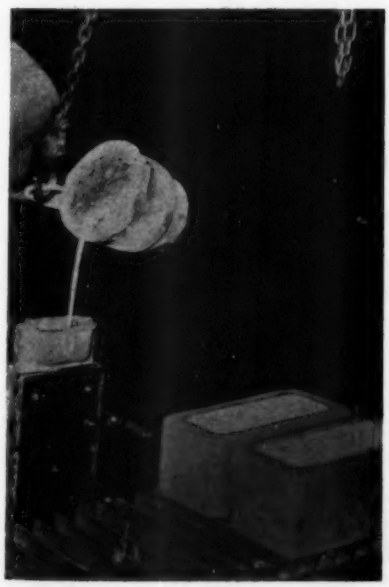


Fig. 5 (left) — Pouring of the test bars used in these tests.



Fig. 6 (right) — The bars were placed on 8 in. centers and a transverse load was applied midway between the centers.

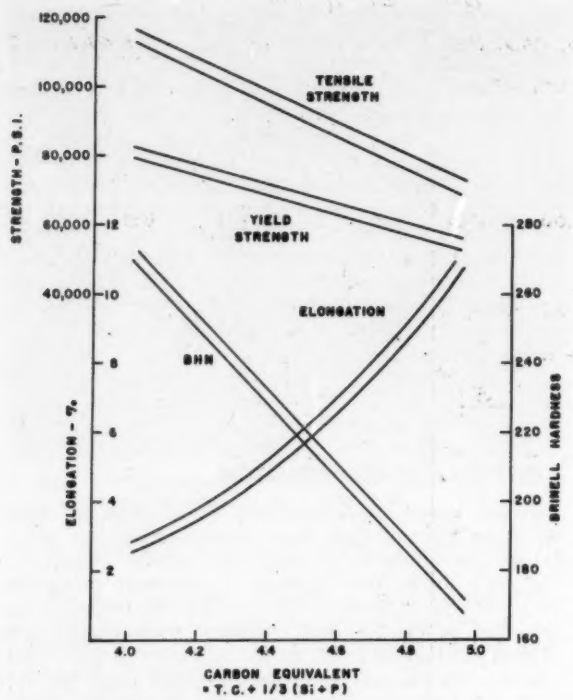


Fig. 7 — Variation in physical properties with carbon equivalent.

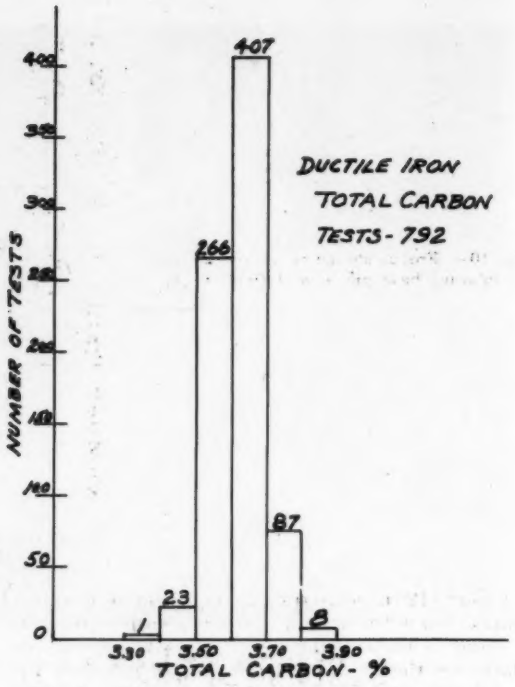


Fig. 8 — Frequency chart for total carbon content.

silicon, and 94 per cent are in the range of 0.90 to 1.30 per cent silicon. The final silicon values are 90.5 per cent within 2.20 to 2.60 per cent and 99 per cent within 2.10 to 2.70 per cent silicon. The base silicon is controlled by the charge materials and operation of the cupola, and the final silicon is controlled by varying the late ferrosilicon addition as indicated by a chill test specimen taken on the base iron.

Figure 10 is a frequency chart of manganese, phosphorus, base nickel and final nickel. The manganese is in the general range of 0.25 to 0.35 per cent, the phosphorus 0.04 to 0.07 per cent, the base nickel 0.30 to 0.50 per cent and the final nickel 0.65 to 0.8 per cent.

The second peak in the final nickel chart at 0.85 to 0.90 per cent nickel is due to the difference in alloying procedure between the shell mold foundry and the green sand foundry. More nickel magnesium alloy is added to the ductile iron poured in the green sand foundry due to the size of the castings, the geometry of the ladles and the longer pour-off time.

Stabilizing Elements

Figure 11 is a frequency chart of the carbide stabilizing elements in the metal. These materials should be kept as low as possible in the charge materials in making as-cast ductile iron. The chromium content is generally 0.06 to 0.12 per cent, the copper 0.12 to 0.18 per cent and the molybdenum less than 0.02 per cent.

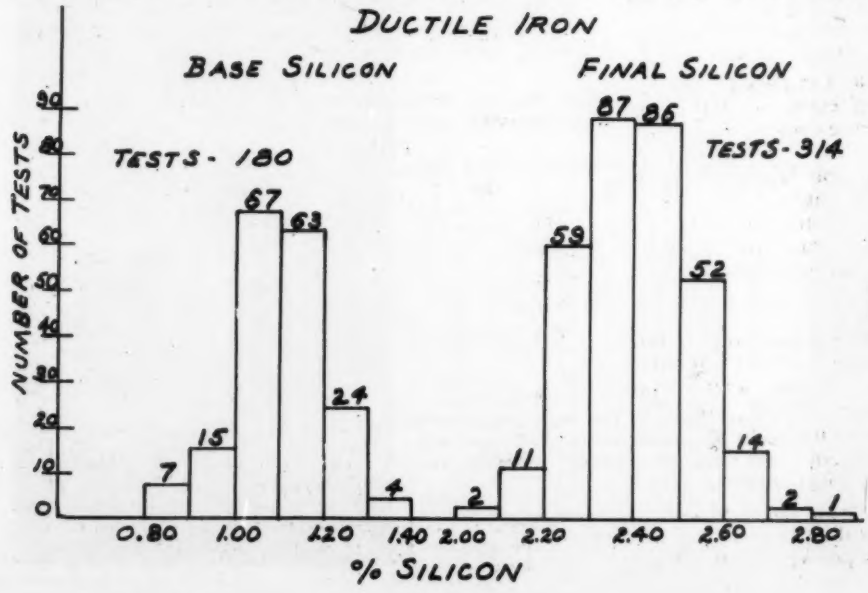


Fig. 9 — Frequency chart of the base silicon and final silicon.

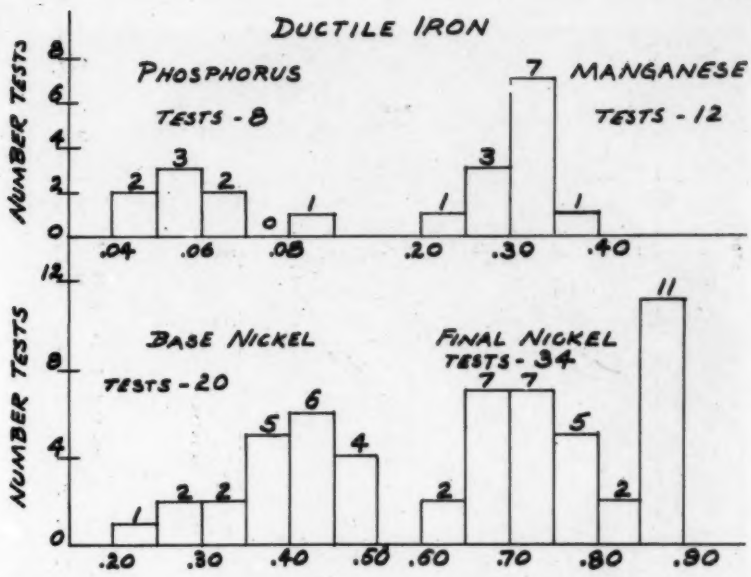


Fig. 10 — Frequency chart of manganese, phosphorus, base nickel and final nickel.

Figure 12 is a frequency chart of the sulfur as tapped from the cupola, and the sulfur after desulfurizing with calcium carbide. No significance is placed on the final sulfur analysis of the treated ductile iron, and therefore the company does not analyze for it. It may be seen that 91 per cent of the cupola sulfurs are in the range of 0.07 to 0.10 per cent; 86 per cent of the sulfurs after calcium carbide treatment are 0.010 to 0.016 per cent and 98 per cent are within the range 0.008 to 0.018 per cent sulfur.

The cupola sulfur is controlled by selection of charge materials, operation of the cupola and selection of coke, whereas the sulfur after desulfurization is controlled by the amount of calcium carbide injected and the operation of the injection technique.

It is felt that the greatest single factor in the successful casting of as-cast ductile iron is maintaining the correct relationship between sulfur, magnesium and cerium in the iron. Since no satisfactory fast analysis for the determination of magnesium and cerium has been found, the control is based on a fast accurate analysis of sulfur and an exact addition of magnesium and cerium based on this analysis.

Figure 13 is a photomicrograph showing carbides in the matrix which cause trouble in machining and tend to reduce the elongation and impact properties. This may be caused by the magnesium content being too high for the section size of the casting, or the carbon equivalent too low.

Figure 14 is a photomicrograph showing wormy graphite in the structure. This causes low physical properties; for example, 65,000 psi tensile, 55,000 psi yield and 4 per cent elongation. It may be caused by low magnesium content or high sulfur content.

Summary

As a summary of the control of composition possible in the preceding procedure of acid melting Table 2 shows the desired analysis and range of the base and final analysis for as-cast ductile iron. The desired total carbon is 3.60 per cent and range ±0.15 per cent. Base silicon 1.10 per cent ±0.15 per cent, the final silicon 2.40 per cent ±0.20 per cent. This gives a carbon equivalent of 4.20 to 4.64, which is an acceptable range for good as-cast ductile iron.

The cupola sulfur is 0.085 per cent ±0.015 per cent, the sulfur after carbide desulfurization is 0.013 per cent ±0.005 per cent, the manganese is 0.30 per cent ±0.05 per cent, the phosphorus is 0.10 per cent (the company's range is 0.04 to 0.08 per cent), the base nickel is 0.40 per cent ±0.10 per cent, the final nickel for the shell foundry is 0.70 per cent ±0.10 per cent, and for green sand 0.85 per cent ±0.10 per cent, the carbide stabilizing elements are

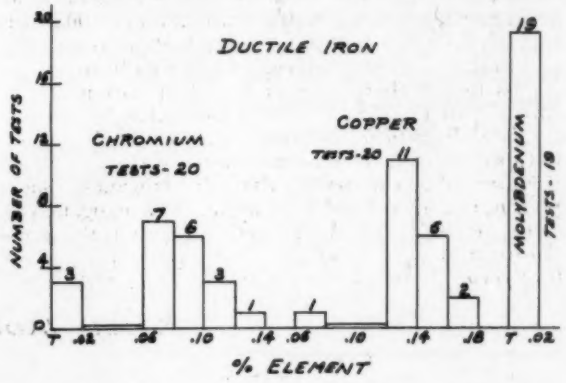


Fig. 11 — Frequency chart of the carbide stabilizing element in the metal.

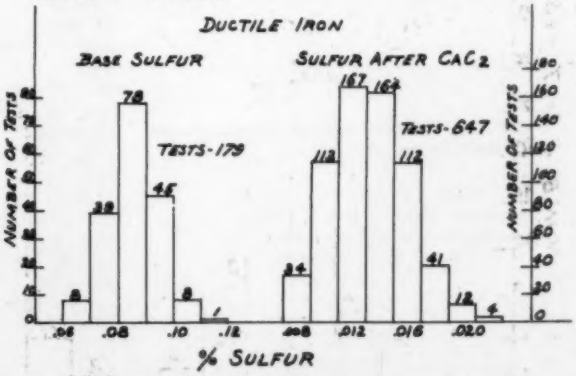


Fig. 12 — Frequency chart of the sulfur as tapped from the cupola, and the sulfur after desulfurizing with calcium carbide.

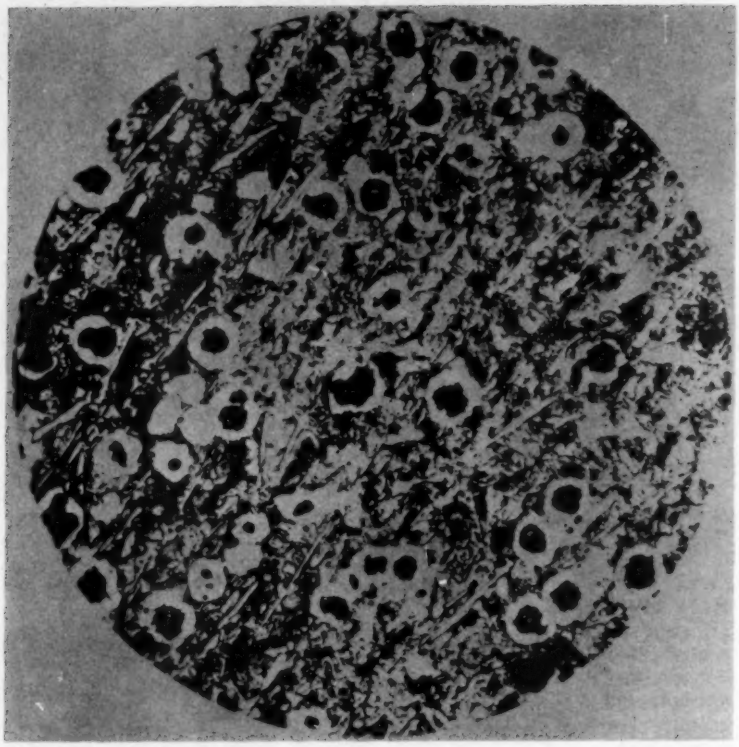


Fig. 13 — Photomicrograph showing carbides in the matrix which cause machining troubles, and tend to reduce elongation and impact properties.

chrome 0.14 per cent max. (the company's range is 0.04 to 0.12 per cent), copper 0.18 per cent max. (the company's range is 0.10 to 0.18 per cent), and the molybdenum is 0.02 per cent max. (the company's range is from a trace to 0.02 per cent).

AS-CAST DUCTILE IRON PHYSICAL PROPERTIES

Figure 15 is a frequency chart of tensile strength and yield point for as-cast ductile iron as measured in a 1 in. keel block. It may be noted that 84 per cent of

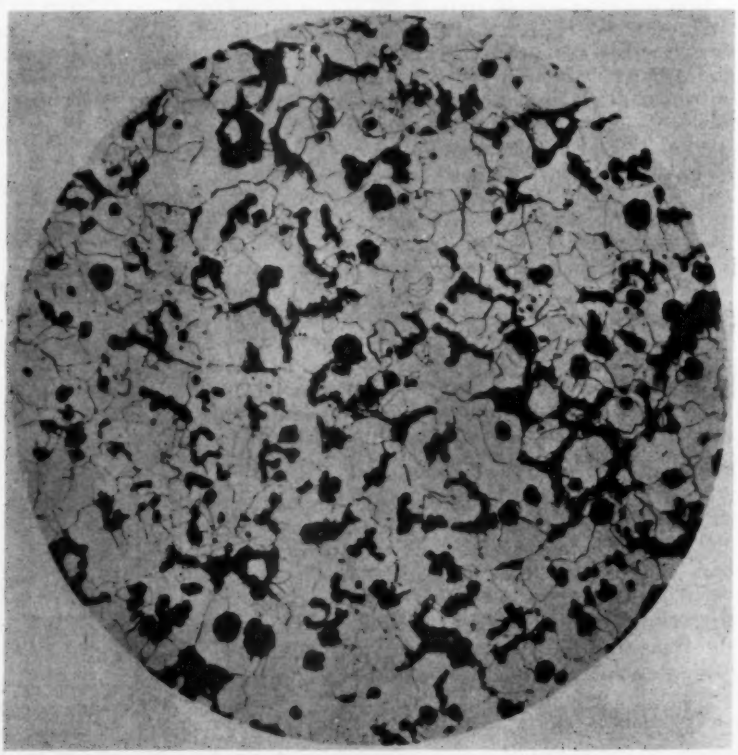


Fig. 14 — Photomicrograph showing wormy graphite in the structure.

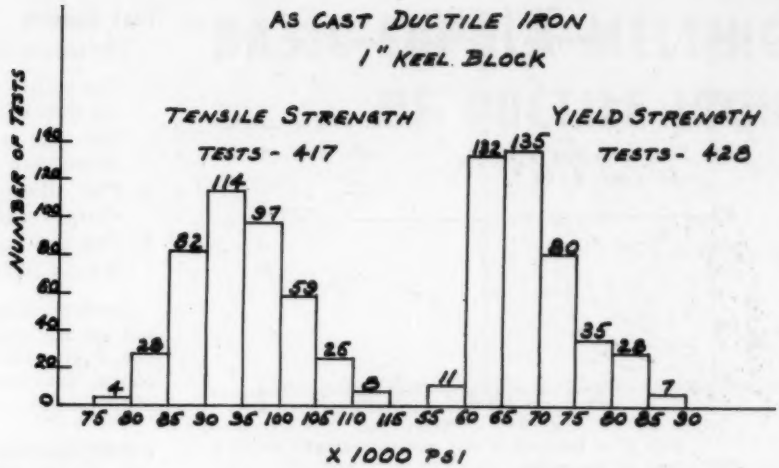


Fig. 15 — Frequency chart of tensile strength and yield point for as-cast ductile iron as measured in a 1 in. keel block.

the values for tensile strength are in the range of 85,000 to 105,000 psi, 96.5 per cent are from 80,000 to 110,000 psi, while 0.95 per cent are under 80,000 psi tensile and, therefore, represent metallurgical scrap. The yield strength values show 89 per cent between 60,000 and 80,000 psi while the values under 60,000 are 2.3 per cent.

Figure 16 is a frequency chart of elongation and Brinell hardness for as-cast ductile iron as measured in a 1 in. keel block. This shows the values of elongation to be 91.5 per cent in the range of 5 to 11 per cent with no values under 3 per cent. The Brinell hardness values are 76 per cent in the range of 200 to 230, and 98 per cent in the range of 180 to 250 Bhn. The tests on the last two frequency charts represent the most questionable ladles of ductile iron, as determined by the transverse test on each ladle.

In addition to the as-cast grade of ductile; i.e., 80-60-03, this iron may be heat treated to grades 60-45-15, 100-70-03 and 120-90-02 or oil quench hardened to Rockwell C hardness of 54 to 58. Also grade 100-70-03 may be made as-cast by raising the manganese to 0.80 to 0.90 per cent and increasing the magnesium alloy slightly.

LATEST TESTS IN BASIC CUPOLA MELTING

It was decided to investigate this field again, incorporating all the knowledge obtained from literature and personal contacts in the industry. These tests were run in the 87 in. I.D. water-cooled cupolas Dec. 15, 1958 and Jan. 8, 1959.

The metal charge consisted of about 60 per cent steel scrap, 30 per cent ductile iron return and cast scrap and 10 per cent low manganese silvery pig iron. The coke charge in Test No. 1 was—15 per cent of a low carbon pickup coke, and in test No. 2—13 per cent of the same coke. The flux charge was 10 per cent dolomite and 2 per cent fluorspar. The air rate

TABLE 2 — RANGE OF COMPOSITION — AS CAST DUCTILE IRON

Element	Desired, %	Range, ± %
Carbon	3.60	0.15
Base silicon	1.10	0.15
Final silicon	2.40	0.20
Base sulfur	0.085	0.015
After CaC_2 sulfur	0.013	0.005
Manganese	0.30	0.05
Phosphorus	0.10 max.	0.02
Base nickel	0.40	0.10
Final nickel	0.70-0.85	0.10
Chromium	0.14 max.	0.04
Copper	0.18 max.	0.04
Molybdenum	0.02 max.	0.01

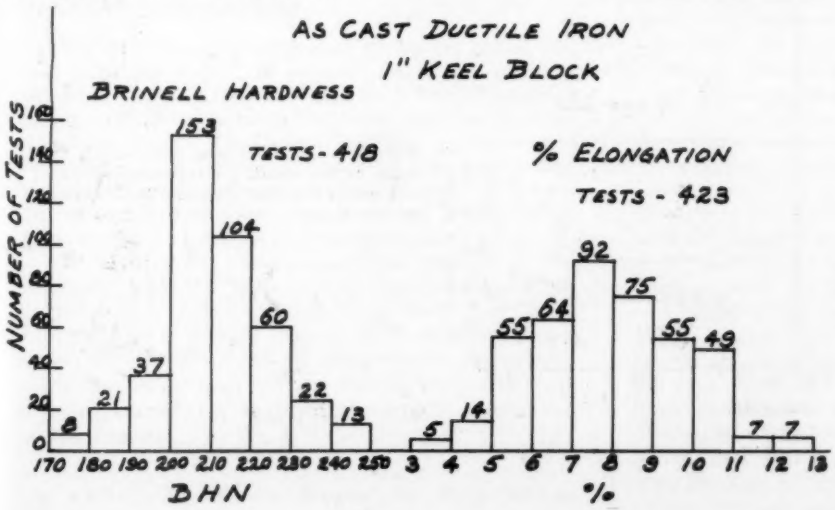


Fig. 16 — Frequency chart of elongation and Brinell hardness for as-cast ductile iron as measured in a 1 in. keel block.

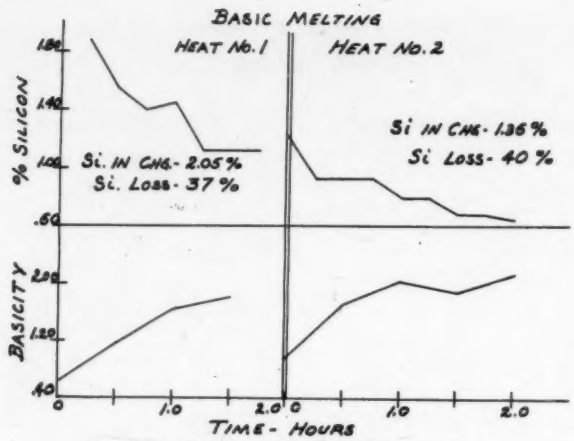


Fig. 17 — Basicity of slag and silicon content of the metal.

was 900 lb/min, and the blast temperature was varied from 600 to 800 F.

Figure 17 shows the basicity of the slag and the silicon content of the metal. It may be noted that as the basicity increases the silicon content decreases. Test No. 1 gave a silicon loss of 37 per cent, and Test No. 2 gave a silicon loss of 40 per cent.

Figure 18 shows the total carbon, and sulfur contents of the metal, the tap temperature of the metal and the MnO and FeO contents of the slag. It may be seen in Test No. 1 that as the tap temperature increases to the maximum of 2940 F the total carbon content reaches its maximum of 3.98 per cent, and the MnO and FeO contents are at a minimum; 0.56 and 0.66 per cent, respectively. Under these conditions the sulfur is reduced to a minimum of 0.037 per cent. Test No. 2 shows that if the tap temperature is maintained at 2800 to 2850 F in order to maintain a maximum carbon content of 3.80, then the sulfur content is in the range of 0.057 to 0.067 per cent.

Test Results

The results of these tests indicate that

1) The sulfur is not reduced to the desired level of less than 0.02 per cent.
2) The total carbon content is above the desired level of less than 3.75 per cent.
3) The silicon loss is about 40 per cent of the charged silicon.
4) The melting rates are only ¾ those of acid melting; i.e., 22½ tons/hr vs. 30 tons/hr.

From these tests it has been concluded that we cannot get the control necessary for making as-cast ductile iron with basic cupola melting, and have, of course, continued to melt acid in the cupolas at both of the author's company's plants.

CONCLUSIONS

Over 4000 tons of ductile iron were poured last year at one of the company's plants, mostly into shell mold castings. Approximately 80 per cent of this production was shipped and used in the as-cast state. The metallurgical, foundry and field losses have been comparable to alloy gray iron castings. All of this iron was melted in the acid cupolas and controlled as given in this report. In addition, six times this amount of gray iron was melted in these same cupolas.

FUTURE WORK

There are, of course, other satisfactory methods of making ductile iron, and there will be new methods and procedures developed. The pressure ladle technique of adding magnesium has been investigated, and the company feels it may have an application on large castings. The plunging technique of adding magnesium is highly thought of, and the company is testing it at the present time. Also the company is continuously working to improve the efficiency of the desulfurizing procedure.

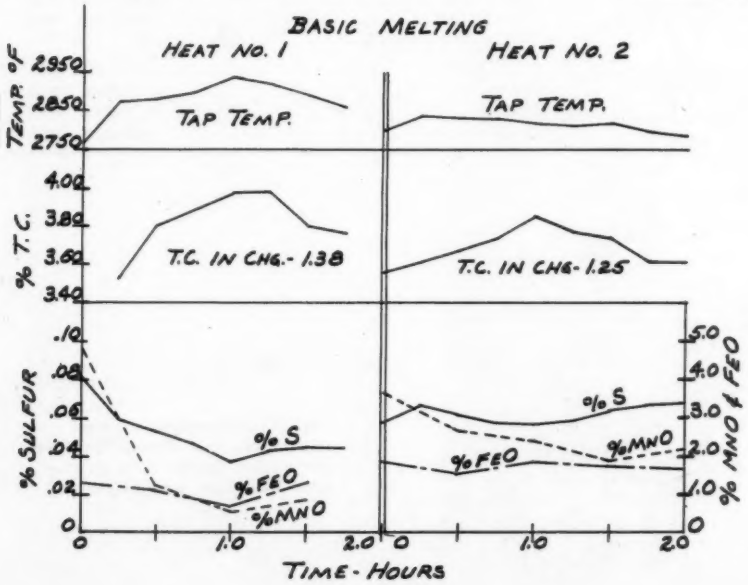


Fig. 18 — Total carbon and sulfur contents of the metal, the tap temperature of the metal and the MnO and FeO contents of the slag.

BASIC CUPOLA MELTING OF DUCTILE IRON

By J. T. Williams

ABSTRACT

The basic cupola melting practice at the author's company is presented, giving the various aspects — slagging technique, metal charge used, basicity required, slag analysis — and other phases.

The advantages and disadvantages of the basic cupola method of melting ductile iron are given, as compared with the acid cupola used at the author's company.

INTRODUCTION

A basic cupola is a variation of the more commonly used acid cupola. It has a basic lining in the melting zone and well, and the charge contains enough lime and magnesia to maintain a basic slag while melting. Basic cupola melting produces a high carbon, low silicon, low sulfur gray iron that has in recent years been used widely in the production of ductile iron and irons subject to thermal shock.

At the author's company a basic cupola for melting ductile iron has been in use since the summer of 1951. The International Nickel Co. has reported that half of the ductile iron produced during 1957 was melted in the basic cupola. The author's company now operates a continuous tap, 78 in. diameter, basic-lined, conventional cupola with a detachable wind box and eight 5 x 7 in. tuyeres.

The cupola is lined with firebrick against the shell followed by magnesite brick in the well and melting zone and a pneumatically placed, stabilized, magnesia enriched, calcined dolomite lining. The melting zone is lined to 48 in.

SLAGGING TECHNIQUE

The front slagging technique of taking metal and slag from the cupola is utilized. Generally speaking the front slagging spout is lined with firebrick, which in turn is covered with carbon brick and a monolithic dolomite lining. This spout regularly gives over 5 hr service. Experience has indicated that a cone bottom bucket, monorail charging system with equipment for accurately weighing each item of every charge provides good control of the ingoing material.

The high carbon, low silicon, low sulfur iron that is tapped from the basic cupola runs over the spout at about 2850 F into a conventional forehearth ladle. From this it is poured into a 500 lb ladle that contains the cerium bearing, magnesium ferrosilicon alloy that is used for making ductile iron. After the alloying reaction is complete, the iron is distributed to the foundry.

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The charge that used in the basic cupola operation averages as follows:

Metal Charge	
Purchased steel scrap, lb	900
Gates and sprues, lb	700
50% Ferrosilicon, lb	42
Total metal charge lb,	1642
Flux Charge	
Limestone, lb	115
Fluorspar, lb	22
Total flux charge, lb	137
Coke Charge, lb	215
Coke Ratio	7.6 to 1

The typical gray iron chemistry resulting from this charge is:

Carbon, %	3.90
Silicon, %	1.40
Manganese, %	0.45
Sulfur, %	0.025
Phosphorus, %	0.025

This chemistry allows the addition of a minimum of alloy to produce ductile iron, thus reducing ladle slag, alloy costs and temperature losses to a minimum.

SCRAP STEEL USE

One of the advantages of operating a basic cupola is that scrap steel, which is considerably less expensive than pig iron, is the only raw material used in the metal charge. The use of such a large amount of scrap steel is possible because the basic slag continually fluxes the ash from the coke and makes a relatively large amount of carbon available for combination with the metal. Carbon pickup in the basic cupola is about 2-2.25 per cent.

Another advantage of this method of melting is that the use of this large amount of low phosphorus steel scrap results in a low phosphorus iron. Several investigations, including work at the Naval Research Laboratory, have shown that the lower the phosphorus content the lower will be the ductile brittle transition temperature of ductile iron. For the same reason it is desirable to keep the final ductile iron silicon content below 2.5 per cent.

A disadvantage of basic melting is also apparent. Where usually less than 10 per cent of the silicon is lost in acid melting, the basic cupola losses are 35-40 per cent. Not only does this make necessary the use of more ferrosilicon, but it also produces more silica (SiO_2), which reduces the basicity of the slag. Use of a hot blast will reduce the silicon loss during melting. Because of this, less silica (SiO_2) is formed to reduce the slag basicity.

BASICITY REQUIRED

A larger percentage of flux is used than in acid melting to achieve the degree of basicity required to produce the type of iron that it is felt meets the company's requirements for making ductile iron. Thus, acid practice fluxing costs are about one-fifth basic fluxing costs. This large quantity of slag becomes too viscous to remove from the cupola, and a secondary flux is needed to give the molten slag fluidity. A 1 x 4 in. lump Mexican grade of fluorspar containing 97 per cent effective calcium floride (CaF_2) has been successfully used for this purpose.

While there are several ways of calculating slag basicity, the following method will be used at this time:

$$\text{Slag basicity ratio} = \frac{CaO + MgO}{SiO_2 + Al_2O_3}$$

Acid cupolas generally operate with slags having a basicity ratio of 0.3 to 0.5. Basic cupolas operate with slags having a basicity ratio over 1.3. Slags with ratios between these limits are generally considered to be neutral. On the average the flux charge results in the following slag analysis:

Lime (CaO), %	52
Magnesia (MgO), %	18
Silica (SiO_2), %	24
Alumina (Al_2O_3), %	3

This slag has a basicity ratio of 2.6 which is considered very basic. I have not mentioned the contribution that lining consumption makes to the above basicity ratio, but it is substantial.

DUCTILE IRON PRODUCTION

Acid cupola sulfurs run 4-5 times higher than the author's company's basic cupola sulfurs. This makes basic melting advantageous to ductile iron production, because the nodularizing alloy added to the base iron must first counteract the effect of certain undesirable elements. Sulfur is the most likely of these to be present. After the undesirable elements have been counteracted, the unconsumed excess of magnesium is available to spherodize the graphite.

Investigations have shown that sulfur reduction takes place when the sulfur in the iron reacts with lime (CaO) and magnesia (MgO) in the slag. The calcium sulfide (CaS) and magnesium sulfide (MgS) formed leave the cupola in the slag. Magnesia (MgO) is less effective in combining with sulfur than lime (CaO); and therefore, a ¾ x 2 in. lump, high calcium limestone is used rather than a dolomitic stone.

The desulfurizing power of the slag is reduced if oxidizing conditions exist and high quantities of iron oxide (FeO) are present, because the chemical reaction tends to reverse itself and return the sulfur to the iron.

Consumption of the dolomitic lining is high, and additional flux is added to the slag as the lining erodes. Comparisons with the acid cupola operations show that the basic cupola consumes at least 50 per cent more lining than does the acid cupola. Considering that acid lining costs about $14.50 per ton, and basic lining costs about $58.00 per ton, the basic patch material cost alone is at least six times the acid patch material. In addition, magnesite brick is more expensive and shows a shorter life than acid firebrick.

WATER-COOLED CUPOLAS

As metal requirements increase and the melting period becomes longer, water-cooled cupolas, usually in conjunction with hot blast, are used. In this case a minimum amount of refractory and lining are used, thus, in part, overcoming the higher basic lining cost.

Research and experimental data now available establish the following basic cupola melting principles:

1) The higher the slag basicity, the greater the carbon pickup.
2) The higher the slag basicity, the lower the final sulfur content.
3) For a given basicity, the higher the temperature, the higher will be the carbon pickup and the lower the sulfur.
4) The higher the slag basicity the greater the silicon loss.
5) There is no change in phosphorus content of the iron during melting.

The basic cupola is a melting unit in which the degree of complexity increases as the basicity of the slag increases. In order to operate a cupola of this type properly, it is imperative that all charge materials be of first quality and that they be weighed accurately. It is also important that the incoming air be controlled closely, that tap temperatures be constant and that chill tests be taken frequently so that adjustments can be made for variations.

ADVANTAGES AND DISADVANTAGES

Reasonably uniform ingoing returns, clean scrap and a skillfully lined cupola for uniform lining consumption will help to insure fairly consistent chemistry. The erosion of a large piece of lining can drastically change slag basicity and will have an unfavorable effect on the operation.

Some of the advantages of basic cupola melting are:

1) Low sulfur iron can be tapped from the cupola.
2) High carbon pickup allows the use of lower cost, low phosphorus steel scrap in the charge.
3) More suitable base iron is available for ductile iron production.
4) Somewhat higher tapping temperature.
5) Good slag fluidity.

Some of the disadvantages of basic cupola melting are:

1) Higher daily patching cost.
2) Higher lining costs.
3) Higher silicon loss during melting.
4) Greater volume of slag to handle with accompanying higher material cost.
5) More skills and more exact control are needed to operate successfully.
6) Higher fluxing cost.

CONCLUSIONS

It can be said that the basic cupola at the author's company has given all who have come in contact with it more mental exercise than any other melting unit in the company has in a good long time. To operate a basic cupola requires a degree of skill and attention far above that required to operate an acid cupola, because the acid cupola seems to adjust itself to a given set of operating conditions. This degree of skill and attention applies to every aspect of the melting operation.

RAMMING, SUPERHEAT AND ALLOYS (TYPE OF METAL) EFFECTS ON METAL PENETRATION

Report of Sand Division, Mold Surface Committee 8-H

Prepared by G. J. Vingas

ABSTRACT

Metal penetration, as discussed here, is that penetration into small masses of sand surrounded by relatively large masses of metal. The major and minor variables which influence this penetration are given. This report is a continuation of work done previously with metals other than iron and steel. The metals used for these tests are aluminum, brass, gray iron, steel and Wood's metal.

INTRODUCTION

Previous reports of the Mold Surface Committee, or those of its chairmen and members, have thoroughly covered the subject of penetration of steel and iron into small masses of sand. These reports extend back in the AFS TRANSACTIONS for a period of over ten years. The following variables have been found to influence penetration into small masses of sand surrounded by relatively large masses of sand and iron:

Major Variables

1) Relationship of mass of sand to mass of metal.
2) Pressure, as determined by the height of liquid metal above the sand surface into which the metal is penetrating.
3) Void size of the sand mass. In all but rare instances this void size is that of the sand as rammed. Under severe conditions, void size to prevent penetration is small, and the controlling factor becomes density, the amount of voids.

Minor Variables

1) Superheat.
2) Metal flow.
3) Washes and coatings, expect when the major variables are not severe.
4) Glazing (sintering).
5) Gas pressure (very minor).
6) Veining (a contributary cause).

All the above variables are interrelated, therefore, a simple analysis based on one variable is usually impossible. It should be emphasized that it has always been possible to explain penetration by one, or more of the preceding factors. In no case have oxides, compounds of any type, volatilization or sublimation been found to cause or effect penetration. The action is mechanical.

All chemical changes in the penetrated metal are those expected, due to oxidation and decarburization of ferrous metals when in contact with air and mold gasses. As will be shown in this report, chemical changes in copper-base alloys are due to normal selective freezing during solidification.

This report is a continuation of previous work with metals other than iron and steel. A statistically controlled, quantitative evaluation has been made of the following variables:

Absolute Temperature, as determined by the metal, aluminum, brass, gray iron, steel and Wood's metal.

Relative Temperature, as determined by three levels of superheat for each metal.

Void Size, a relatively small change in void size as determined by four levels of ramming of the same mold.

Test cores were made of a blended sand with the properties listed in Appendix 1. The cores contacted the metal under severe conditions of mass and pressure in a modified Gertsman test casting, as in Fig. 1. The degrees of penetration were evaluated by the members of the committee in degrees of severity from 0 to 9 (Fig. 2) according to the method outlined in the previous committee report. The individual evaluations were analyzed statistically, as detailed in Appendix 2.

The five metals were poured at constant levels of superheat, measured in Btu/cu in. above each liquidus level. In this way superheat can be separated

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Mold Surface Committee (8-H) consisted of the following personnel: G. J. Vingas, *Chairman;* J. E. Haller, *Vice Chairman;* J. G. House, *Secretary;* J. B. Caine; R. Dietert; R. Gregory; C. E. McQuiston; A. E. Murton; J. A. Ridderhof; R. W. Ruddle; R. R. Schaaf; C. J. Schwetz; C. C. Sigerfoos; and D. C. Williams.

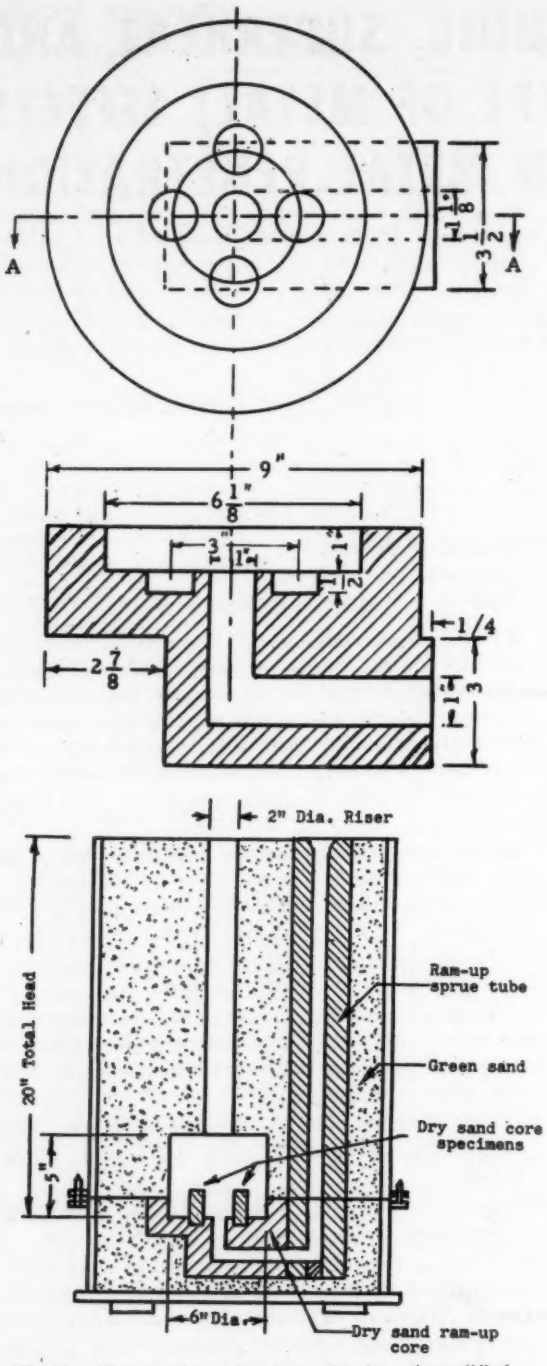


Fig. 1 — Dry sand ram-up cores (top) and modified Gertsman test mold (bottom).

from absolute temperature as determined by the liquidus level of each metal. The degrees of penetration for the five metals are shown in Table 1.

The following general conclusions can be drawn:

1) The type of metal must be added to the list given previously as a major variable.
2) Temperature per se has no influence on penetration. This point is accented by the Wood's metal experiments. These experiments were designed with just this point in mind. By far the worst penetration was experienced with this metal poured at a temperature below boiling water.

CONCLUSIONS

A statistical analysis of the minor variables leads to the following conclusions:

1) The experimental error is low. Within the limits of the experiment, the causes of penetration have been found.
2) For aluminum, ramming affects penetration while superheat does not.
3) In the case of brass, ramming does not affect penetration while superheat is an exceedingly important variable.
   For iron, both ramming and superheat have an effect.
4) The sands of Table 1 were penetrated in both steel and Wood's metal regardless of the pouring temperature differential and pressure head.

Pressure for a given metal was constant. Differences in pressure due to the densities of the various metals is one factor causing the great difference in penetration of the five metals.

The penetrated metal, of the brass and Wood's metal castings, was checked for changes in chemical composition.

Brass			Wood's Metal		
	Casting, %	Penetration, %		Casting, %	Penetration, %
Cu	84.6	69.8	Bi	48.0	49.4
Pb	4.6	14.6	Pb	22.8	20.3
Sn	4.8	11.3	Sn	13.1	17.4
Zn	5.5	3.2	Cd	16.7	13.0
P	0.029	0.032			

It will be noted that in the case of the brass casting the percentages of the lower melting constituents, except zinc, are higher in the penetrated metal. This situation should be expected from selective freezing. Volatilization of zinc is an added variable. Wood's metal is close to a quatenary eutectic, and no great change in chemical composition should be expected.

It should be emphasized that penetration, as dis-

TABLE 1 — PENETRATION OF TEST CASTING (AVG. OF SEVEN EVALUATIONS)

No. of rams	Density, % solid	Superheat Levels: Aluminum	Brass	Iron	Steel	Wood's Metal
0	54.5	0.71 0.86 1.0	6.8 8.3 8.8	1.0 1.4 1.8	over 9 for	completely
¼	57.2	0.43 0.43 0.43	6.4 8.3 8.8	1.0 1.3 1.4	all degrees	penetrated
1	59.0	0.14 0.14 0.0	6.4 8.0 9.0	0.28 0.71 0.71	of ramming	under all
3	60.7	0.0 0.0 0.0	6.3 8.0 8.8	0.0 0.28 0.43	and super-heat	conditions

cussed in this report, is that into small masses of sand surrounded by relatively large masses of metal. Such penetration may differ from slight penetration that causes variation in surface smoothness when the relative masses of sand and metal are reversed. This latter case is now the concern of the Committee.

APPENDIX 1

Cores were made of the following sand mix:

Sand	Per Cent
Port Crescent	38
Juniata Bank	47
Silica Flour	5
Cereal	1.25
Core Oil	1.75
Iron Oxide	1.50

Mulled	Time, min
Dry Additions	1
Water	2
Core Oil	3
Total	6

Baked at 400 F, 1½ hr
Distilled water, pH 6.2, added to 4.9 per cent

Screen analysis of the sand

Screen Size	Retained, %
43	trace
53	trace
60	12.1
70	37.0
100	27.2
140	11.8
200	6.1
270	2.8
Pan	3.0
AFS fineness no.	77

The physical properties of the sand

Green permeability	56
Green strength, psi	1.5
Green hardness	63
Flowability (Dietert)	84
Deformation, in./in.	0.050
Baked compressive strength, psi	789
Baked tensile strength, psi	166
Baked scratch hardness	90-95
Overhang, no. jolts, low cam	27
Overhang, no. jolts, high cam	9
Cracking, no. jolts, low cam	175
Cracking, no. jolts, high cam	67
Sagging, at 5 jolts, in.	0.0095

No spalling at 2 or 12 min at 2500 F; a few hairline cracks.

	0 Ram	¼ Ram	1 Ram	3 Rams
Free expansion, 1500 F, in.	0.030	0.029	0.026	0.027
Baked density, % solid	54.5	57.2	59.0	60.7

APPENDIX 2

To calculate the degree of superheat, the following equation was used:

$$Q = WC\,(t_2 - t_1)$$

where:

Q = heat addition in Btu.
W = lb of metal.
C = ave. specific heat.
t_1 = melting point.
t_2 = temp. attained after heat addition.

It was decided to add Btu/in.² of molten metal as follows:

Btu/in.² added	Gray Iron	Copper	Aluminum	Wood's Metal	Steel
300	2395	1935	1285	360	2785
600	2490	2020	1415	560	2850
900	2550	2110	1545	770	2910

When Wood's metal was cast it could not be contained in regular fine sand, therefore, it was decided to calculate the superheat on a ratio of pouring temperature relative to the other metals, as:

TEMPERATURE OF SUPERHEAT

Btu/in.² added	Metal Temperature: Gray Iron	Copper	Aluminum	Steel
300	1.015	1.045	1.707	1.022
600	1.07	1.102	1.220	1.045
900	1.09	1.140	1.330	1.062

Wood's metal was then calculated, based on the ratio of aluminum; 1.1, 1.2 and 1.3, and was poured at 174, 189 and 206 F.

CHEMICAL AND THERMAL PROPERTIES OF METALS

	Chemical Composition: Gray Iron	Copper Alloy	Aluminum Alloy	Wood's Metal	Steel
	C, 3.0-3.2	Cu, 82.5	Al, 92.21	Pb, 26.0	C, 0.25
	Mn, 0.78	Sn, 4.53	Cu, 4.66	Sn, 13.0	Mn, 0.70
	P, 0.12	Pb, 4.72	Si, 2.40	Cd, 12.0	Si, 0.50
	S, 0.06	Zn, 5.45	Mn, 0.17	Bi, 49.0	
	Si, 1.75		Fe, 0.56		
Specific heat, Btu/lb/F	0.165	0.110	0.230	0.041	0.18
Latent heat of fusion, Btu/lb	41.4	79.9	93.0	17.2	52.5
Melting point, F	2330	1850	1160	158	2725
Density, Lb/in.³	0.26	0.316	0.101	0.36	0.28

APPENDIX 3

STATISTICAL ANALYSIS OF TESTS PERFORMED

A three way analysis of variance, i.e., three independent variables, namely metal, superheat and ramming acting on the dependent variable, metal penetration, was performed. In an analysis of variance, the hypothesis tested is that no independent variable or combination of independent variables influence the dependent variable.

If the amount of variation of the dependent variable caused by an independent variable or combination exceed the chance probability, then the basic hypothesis is rejected and the independent variable is said to be significant. This means that variation caused by the independent variable can not be due to chance, and thus must be a cause and effect relationship.

In this analysis the significance level was chosen as 0.95, i.e., assignable cause of variance could be made with a probability of accuracy 95 times out of 100. $F_{0.95}$ is the F ratio for this 0.95 probability of being correct.

By examination of the above table the following conclusions can be drawn based on the experiment:

1. The experimental error is low; the assignable causes of penetration have been investigated.
2. Two interactions were found to cause penetration; metal-ramming and metal superheat.

When interactions are found to be significant, as above, the independent variables can not be tested

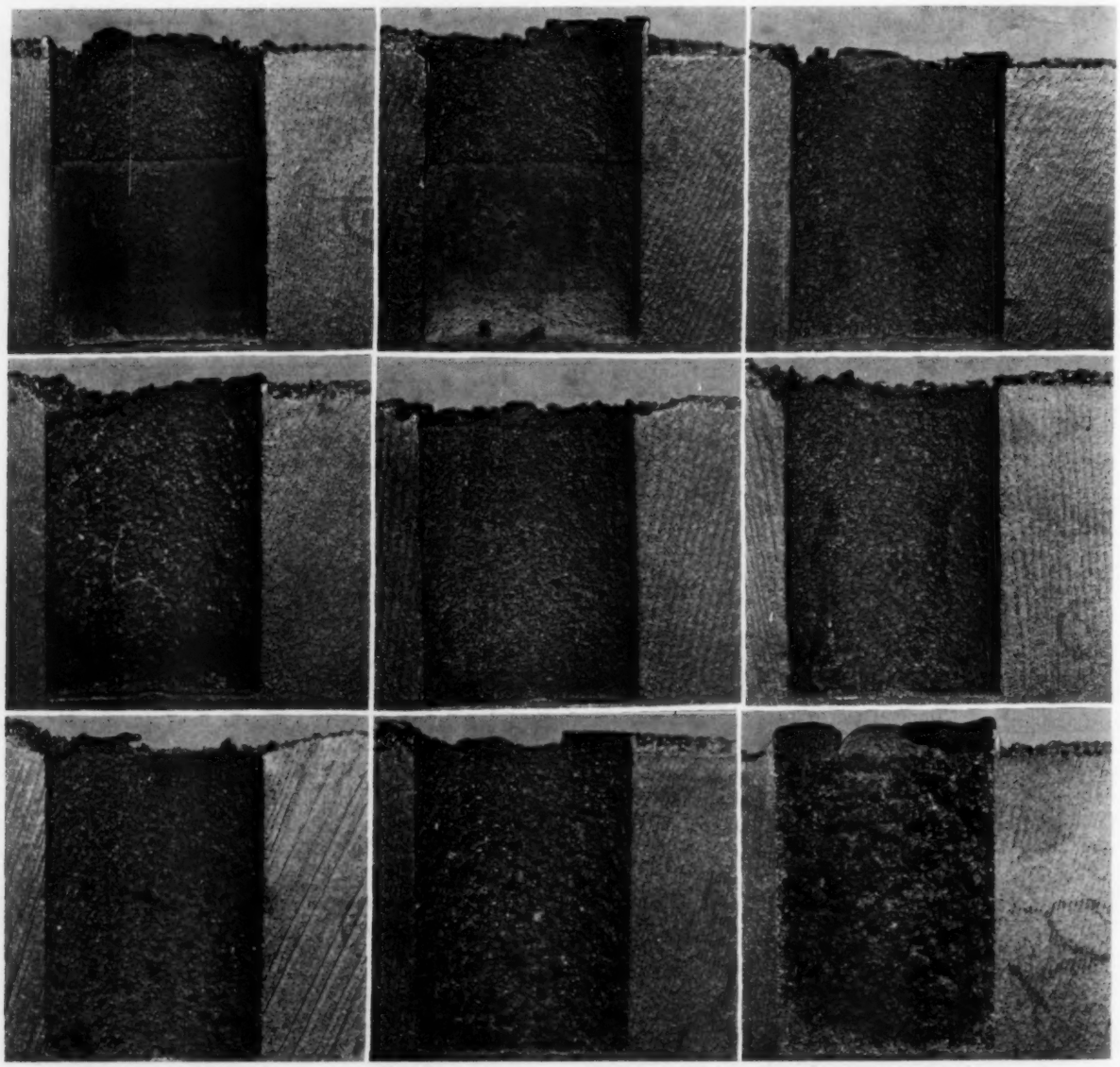


Fig. 2 — Typical results of standard Gertsman tests (re: C.C. Sigerfoos). *Top row* (left to right) — Silica flour, X 300 normal; Silica flour, X 200 normal; Crystalline graphite, X 200 skew right. *Center row* (left to right) — Mexican graphite, X 300 normal; Mexican graphite, X 200 normal; Crystalline graphite, X skew right. *Bottom row* (left to right) — Crystalline graphite, X 300 skew right; Crystalline graphite, X 300 normal; Uncoated, X skew right.

for significance singly. To illustrate, the interaction is analogous to $x \cdot y = a$. It is not possible to plot a knowing only x.

To overcome this problem, a further analysis was performed called a breakdown analysis of variance. Here the level of variables is reduced by one. In this case a breakdown analysis was performed for each metal. The results are given in the table of Analysis of Variance.

It should be remembered that the following conclusions are limited to this experiment, and any generalization to actual practice should take into consideration the limits of this experiment.

The table following is the result of the analysis of variance for this experiment.

Source of variation	Sum of square	d.f.	Mean square	F	$F_{0.95}$	Sign
Metal	419.74	2	109.87			
Superheat	6.04	2	3.02			
Ram	3.19	3	1.06			
Interactions						
Metal superheat	6.41	4	1.60	7.37	3.26	Yes
Superheat, ramming	0.02	6	0.00333	<1.0	3.00	No
Metal-ramming	1.06	6	0.1766	8.14	3.00	Yes
Error	0.26	12	0.0217			
Total	436.72	35				

d.f. = degrees of freedom.

The conclusions that can be drawn for the analyses are:

1) The causes of penetration in this experiment have been found, as the experimental error is low.
2) For aluminum, ramming affects penetration while superheat does not.
3) In the case of brass, the ramming does not affect penetration, while superheat is an exceedingly important variable.
4. For iron, both ramming and superheat have an effect.

ANALYSIS OF VARIANCE

Source of variation	Sum of square	d.f.	Mean square	F	$F_{0.95}$	Sign
M_1 Aluminum						
Ramming	1.42	3	0.473	82.98	4.35	Yes
Superheat	0.01	2	0.005	0.88	4.74	No
Error	0.04	7	0.0057			
Total	1.47	12				
M_2 Brass						
Ramming	0.11	3	0.0366	1.60	4.35	No
Superheat	11.91	2	5.96	260.26	4.74	Yes
Error	0.16	7	0.0229			
Total	12.18	12				
M_3 Iron						
Ramming	2.73	3	0.91	91.0	4.35	Yes
Superheat	0.53	2	0.265	26.5	4.74	Yes
Error	0.07	7	0.01			
Total	3.33	12				

d.f. = degrees of freedom.

COMPARISON OF X-RAY QUALITY AND TENSILE PROPERTIES IN CAST HIGH STRENGTH STEEL

By H. R. Larson, H. W. Lloyd and F. B. Herlihy

ABSTRACT

Results of tests performed with x-rays to correlate soundness with tensile properties are given. Plates ½ and 1 in. thick were used for this investigation. Tensile test bars 0.505 or 0.252 in. in diameter were machined from these plates. Radiographs were taken of ⅛-in. slices ground to 0.020 in. and examined for traces of shrinkage at 20×. Results indicate that slight traces of microshrinkage have a rather pronounced effect on ductility of tensile specimens of cast ultra high strength steel.

INTRODUCTION

The problem of developing high integrity steel castings for aircraft applications is obviously not a simple one. To be competitive with other materials they must be heat treated to high strength levels, 180,000 psi or above. At these high strength levels ductility and impact strength are lower than at the strength levels commonly used today. Because of this increased notch sensitivity, it is quite possible that casting defects will have a more detrimental effect on properties than the same defects at lower tensile strengths.

Although random dirt and slag defects, blowholes, etc., can be minimized by careful foundry practice, microshrinkage may be present to some extent in all except the most carefully fed areas. The effect of varying amounts of microshrinkage on the tensile properties at high strength levels is an unknown factor.

Before investigating the effect of unsoundness, it is necessary to determine the ultimate properties which can be developed in well fed test bars. The authors, in another paper,[1] have described alloy compositions, melting practices and heat treatments which gave improved mechanical properties in test bar castings. Although the ultimate goal is to produce castings in which metal quality approaches these test bar properties, it seemed that a logical intermediate step would be to investigate properties in simple shapes in which unsoundness would be present in varying degrees.

Thus, it would be possible to correlate mechanical properties with x-ray quality. It would also offer a rough guide as to the distance from risers and chills that one could expect adequate mechanical properties.

In an extensive series of investigations[2,3,4] Pellini and coworkers have shown that the distance which plate and bar castings can be fed sound is a function of the thermal gradients set up by the heating effect of risers and the chilling effect of edges, ends and chills. They used x-rays of the various sections to establish the limits of soundness. However, no effort was made to correlate soundness with tensile properties.

PROCEDURE

Casting Practice

The simple shapes chosen for study were plates with thicknesses of ½-in. and 1 in. The 1-in. thick plates were 6½ in. wide and 3¾, 5¾ and 7¾ in. long with a 3 in. x 6½ in. x 6 in. high hot metal riser. The ½-in. thick plates were 6 in. wide by 2, 3, 4 and 6 in. long with a 2 in. x 6 in. x 6 in. high hot metal riser.

These plate castings were poured in both oil bonded dry sand and ceramic molds, and both with and without steel chills at the ends of the plates opposite the risers. The chills were 1½ in. x 2 in. x 9 in. for the 1 in. plates and 1½ in. x 3 in. x 7 in. for the ½-in. plates.

A test bar casting, Fig. 1, was also poured with each heat. This "octabar" casting was designed for use with cast steel and yields eight 1 in. x 1 in. x 6 in. test coupons per casting. Both oil bonded dry sand and ceramic molds were used for these castings.

All castings were poured between 2950 F and 3050 F with the risers immediately covered with an exothermic hot top compound.

After cleaning and removal of the risers, the plates were x-rayed with a sensitivity of two per cent or better. They were then sectioned into alternate tensile blanks and ⅛-in. slices for microradiographs.

The tensile blanks were heat treated as follows: 1850 F–2 hr-A.C., 1575 F–2 hr-oil quench, 400 F–6 hr-A.C. Tensile bars with diameters of 0.505 or 0.252 in. were then machined from the 1 in. and ½-in. plates, respectively. The tensile specimens were shoulder grip rather than threaded specimens. A drawing of the 0.505 in. specimen is shown in Fig. 2.

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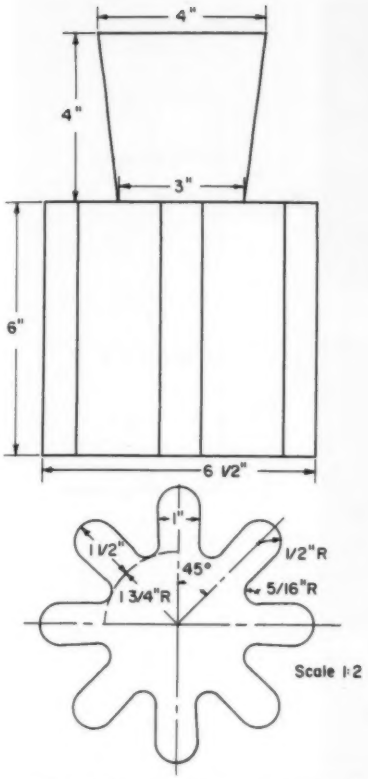


Fig. 1 — Test coupon design for cast steel.

The 1/8-in. slices were ground to 0.020 in. and radiographed on fine grain film. The radiographs were examined for traces of shrink under a microscope at 20×. Typical x-rays of the 0.020 slices magnified 10× are shown in Fig. 3. Absolutely sound x-rays were rated zero, and those with shrink were rated 1 through 4. The portion of the cross-section of the 1/2-in. plate tested by a 0.252 in. diameter tensile bar has been marked on the photographs.

Melting and Deoxidation Practice

Four heats of 500 lb were made in an induction furnace of 650 lb capacity which was lined with a rammed proprietary composition of $MgO{\cdot}Al_2O_3$. One heat of 220 lb was made in an induction furnace with a magnesia crucible.

These heats were made with high purity charge materials including special sponge iron briquets, ferromolybdenum, electrolytic chromium, nickel, graphite and high purity revert. Sponge iron heads form a considerable amount of slag on melting which was skimmed off before additions. After the bath reached 3000 F to 3050 F, ferrosilicon and electrolytic manganese were added.

All heats were tapped at 3200 F into basic lined ladles. CaMnSi in the amount of 6 lb/ton was added to the bottom of the ladle just prior to the tapping. Aluminum wire (0.06 per cent) was wrapped on a steel rod and plunged at about half tap.

Analysis of heats is shown in Table 1.

Tensile data and microradiographic results for 1/2-in. plates are presented in Figs. 4 and 5, and for 1 in. plates in Figs. 6 and 7. Figures 5 and 7 also present tensile data from 0.505 in. tensile specimens cut from the octabar test castings for comparison.

Fig. 2 — Shouldered round tension test specimen.

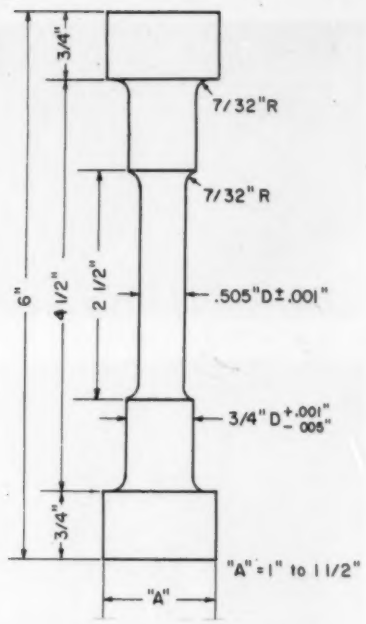


TABLE 1 — CHEMICAL ANALYSIS OF HEATS

	Composition, %								
Heat No.	C	Mn	P	S	Si	Ni	Cr	Mo	Al
58-327-1	0.39	0.68	0.009	0.013	0.36	0.59	0.85	0.40	0.031
58-333	0.42	0.77	0.015	0.011	0.38	0.60	0.89	0.40	0.046
58-339	0.40	0.67	0.010	0.013	0.30	0.59	0.90	0.40	0.050
58-349	0.36	0.59	0.010	0.013	0.30	0.59	0.82	0.41	0.072
58-368	0.34	0.89	0.010	0.014	0.39	0.59	0.85	0.40	0.044

RESULTS

It is evident that as the length of the plate increases the tendency for shrinkage and inferior properties in the center of the plate increases also. The correlation between microradiographs and ductility is rather good. Unless the test bar contains gross defects there is only a negligible change in yield or tensile strength.

The microradiograph technique applied to the 0.020 in. sections is a much more sensitive method of detecting shrinkage than is the x-ray of the plates as a whole. For example, all x-rays of the 1 in. plates poured in ceramic molds were free of shrinkage. On the other hand microradiographs of thin sections indicated considerable unsoundness in 6 in. long plates and lesser amounts in the other plates. Similarly, in the 1/2-in. plates poured in ceramic only the 6 in. plates had unsoundness by x-ray. The thin slices showed unsoundness in 3 and 4 in. plates as well. Because of surface roughness and dirt, it was difficult to interpret the x-ray of the sand cast plates.

Chilling increases the feeding distance, as would be expected. This can be seen most readily by comparing data for chilled and unchilled 1/2-in. plates with lengths of 4 in. in Fig. 4, and 1 in. plates 6 in. long in Fig. 6. Also, the sections next to the chills exhibit some exceptionally good properties. For example, one

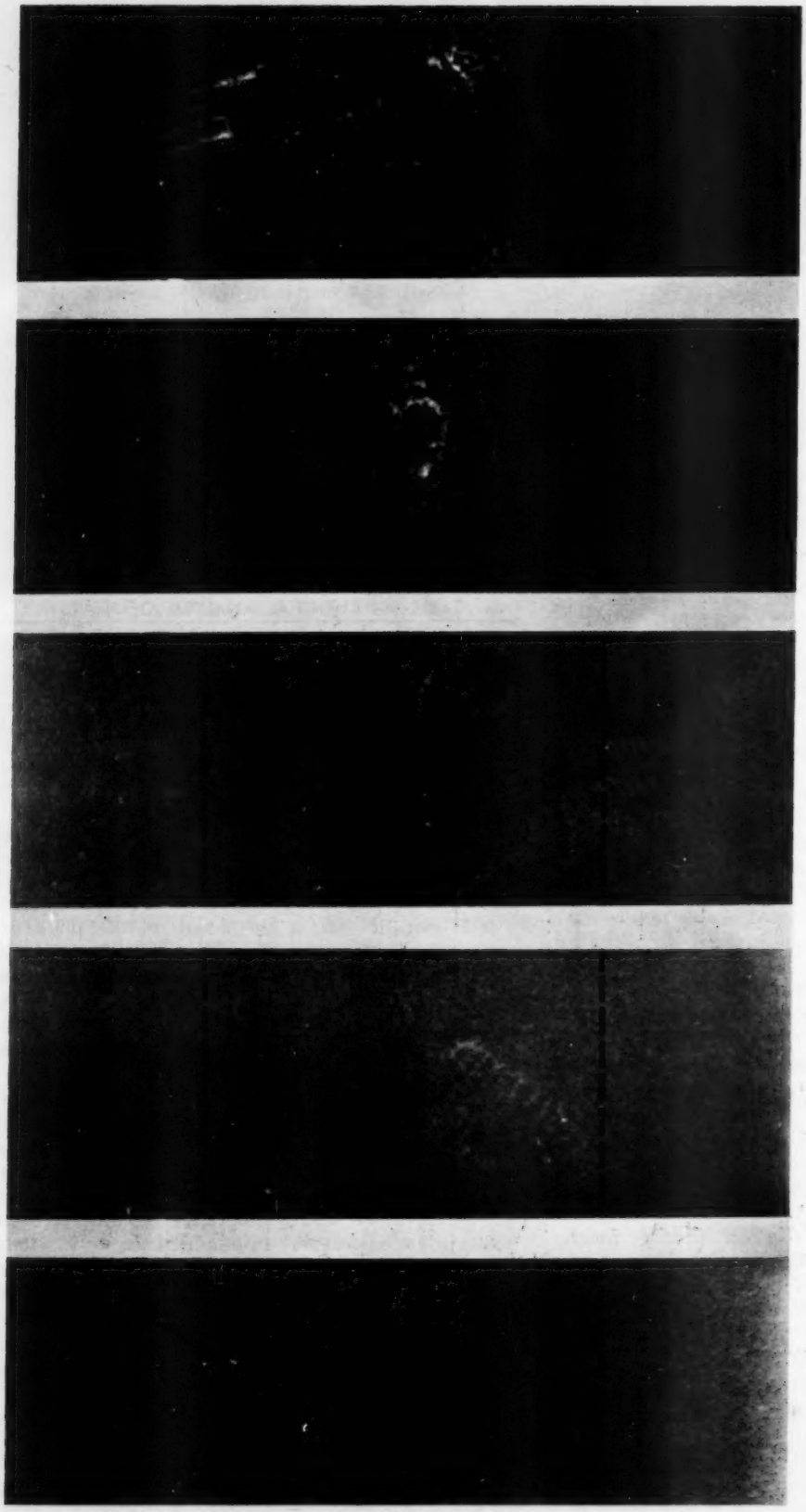


Fig. 3 — Classification of shrinkage. Radiograph of 0.020 in. slice from a ½-in. thick plate. Dotted lines show extremes of 0.252 in. test bar. 10×.

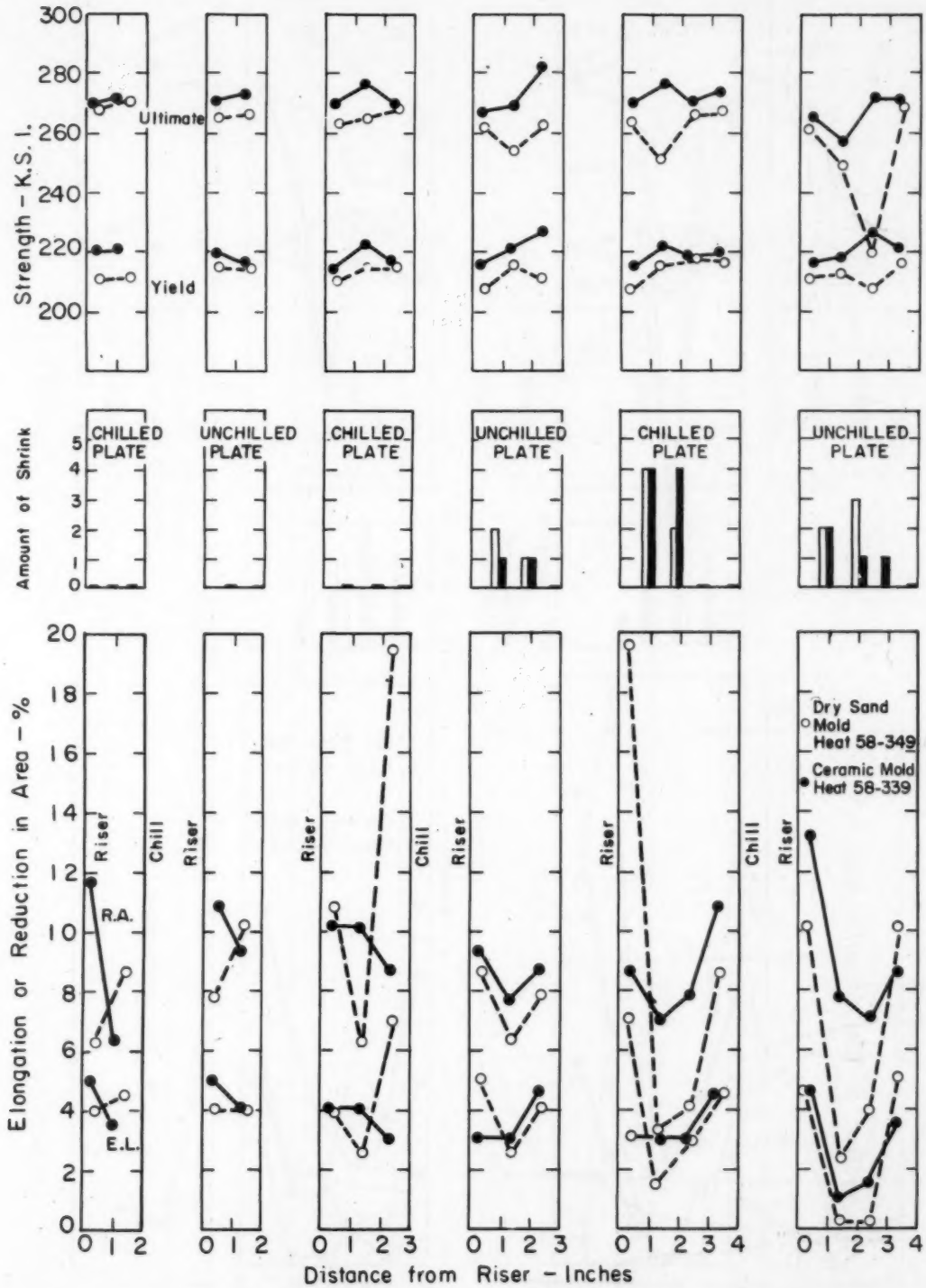


Fig. 4 — Feeding distance in ½-in. plates — 2, 3 and 4 in. long. Test bars are 0.252 in. diameter.

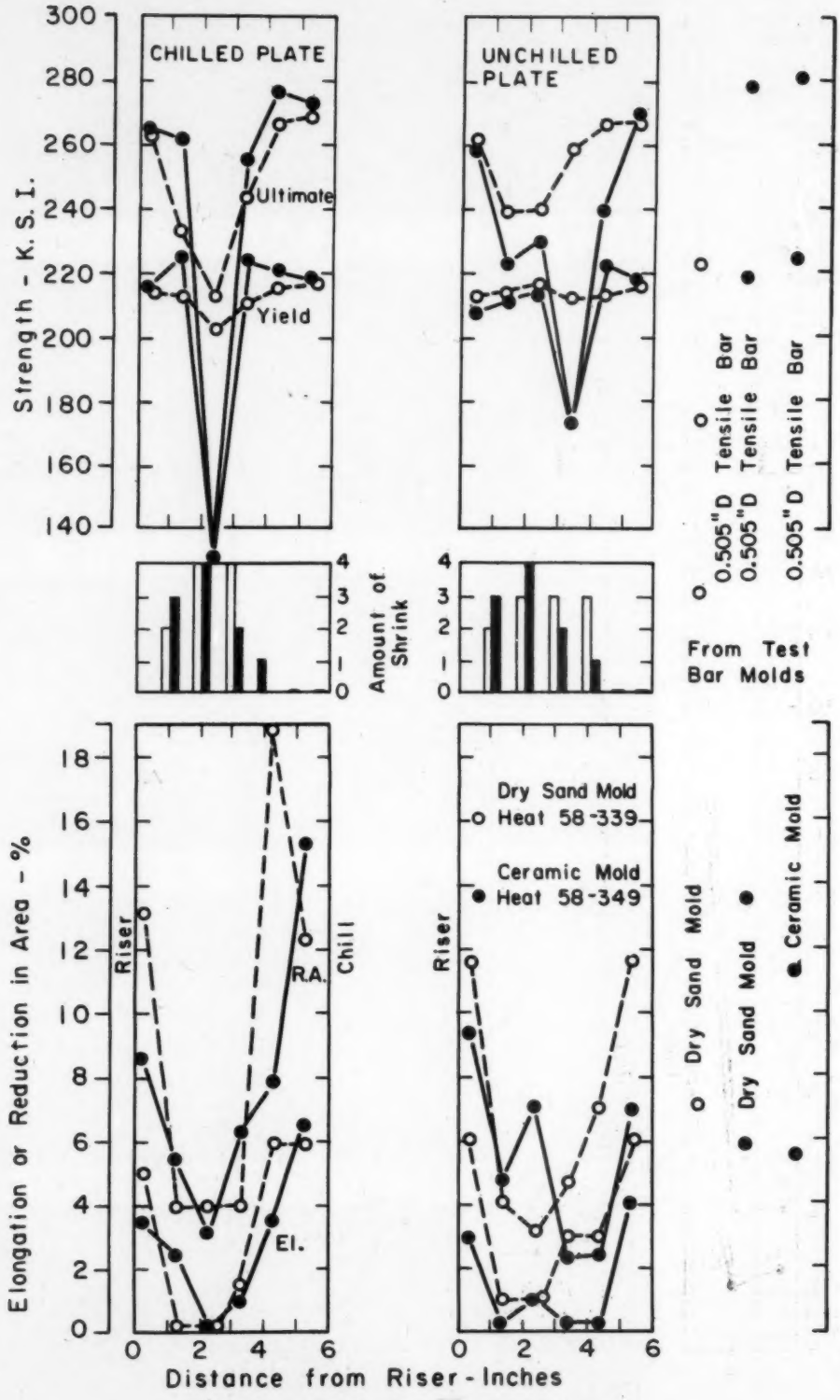


Fig. 5 — Feeding distance in ½-in. plates — 6 in. long. Test bars are 0.252 in. diameter.

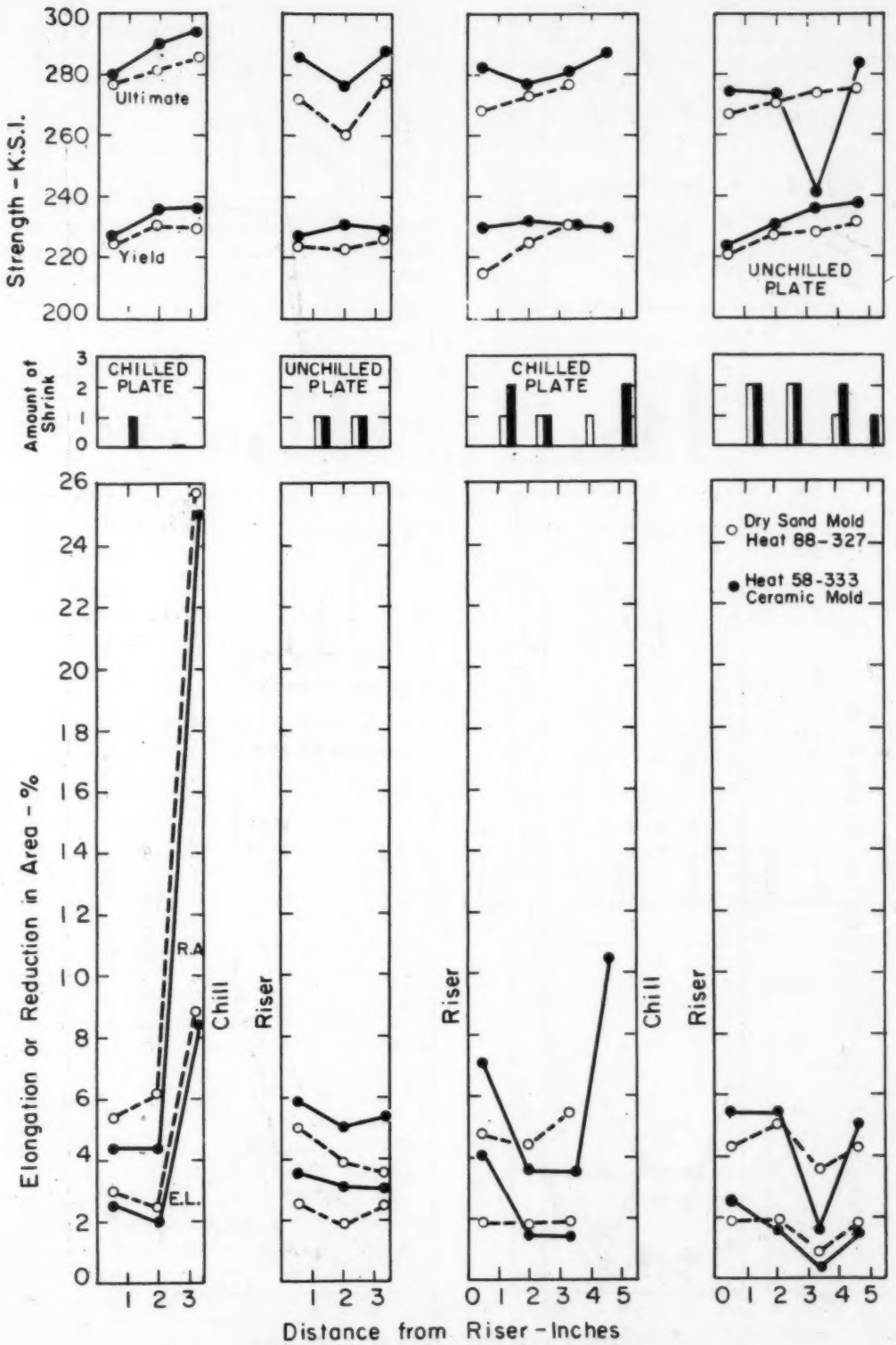


Fig. 6 — Feeding distance in 1 in. plates — 4 and 6 in. long. Test bars are 0.505 in. diameter.

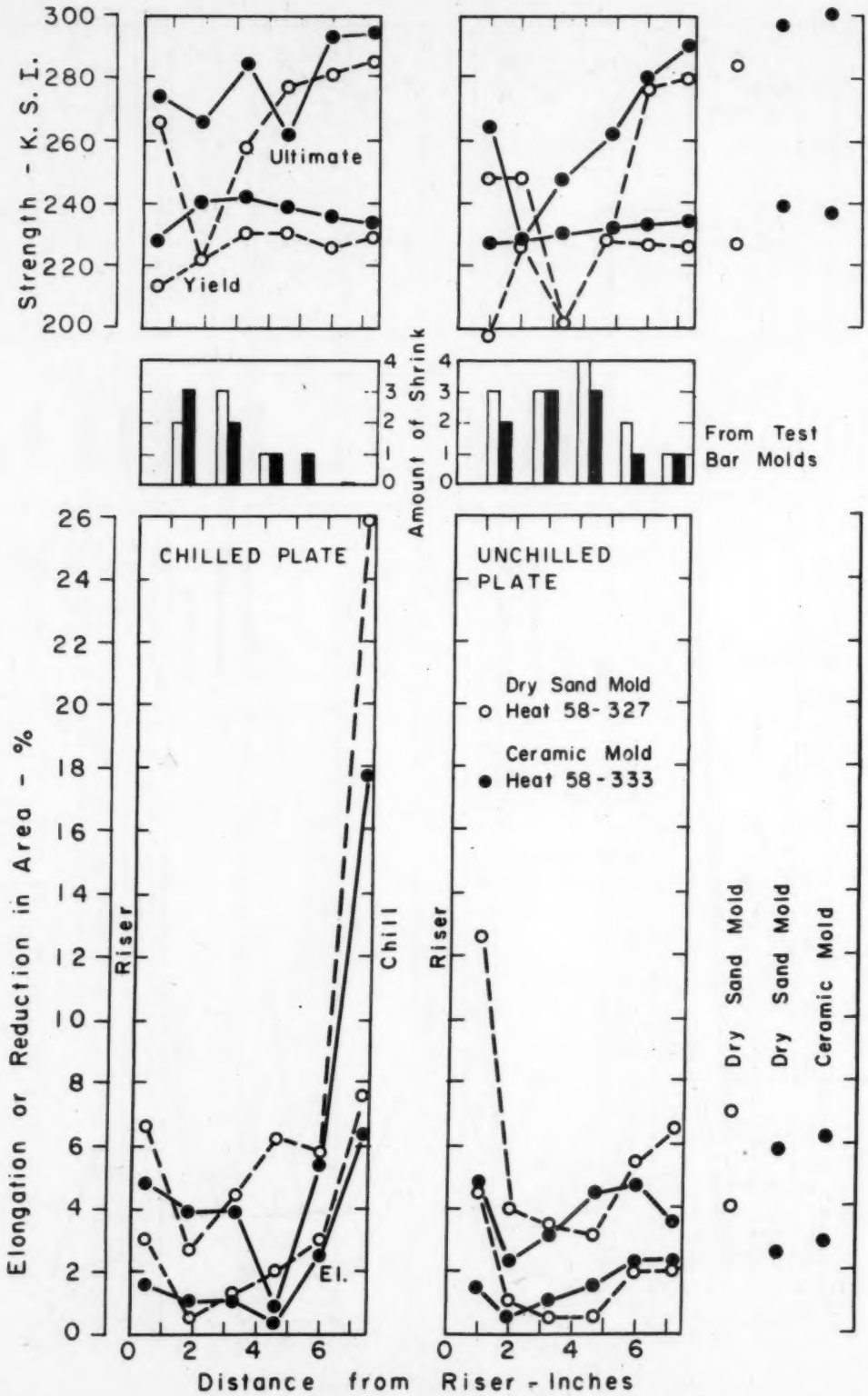


Fig. 7 — Feeding distance in 1 in. plates — 8 in. long. Test bars are 0.505 in. diameter.

bar from a plate 1 in. in thickness and 3 in. long had an elongation of 8.5 per cent at 295,000 psi tensile strength.

A quantitative description of the maximum length plate which can be cast sound depends upon an arbitrary decision as to what level of soundness or ductility is satisfactory. Table 2 shows the limits based on zero or 1 x-ray rating and another set of figures based on 2 per cent minimum elongation.

Two plates (1 x 6½ x 8 in.) from another heat of 8740 (58-368) were cast with end chills in sand molds. They were homogenized at 1850 F and oil quenched from 1575 F. One plate was tempered at 400 F and the other at 950 F. The plates were then sectioned into ⅛-in. slices which were ground into 0.020 x 0.875 x 6½ in. specimens. These were x-rayed on fine grained film and the radiographs examined for shrink at 20×. They were then pulled

TABLE 2 — FEEDING DISTANCE OF ½ AND 1 IN. PLATES

Thickness, in.	Mold Material	Unchilled Plate, in.	End Chilled Plate, in.
	Based on x-ray ratings of 1 or better		
1	Sand	4 (4T)	6 (2T+2)
1	Ceramic	4 (4T)	4 (4T)
½	Sand	2 (4T)	3 (4T+1)
½	Ceramic	2 (4T)	3 (4T+1)
	Based on 2% Minimum Elongation		
1	Sand	4 (4T)	6 *(4T+2)
1	Ceramic	4 (4T)	6 *(4T+2)
½	Sand	3 (6T)	3 (6T)
½	Ceramic	3 (6T)	4 (6T+1)

*Elongation limit relaxed somewhat since ductility in sound test bars only 2-3 per cent.

in the tensile machine. Tensile properties and radiograph results are presented in Fig. 8.

Again the properties and radiographs correlate quite well. Soundness was quite good to a distance of about 4 in. from the chill although ductility dropped off in this region. The chill probably has some benefit aside from inducing soundness; as cast grain may be smaller and microsegregation may be less in the immediate vicinity of the chill.

After the specimens were broken, the broken halves were radiographed. In every case fracture went through microshrink even when only one tiny spot was present. Most of the specimens which were completely sound broke at an angle of 45 degrees across the width of the specimen. Where unsoundness was present, the fracture was 90 degrees with respect to the length of the specimen.

Microshrinkage, although a major variable, is not the only property which changes along the length of the plate. Samples were cut from tensile specimens adjacent to the chill and the riser of the sand cast plate measuring 1 in. in thickness and 8 in. long (Heat 58-327). As shown in Fig. 9, the inclusions near the chill are small, globular and randomly distributed. Near the riser the inclusions are larger, irregular and have a tendency to be clustered.

Although in general the tensile properties correlate rather well with radiographs of thin slices, some apparent anomalies need explanation. The general characteristics of the shrinkage pattern are most apparent in Fig. 8 where a series of thin slices was cut from a 1 in. plate. Since standard tensiles were

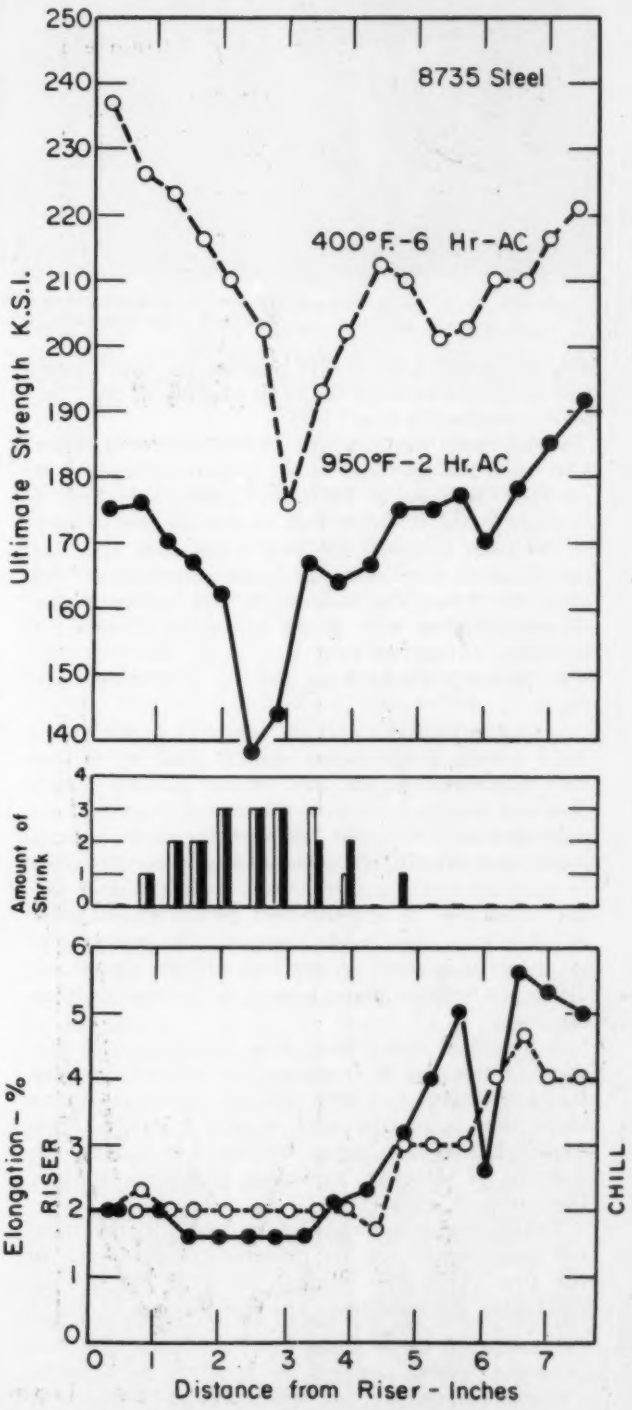


Fig. 8 — Properties in 1 in. plate — 8 in. long. Tensile specimens are 0.020 in. x 0.875 in. Heat treatment— 1800 F, 3 hr — AC; 1575 F, 2 hr — OQ. Heat no. 58-368.

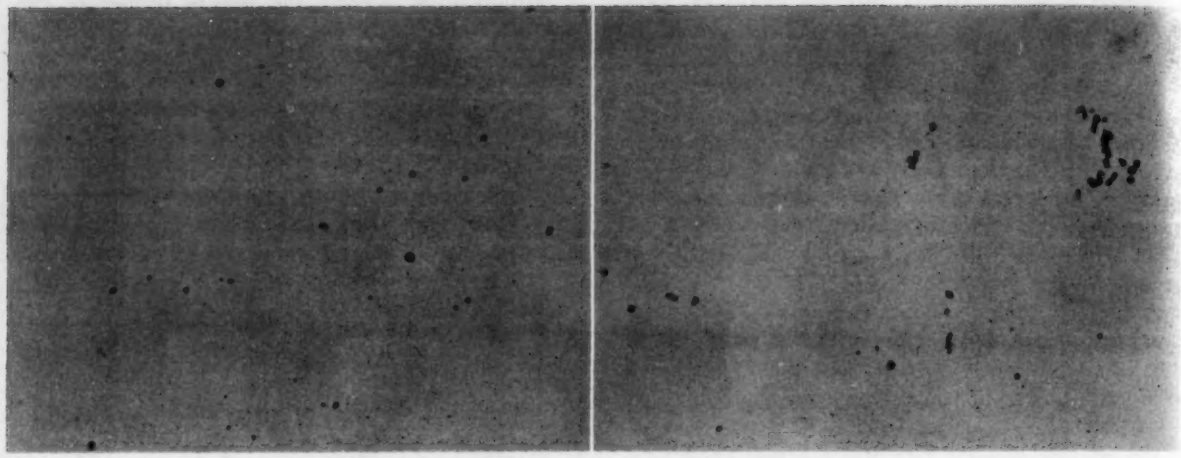


Fig. 9 — Micrographs of the chill and riser ends of plate 1 in. thick and 8 in. long. 250×. Left, chill end. Right, riser end.

not cut in this case, it was possible to obtain more and closer spaced slices than was possible in the 1 in. plates described in Figs. 5 and 7.

Sound metal was obtained over a distance of about 4 in. from the chill and in the immediate vicinity of the riser. In most of the cases described in Figs. 4 through 7, the ductility rose sharply in the vicinity of the riser, although the first x-ray slices were far enough from the riser that some unsoundness was apparent. Thus, the radiograph and tensile results do not correlate well in the immediate vicinity of the riser. Taking an x-ray slice before the first tensile specimen would have been a preferable technique.

It was surprising to obtain complete soundness in the 1 x 8 in. long ceramic molded plate by normal x-ray procedures (2 per cent sensitivity). This length of sound plate is greater than expected from Pellini's original work. Of course, his work was done in sand molds, and the difference in molding materials might be expected to cause a difference, especially since this particular type of ceramic mold gives a slower solidification rate than sand. However, the microradiographs of thin slices do not indicate any significant difference between plates poured in the two molding materials.

It should be noted that taking radiographs of thin slices in this way is a sensitive method of detecting traces of shrink. A 1 or 2 rating in Fig. 3 represents metal that is actually quite sound by normal techniques. However, as shown by Figs. 4 to 7, even this quantity of shrinkage can effect properties at these high strength levels.

Taking tensile specimens from plates in this manner can lead to unduly pessimistic conclusions. In the first place the soundest metal is removed in machining the specimen. The dotted lines in Fig. 3 show the diameter of the gage length compared to original thickness of the plate. It is obvious that the poorest metal is being tested. Also machining in this manner can put shrinkage defects at the surface of the tensile specimen where, according to observations, they have the most detrimental effect.

Finally, defects at the center may have little effect on actual performance of castings in service. In bending or torsion the presence of defects along the neutral axis would have only a minor influence. Also in fatigue type service, defects at the surface are the most important.

Although testing specimens from a casting may not give a reliable picture of quality of the casting as a structural component, these results indicate that slight traces of microshrinkage have a rather pronounced effect on the ductility of tensile specimens of cast ultra high strength steel.

REFERENCES

1. Hugo R. Larson and F. B. Herlihy, "Development of Low Alloy Steel Compositions Suitable for High Strength Steel Castings," AFS TRANSACTIONS, *vol. 67*, 1959.
2. E. T. Myskowski, H. F. Bishop and W. S. Pellini, "Application of Chills to Increasing the Feeding Range of Risers," AFS TRANSACTIONS, *vol. 60*, pp. 389-399, 1952.
3. H. F. Bishop, E. T. Myskowski and W. S. Pellini, "Contribution of Riser and Chill Edge Effects to the Soundness of Cast Steel Plates," AFS TRANSACTIONS, *vol. 58*, pp. 185-197, 1950.
4. H. F. Bishop, E. T. Myskowski and W. S. Pellini, "The Contribution of Riser and End Effects to the Soundness of Cast Steel Bars," AFS TRANSACTIONS, *vol. 59*, pp. 171-180, 1951.

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APPLICATION OF THEORY IN UNDERSTANDING FLUIDITY OF METALS

By J. E. Niesse, M. C. Flemings and H. F. Taylor

ABSTRACT

A theoretical analysis of fluidity of pure metals was carried out employing heat flow and fluid flow relationships. Equations were developed to predict fluidity of pure metals at their melting point and to predict the effect of superheat on fluidity.

An experimental investigation was conducted using the "vacuum fluidity test"; progress of metal flow was measured by high-speed motion picture photography. Fluidity tests were made on pure tin and tin-lead alloys; quantitative agreement between theory and experiment was obtained for pure tin.

Additions of lead to tin decrease fluidity markedly. Fluidity of a tin-0.01 per cent lead alloy was found to be approximately 70 cm; for a tin-0.10 per cent lead alloy it was 45 cm; and for a tin-1.0 per cent lead alloy it was 22 cm (all tests conducted at the liquidus temperature). In most other metals, small additions of alloys also decrease fluidity, and an explanation for this behavior is set forth herein. It appears that alloying elements which decrease fluidity do so by altering the mode of solidification of the pure metal.

A discussion is presented regarding the location in a fluidity spiral where solidification first chokes off liquid flow. It is concluded this solidification occurs either at the entrance to the flow channel, within the flow channel or at the tip of the flowing stream, depending (predictably) on alloy analysis and superheat.

Experiments were made on fluidity of aluminum-tin alloys and on low-carbon steel. The results obtained showed the understanding of fluidity gained from studying low melting metals was applicable to other systems. Low-carbon steel was found to have flow and solidification characteristics similar to certain of the tin-lead alloys.

INTRODUCTION

Metal "fluidity" (in the casting sense) has been defined as the property of a metal which allows it to flow freely and evenly into a mold cavity and fill it before such freezing occurs as would obstruct further flow.[1] The study of fluidity is, of course, of great practical importance, and many different tests have been developed to evaluate the various factors affecting fluidity. Almost all these tests involve one basic feature; metal is caused to flow into a mold channel of small cross-sectional area. The distance metal flows before being stopped by solidification is the measure of fluidity. Several excellent reviews of fluidity and fluidity testing may be found in the literature.[1-3]

The present work is primarily a study to determine the fundamental quantitative effects of such variables as metallurgy, fluid flow and heat flow upon fluidity. The work is, in a sense, a continuation and extension of a previous basic study in the M.I.T. foundry laboratory.[4,5] It deals with 1) an analysis of factors affecting fluidity of pure metals when heat flow rate varies with time (as in a sand mold), 2) an analysis of the effect of superheat on fluidity and 3) discussion of how alloying elements change the fluid flow characteristics and pure metals.

Experimental work, supporting the theoretical studies, has been carried out using a carefully controlled "vacuum fluidity test." Research has been devoted primarily to studies on pure tin and tin-lead alloys, but tests have also been made (using the same apparatus) on the fluidity of a low-carbon steel.

FLUIDITY OF PURE METALS

Figure 1 is a schematic diagram of metal flowing in a fluidity test piece. The metal travels down a sprue and into a thin tubular mold cavity "pushed" by the pressure head (Z_1-Z_e). There are a variety of fluid flow resistances in such a system which act to reduce the velocity of metal; these include internal friction, wall friction, etc. However, flow does not cease entirely until some section of the channel is sufficiently solid to prevent passage of further liquid.

Because it is solid metal that ultimately stops flow in a channel, rate of solidification is a factor which must be considered in any analysis of fluidity. Another equally important consideration is mode of solidification; i.e., the nature of solid grains which form and their location in the flowing stream. While much still remains to be learned about the effects of solidification variables on fluidity, several previous researches[4,6] have yielded an engineering understanding of how solidification of a pure metal takes place in a fluidity spiral. This is illustrated in Fig. 2 for a metal poured at its melting point. Subsequent

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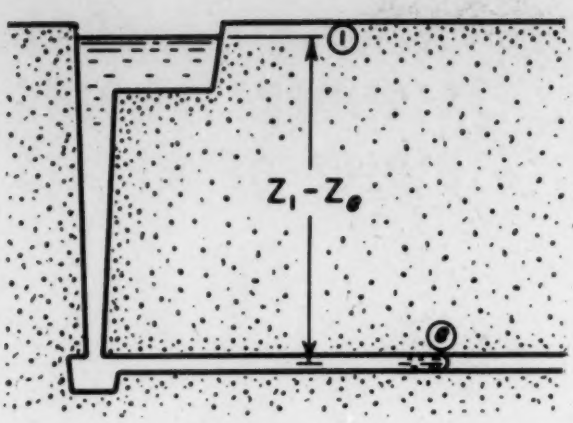


Fig. 1 — Schematic diagram of flow in a fluidity test.

analyses in this paper are based on the model of Fig. 2.

Pure Metal Solidification

As a pure metal with no superheat enters the fluidity test channel (Fig. 2), solidification must begin immediately, since heat flows from the metal into the mold. The absolutely pure metal freezes on the mold wall with a smooth liquid-solid interface, as shown in Fig. 2a. As it continues to flow downstream, solid continues to form and freeze along the wall; however, the location nearest the entrance to the flow channel began to freeze earliest, and it is here that complete solidification of the cross-section occurs first (Fig. 2c).

Certain assumptions are implicit in the above description, including conditions of radial heat flow and of no undercooling. Earlier works discuss these assumptions and present ample theoretical and experimental evidence that fluidity spirals of pure metals, poured without superheat, do in fact freeze first at the channel entrance.[4,6]

The effect of superheat on solidification of a pure metal in a fluidity spiral is sketched in Fig. 3 and discussed in detail later in this paper. Essentially, it has been concluded that superheat moves the point downstream where solid grains form and block the further flow of metal. The basic mechanism of solidification and flow stoppage, however, remains unchanged.

The following paragraphs present heat flow and fluid flow analyses which describe quantitatively the models of Figs. 2 and 3 for the various conditions expected to occur in sand molds and compare these analyses with experimental data.

Heat Flow

Consider a short length of a fluidity channel of circular cross-section wherein a pure metal is flowing and freezing (Figs. 1-3). Heat, flowing from metal to mold, must traverse 1) liquid metal to the solid-liquid interface, 2) solid-liquid interface, 3) solid metal, 4) metal-mold interface and 5) mold. Each of the preceding zones offers a barrier to heat flow, but for the various conditions encountered in ordinary fluidity tests only the last two are important.

This is so because of the high thermal conductivity of metals, the fact that much of the heat originates from heat of fusion and does not pass through the liquid and the fact that practically no specific heat is lost from the solid metal. The latter can easily be

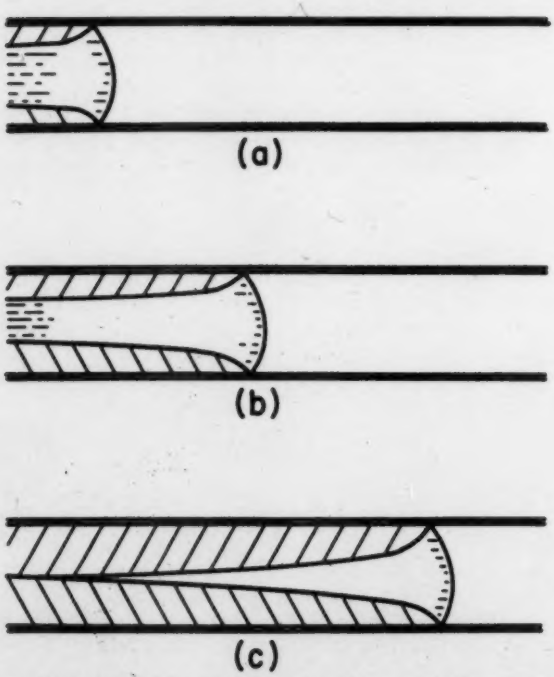


Fig. 2 — Schematic diagram of flow and solidification of a pure metal in a fluidity channel (no superheat). (a) Liquid metal enters flow channel; columnar grain formation (with smooth liquid-solid interface) begins. (b) Columnar grains continue to grow. (c) Choking off and flow cessation occurs.

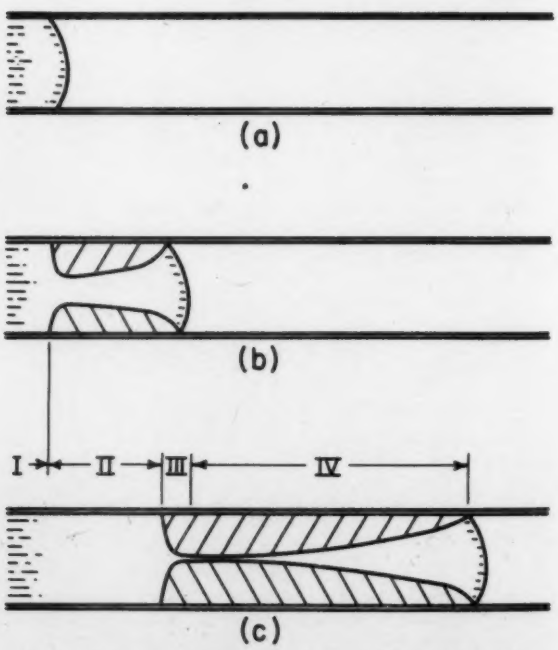


Fig. 3 — Schematic diagram of flow and solidification of a pure metal in a fluidity channel (with superheat). (a) Liquid enters flow channel and dissipates superheat before solidification begins. (b) Solidification begins by growth of columnar grains. (c) Choking off occurs.

shown by use of Adams' analysis of solidification in sand molds.[7]

In the following analysis, resistances due to the metal-mold interface and to the mold itself have been considered. Heat flow rate in the mold has been assumed to be inversely proportional to the square root of time (θ type) as is found in sand molds. The equations developed by Adams[7] apply and have been used herein. Figure 4 shows schematically the temperature distribution across a radial cross-section of a solidifying fluidity spiral, and equation (1) below is the resulting analytical expression describing heat flow as a function of time (Appendix A gives details).

$$Q = \frac{2\pi R h (T - T_r)}{B}\left[\theta_h + 2A\left(1 - \frac{1}{B}\right)\sqrt{\theta_h} - \frac{2A^2}{B}\left(1 - \frac{1}{B}\right)\ln\left(1 + \frac{B}{A}\sqrt{\theta_h}\right)\right] \qquad (1)$$

Symbols used in this and in subsequent equations are defined in Appendix B.

A typical solution of equation (1) is shown graphically in Fig. 5. Heat flow rate, dependent on mold resistance as well as interface resistance, decreases with time. Equation (1) applies to mold channels of circular cross-section in sand, although similar procedures may be used for analyzing other types of molds.

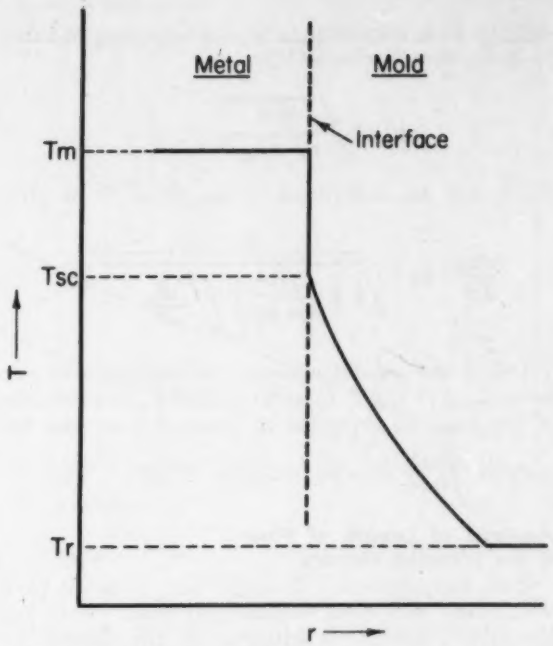


Fig. 4 — Schematic temperature distribution in metal and mold. Temperature distribution is across a radial cross-section of the fluidity spiral and into the sand or glass mold.

Fluid Flow

To arrive at an analytical expression for fluidity it is necessary to determine fluid flow as well as heat flow characteristics. Heat flow analysis allows one to predict the rate and extent of solidification, but the incorporation of fluid flow variables is essential to determine how far the liquid metal travels before solidification imposes enough frictional resistance to stop the flow.

Consider fluid flow in the case of a pure metal flowing and freezing with a smooth solid-liquid interface. An energy balance between the leading tip of the stream (e), Fig. 1, and the free surface of liquid metal (1) supplying the fluidity channel yields:

$$2gZ = V_e^2(1 + \phi_{ent}) + \int_0^{L_e} f V^2 \frac{dL}{2r} \qquad (2)$$

where the wall friction energy is integrated over the length of the flow in the last term of equation (2).

"Z" is the "head of metal" given by:

$$Z \equiv Z_1 - Z_e + \frac{P_1 - P_e}{\rho} \qquad (3)$$

This term, Z should include effects, if any, of surface tension. However, due to the uncertainty of data, especially for alloys, these effects are neglected. The result of surface tension would be a simple decrease in Z and would have little effect on the remainder of this analysis.

Conservation of matter requires that:

$$\pi r^2 V = \pi R^2 V_e$$

$$V = \frac{R^2}{r^2} V_e \qquad (4)$$

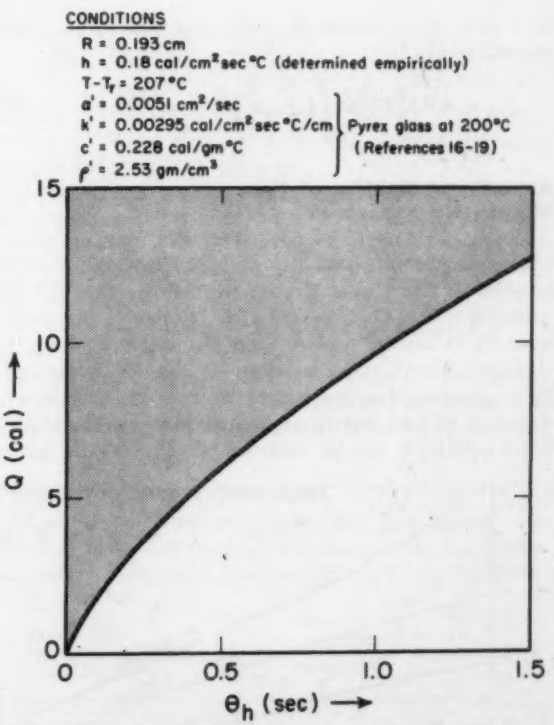


Fig. 5 — Typical heat flow curves for a cylinder mold.

which, substituted in the last term of equation (2), gives:

$$V_e = \sqrt{\frac{2gZ}{1 + \phi_{ent}}} \frac{1}{\sqrt{1 + \frac{fR^4}{2(1+\phi_{ent})}\int_0^{L_e} \frac{dL}{r^5}}} \qquad (5)$$

When L_e is zero (that is, at the beginning of flow) V_e is the initial velocity V_i:

$$V_i = \sqrt{\frac{2gZ}{1+\phi_{ent}}} \qquad (6)$$

which may be substituted in equation (5) to give:

$$\frac{dL_e}{d\theta} = V_e = \frac{V_i}{\sqrt{1+\frac{fR^4}{2(1+\phi_{ent})}\int_0^{L_e}\frac{dL}{r^5}}} \qquad (7)$$

This is the general equation for fluidity with variables L_e, L, r and θ. In order to find L_e as a function of θ, r must be expressed in terms of L so that the integral $\int \frac{dL}{r^5}$ may be evaluated.

Analysis of Length of Flow If No Freezing Occurs

If no freezing occurs, as in the case of initial parts of fluidity tests when superheated metal (pure and alloys) is poured, r is equal to R, the channel radius. Then equation (7) becomes

$$\frac{dL_e}{d\theta} = V_e = \frac{V_i}{\sqrt{1+\frac{f}{2(1+\phi_{ent})R}L_e}} \qquad (8)$$

in which the variables L_e and θ are separable. Integration yields:

$$L_e = \frac{2(1+\phi_{ent})R}{f}\left[\left(\frac{3}{4}\frac{fV_i\theta}{(1+\phi_{ent})R}+1\right)^{2/3}-1\right] \qquad (9)$$

Analysis of Fluidity of Pure Metals, No Superheat

When pure metal is poured at the melting point, freezing takes place on the mold walls. The amount of solid is controlled by the heat flow; that is, the value of r may be found if θ_h is known. Using equation (1) to find Q and r, then numerically summing elements of $\Delta L : r^5$, equation (7) can be evaluated. By repeating this numerical process starting at V_e equal to V_i and continuing until the stream velocity is zero, fluidity can be determined. The Table shows the initial conditions of such a numerical process. Figure 6 shows the shape of the solid-liquid interface, and Fig. 7 the calculated variation of stream velocity V_e as flow progresses.

From analysis of Fig. 6, r is a parabolic function of L of the type

$$r = R - K\sqrt{L_e - L} \qquad (10)$$

where K is a function of V_e. For the curves of Fig. 6:

$$K = \sqrt{\frac{0.0045}{V_e - 4} - 0.00002} \qquad (11)$$

Unfortunately, this parabolic relation does not hold as the velocity approaches zero. The relationship of equation (11) is indeterminate when V_e is 4 cm/sec. For the conditions of Fig. 6, excellent agreement is obtained between the point solutions of equation (10), and curves obtained by numerical analysis until the elapsed time θ is over 1.06 sec (or until V_e is about 15 cm/sec). This is illustrated by the points plotted on Fig. 6.

CONSTANTS EMPLOYED FOR SOLUTION OF EQUATION (12)*

Conditions:

$R = 0.193$ cm	$\rho' = 2.53$ gm/cm³ (pyrex)[19]
$h = 0.18$ cal/cm² C sec	$\rho H = 106$ cal/cm³ (tin)
$c' = 0.228$ cal/gm C[17,18]	$f = 0.04$
$k' = 0.00295$ cal/cm² sec C/cm[16]	$\phi_{ent} = 0.95$
$V_i = 180$ cm/sec	$\Delta L = 2$ cm
	$T\text{-}T_r = 207$ C

*Solution drawn in Fig. 9.

Substituting equation (10) into (7) eliminates the variable r; integration yields:

$$\frac{dL_e}{d\theta} = V_e = \frac{V_i}{\sqrt{1+\frac{fR}{12K^2(1+\phi_{ent})}\left(1+\frac{4\frac{K}{R}\sqrt{L_e}-1}{\left(1-\frac{K}{R}\sqrt{L_e}\right)^4}\right)}} \qquad (12)$$

Equation (12) can be solved numerically from $V_e = V_i$ to $V_e = 15$ cm/sec. At the latter point the

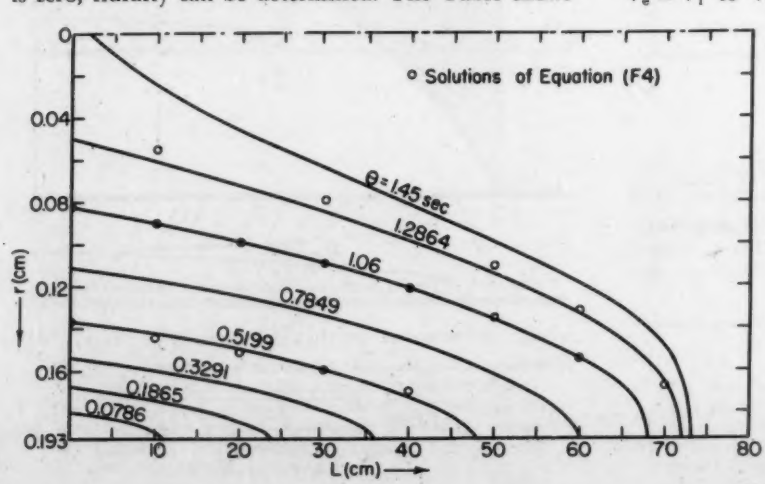


Fig. 6 — Solid-liquid interfaces for V_i = 180 cm/sec. Curves calculated for pure tin poured at its melting point into a fluidity test channel 0.386 cm in diameter.

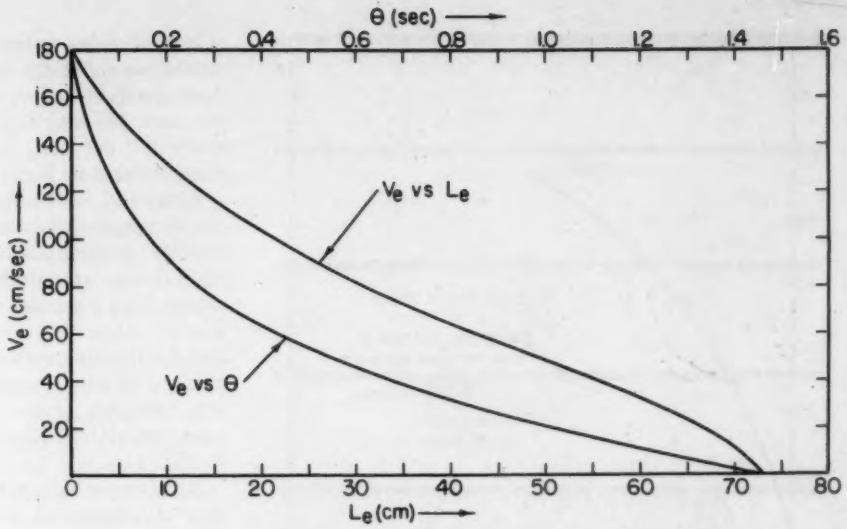


Fig. 7 — Stream velocity as a function of time θ and length L_e. Curves calculated for pure tin poured at its melting point into a mold 0.386 cm in diameter.

metal stream has completed about 95 per cent of its total flow length.

Superheat Effect on Fluidity of Pure Metals

Figure 3 illustrates qualitatively the effect of superheat on fluidity of a pure metal. Essentially, superheat prevents solidification from occurring near the entrance to the flow channel so that choking off of metal flow occurs further downstream and at a later time. Experimental evidence indicates this does, in fact, occur.[6]

Employing the assumptions made earlier in this paper, it is possible to describe in more detail the effects of superheat and to derive a mathematical expression covering a particular case. The assumptions that liquid metal cannot support a significant temperature gradient and that superheated liquid and solid metal cannot coexist for any length of time are employed. This appears to be realistic since flow of the liquid metal is highly turbulent and heat diffusivity of metals is high.

During flow of a superheated metal in a fluidity spiral there exists as many as four distinct zones (Fig. 3c):

I) A zone where no freezing occurs.
II) A zone where freezing once occurred but where superheated metal has caused re-melting so that no solid remains.
III) A zone where solid exists but remelting is occurring.
IV) A zone where solidification is progressing.

When the first metal enters a fluidity channel, it flows without freezing until its superheat is dissipated; the length to the point where solidification begins constitutes zone I. Subsequent small sections of metal which travel down the spiral do not dissipate their heat so rapidly since rate of heat flow into the mold decreases with time at any one location. The small sections of liquid arrive at the beginning of the partially solid zone with some superheat remaining; they therefore remelt this solid in the region termed zone III. At any instant of time the length that has been completely remelted is zone II. In zone IV, liquid and solid coexist (both at the melting point) and solidification takes place exactly as in the case of metal poured at its melting point.

On the assumptions that heat flow is radial and that thermal gradients in the liquid are small, zone III must be of negligible length. Then, with zones I and II identical from a friction-loss standpoint, equation (7) becomes

$$\frac{dL_e}{d\theta} = V_e = \frac{V_i}{\sqrt{1+\frac{fR^4}{2(1+\phi_{ent})}\left[\frac{L_1}{R^5}+\int_{L_1}^{L_e}\frac{dL}{r^5}\right]}} \qquad (13)$$

Where L_1 is the distance from the channel entrance to the beginning of solid metal.

Using equation (10), r may be eliminated and the friction integral evaluated:

$$\frac{dL_e}{d\theta} = V_e = \frac{V_i}{\sqrt{1+\frac{fR^4}{2(1+\phi_{ent})}\left[\frac{L_1}{R^5}+\frac{1}{6K^2R^3}+\frac{4K\sqrt{L_e-L_1}-R}{6K^2(R-K\sqrt{L_e-L_1})^4}\right]}} \qquad (14)$$

Unfortunately, to evaluate equation (14) a laborious numerical procedure is required. Equation (13) must first be solved to determine how L_1 and K are related to θ, L_e and V_e.

Comparison With Experiment

Experimental data for fluidity of a pure metal were obtained using the vacuum fluidity test developed by Ragone, Adams and Taylor.[5] In this test (Fig. 8)

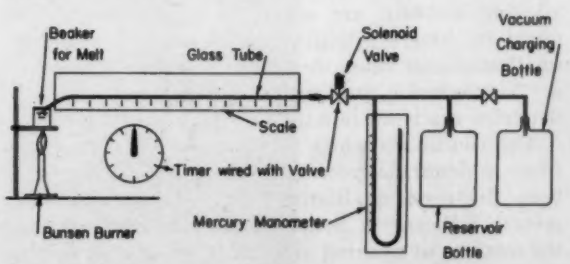


Fig. 8 — Schematic sketch of vacuum fluidity apparatus.

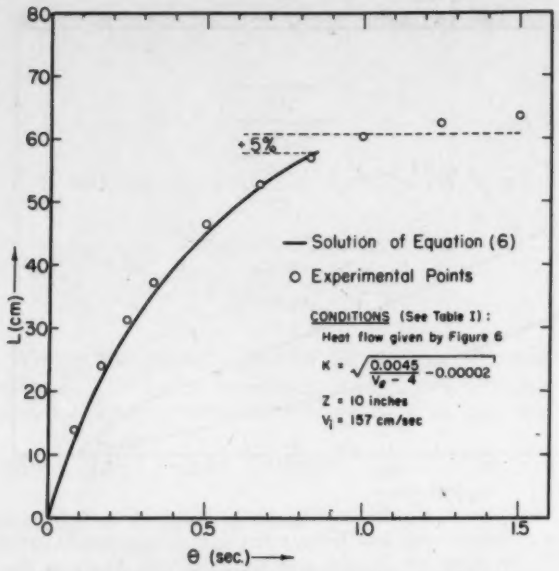


Fig. 9 — Curves of fluidity L vs. time θ comparing analytical, equation (6), to experimental results. Experimental results obtained for pure tin poured at its melting point into a test channel 0.386 cm in diameter, and under 10 in. of metal head.

liquid metal is drawn into a glass tube by a partial vacuum. Advantages of the test compared to other fluidity tests include 1) close control over many mold and fluid flow variables, 2) ability to perform many tests quickly and reproducibly and 3) opportunity to photograph metal flow with a high-speed motion picture camera. In addition, heat flow characteristics of the glass mold are quite similar to those of a sand mold.

Figure 9 illustrates experimental data for fluidity of highly pure tin at its melting point compared with a calculated curve. The calculated curve is based on equation (12) using the constants shown in the Table, except that initial velocity is 157 cm/sec corresponding to a metal head Z of 10 in. Comparison between calculated and experimental values is good up to the limitation of equation (12); this limitation is at approximately 95 per cent of the total length.

FLUIDITY OF ALLOYS

General Discussion

In the absence of undercooling effects, absolutely pure metals freeze with a smooth "plane front" interface between the solid and liquid. This mechanism of solidification is illustrated in Figs. 2 and 3. When alloying elements are added to a pure metal, the metal no longer solidifies with a smooth interface; small additions cause the front to become "rippled," or "corrugated," and with slightly larger additions dendrites reach out into the flowing stream.

The conditions which cause transition from plane front to dendritic growth in solidifying castings have been discussed qualitatively and quantitatively by several authors.[8-10] Briefly, these conditions include the amount of alloying element, type of alloy system, characteristics of alloying element and thermal gradients in the casting.

In most alloy systems small amounts of alloying elements are sufficient to bring about dendritic growth during solidification. In some systems, less than 0.0 per cent by weight is adequate, and in the present study 0.1 per cent lead dissolved in tin was more than enough to break the plane front.

Figure 10 illustrates the effect that small amounts of alloying elements have on solidification in a fluidity casting. Solidification and flow stoppage mechanisms are similar to that illustrated in Fig. 2, except that here the solidification front is no longer smooth. Dendrites project into the flowing stream, and friction is much greater than it would be in the absence of such projections. Small amounts of alloying elements almost always decrease fluidity of a pure metal by affecting solidification, as illustrated in Fig. 11.

In alloys which contain somewhat higher percentages of alloying element than that illustrated in Fig. 10, dendrites may reach even further into the stream and create such high friction that metal flow ceases when only a small fraction of any cross-section is completely solid. Based on this assumption (that flow ceases after only a negligibly small amount of solidification takes place) Rightmire and Taylor[11] have analyzed and successfully correlated a large amount of experimental data on fluidity of cast steels.

Equiaxed Dendritic Solidification

There is another type of solidification which is known to occur in fluidity test castings. Alloys which contain a substantial percentage of solute, and in which nucleation is not a problem, solidify with equiaxed dendrites.[6,12] As the equiaxed grains form and

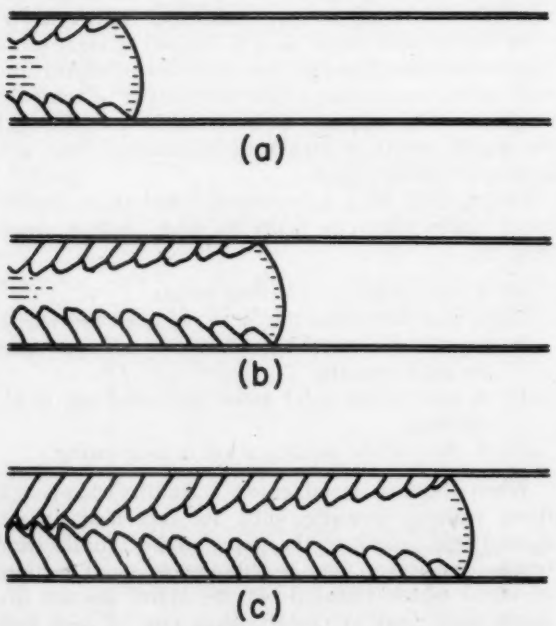


Fig. 10 — Schematic diagram of flow and solidification of a dilute alloy in a fluidity test channel (no superheat). (a) Liquid enters flow channel; columnar grain formation with jagged liquid-solid interface begins. (b) Columnar grains continue to grow. (c) Choking off occurs at flow channel entrance, cross-section is not completely solid but is sufficiently solid to prevent flow.

grow, they flow downstream until frictional effects become large enough to stop flow. The degree of this friction has been shown to depend on the size and shape of the solid particles as well as on the amount of solid.[12]

When the solid is growing in dendrite-like shapes, small percentages of solid can effect flow stoppage; Fig. 11 illustrates this mechanism schematically. Fine grains form at the tip and flow downstream. Metal flow is eventually choked off at the tip of the stream when frictional effects due to the solid grains become sufficiently high.

Figures 12 and 13 present macrostructures taken from fluidity spirals of two aluminum-tin alloys; these structures illustrate the two mechanisms of solidification described above. In Fig. 12 (aluminum containing 0.1 per cent impurities) columnar dendrites formed and grew on the mold wall. These dendrites choked off flow at a point near the entrance of the fluidity spiral. In Fig. 13, illustrating the macrostructure of an aluminum-10.0 per cent tin alloy, fine grains formed and traveled downstream with the flowing metal. More detailed discussion of flow and solidification of these alloys can be found in the original sources from which Figs. 12 and 13 were abstracted.[6,13]

Earlier in this paper a model was presented which described solidification of a pure metal in a fluidity test casting (Figs. 2 and 3). This model lent itself to

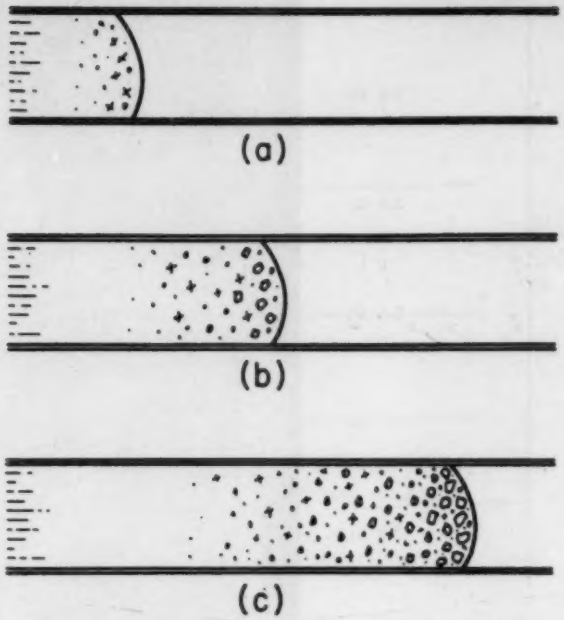


Fig. 11 — Schematic diagram of flow and solidification of an alloy which readily nucleates fine grains (no superheat). (a) Liquid enters flow channel, fine grains nucleate at the tip. (b) Nucleation continues and fine grains grow rapidly as flow progresses. (c) Flow ceases when a critical concentration of solid grains is attained at the tip.

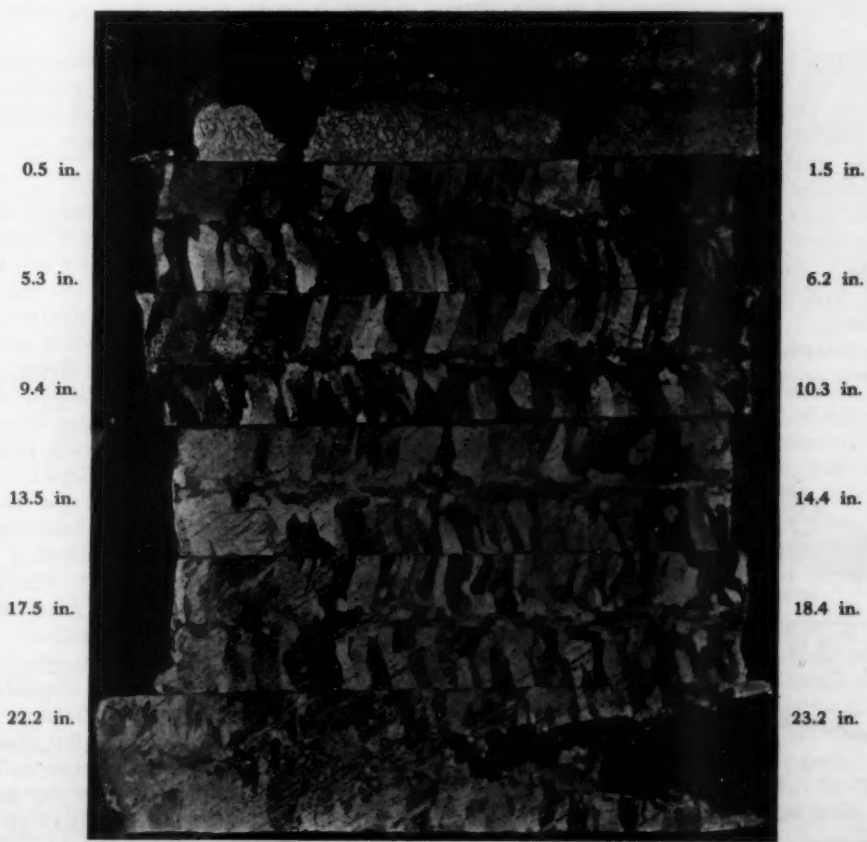


Fig. 12 — Longitudinal macrostructures of an aluminum (99.99 per cent) fluidity casting. Temperature, liquidus plus 150 F (83 C); metal flow left to right. Figures at sides represent distance along the tube.[6,13] 4×.

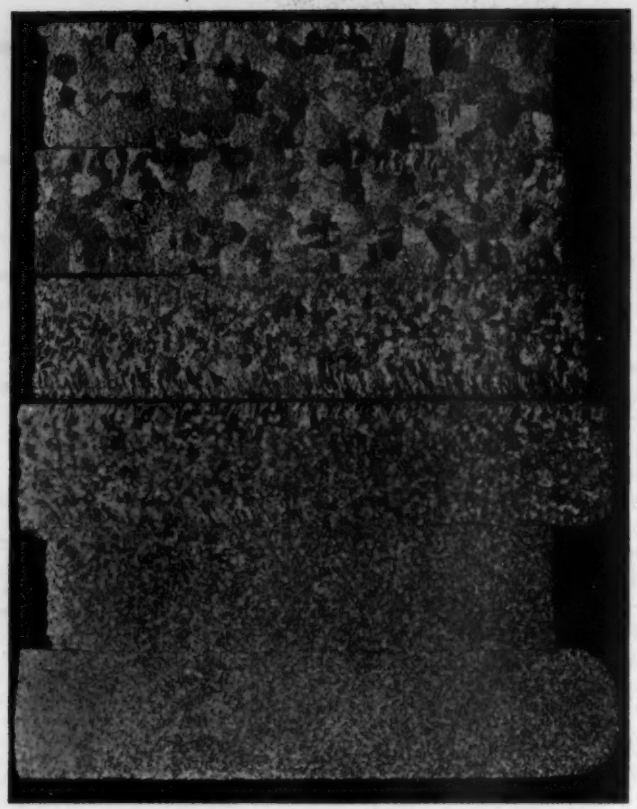


Fig. 13 — Longitudinal macrostructures from an aluminum-5.0 per cent tin fluidity casting. Temperature, liquidus plus 147 F (82 C); metal flow from left to right. Figures at sides represent distance along the tube.[6,13] 4×.

mathematical analysis, and equations were developed from this analysis for fluidity of pure metals. The models which have been drawn for alloys (Figs. 10 and 11) do not lend themselves as readily to mathematical treatment primarily because friction effects are difficult to evaluate. This is so because it is difficult to determine quantitatively 1) total flow area available between dendrites which adhere to the mold wall, 2) roughness of these dendrites and 3) increased friction resulting from fine grains which flow downstream with the metal.

Therefore, quantitative analysis of the fluidity of alloys has not been undertaken in this work. The remainder of this paper presents experimental results illustrating the effect of alloying on fluidity and fluid flow and illustrating the applicability of models for understanding the flow behavior of alloys.

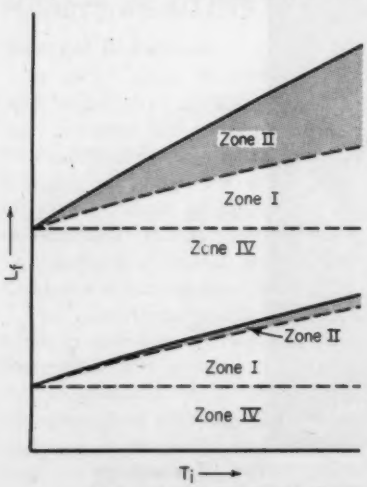


Fig. 14 — Schematic curves showing increased influence of zone II with increased fluidity.

Superheat Effect

Because the mechanism of freezing cannot be put in fully quantitative terms, only a rough analysis of the effect of superheat on fluidity can be made. Referring again to Fig. 3, fluidity test castings are considered to be made up of four distinct zones. These zones apply to alloys which solidify, as illustrated in Figs. 10 and 11, as well as to pure metals.

For both pure metals and alloys, the extent of zone I depends primarily on degree of superheat. The extent of zone II, however, depends primarily on the length of time during which liquid flows through zone I; this time is, of course, the fluid life. As a result, superheat is effective in increasing fluidity of alloys which are already highly fluid at their melting point (or liquidus temperature). The superheat increases zone I, but substantially increases zone II because of the large amount of remelting.

Conversely, superheat is not as effective in increasing fluidity of alloys which are sluggish at their liquidus; the superheat may increase zone I as much as in the case of the highly fluid alloy, but the increase of zone II is small. This situation is depicted schematically in Fig. 14; the greater the initial fluidity, the greater the effect of superheat on the fluidity. The effect has been observed by other investigators[5,6,14,15] but has not been previously explained;

experimental results showing the effect are also presented later in this paper (Fig. 16).

Experimental Results and Discussion

The vacuum fluidity test (Fig. 8) was used for all experimental work described herein. Work was devoted primarily to study of tin-lead alloys, but limited data are also presented for an aluminum-tin alloy and for a low-carbon steel.

Figure 15 illustrates the effect of alloy additions on metal fluidity. The data are for lead added to tin, and are plotted on a logarithmic scale to more clearly illustrate the effects of small additions. Additions up to about 0.07 per cent lead had little effect on fluidity (length of flow) or on fluid life (duration of flow as measured by motion picture photography). Additions over about 0.07 per cent lead sharply decreased fluidity.

It appears that there is a critical purity affecting fluidity; metals more pure than this critical composition are not much more fluid, but less pure metals are much less fluid. The explanation for this behavior lies in the fact that a change in the mode of freezing due to alloying occurs at a critical composition which depends also on the particular conditions of the test. The change in the mode of freezing in these nearly-pure metals is a change from plane front (typical of pure metals where the solid-liquid interface is an isotherm) to a dendritic type of freezing.

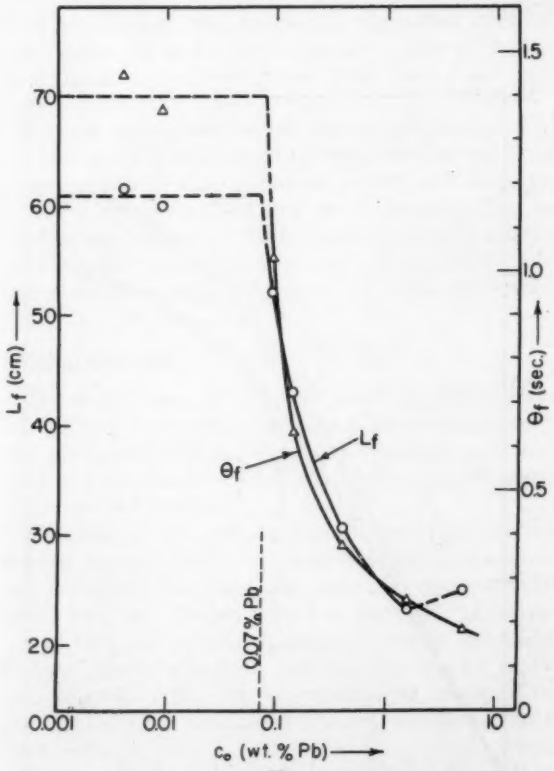


Fig. 15 — Fluidity and fluid life as a function of log composition of weight per cent lead in pure tin. In practice, several tests were conducted at temperatures slightly above the liquidus and results extrapolated to the liquidus for presentation. All runs were conducted in glass tubes 0.386 cm in diameter under a metal head of 10 in.

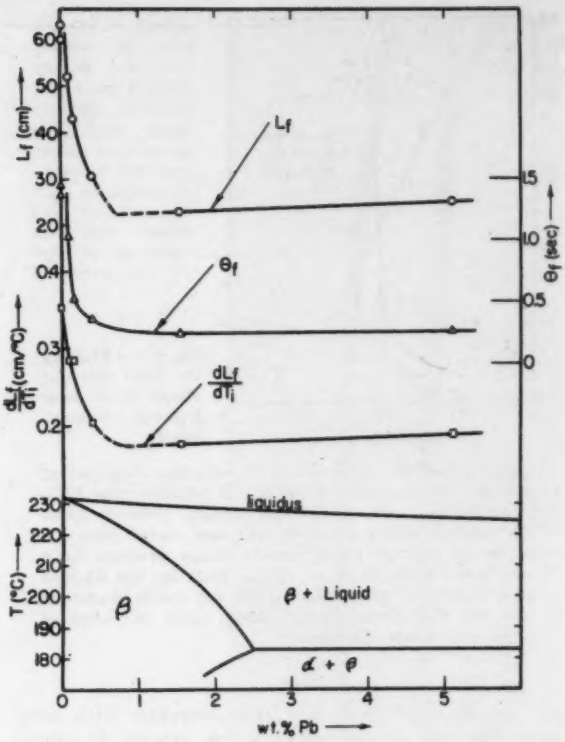


Fig. 16 — Fluidity, fluid life and slope of fluidity-temperature curve as a function of composition of weight per cent lead in pure tin at the liquidus temperature.

Figure 16 presents the fluidity data of Fig. 15 in more conventional fashion. Also shown are 1) phase diagram for the alloy system, 2) fluid life (θ_f) of the alloys studied and 3) the slope ($dLf:dt_i$) of the fluidity vs. superheat curves for the alloys. The shape of the fluid life curve is similar to that of the fluidity curve. The temperature coefficient of fluidity ($dLf:dt_i$) decreases with decreasing fluidity for reasons outlined, shown schematically in Fig. 14.

Solidification Mechanisms

Two quite different mechanisms have been discussed whereby solidification of alloys may occur in a fluidity test. In both cases growth is dendritic, but in one the solid which forms adheres to the mold walls, in the other it moves downstream. The mechanism of solidification which actually occurs for a given alloy is of importance in determining certain fluidity data, particularly fluid life as described in the following paragraphs.

If solid metal adheres to the mold walls, then fluid life (extrapolated to the melting point) should remain the same regardless of metal head, This is so because in the absence of superheat, rate of solidification and hence, fluid life are dependent only on mold and metal physical constants, not on amount of liquid flowing. For alloys in which the solid travels downstream (Fig. 11) fluid life should decrease with increasing metal head (even though fluidity increases). This apparent anomaly is explained by

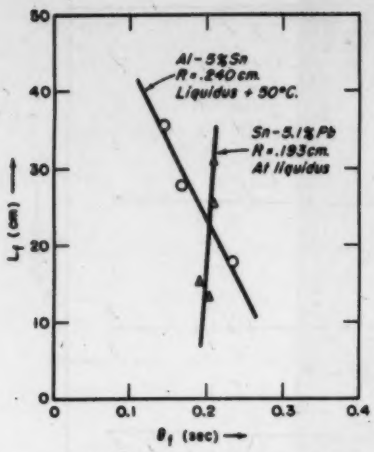


Fig. 17 — Fluidity vs. fluid life for alloys at or near liquidus temperature.

Data for lead-tin alloys are at the liquidus temperature; data for aluminum-tin alloys are at liquidus plus 50 C. Fluidity (L_f) was varied by varying pressure head. For lead-tin alloys pressure head was varied from 3.5 in. to 20 in.; for aluminum-tin alloys pressure head was varied from 10 in. to 100 in. Data for the tin-lead alloys apply to glass tubes 0.386 cm inside diameter; data for the aluminum-tin alloys apply to tubes of 0.480 cm inside diameter.

the fact that the heat flow rate decreases with time, and thus the tip of the flowing stream is always exposed to the highest cooling rate or chill. Increasing the flow velocity increases the chill which the tip experiences so the stream will freeze earlier even though it flows farther.

In Fig. 17, fluidity is plotted vs. fluid life for a tin-5.1 per cent lead alloy and for an aluminum-tin alloy. Data are for constant pouring temperature; fluidity (L_f) was varied by varying the pressure head.

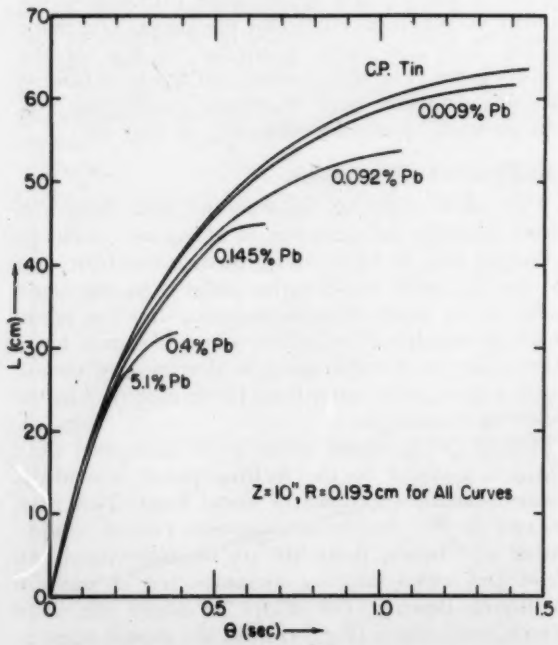


Fig. 18 — Distance L vs. time θ for tin-lead alloys at 469 F (240 C) pouring temperature.

Fluid life (θ_f) was determined by motion picture photography. Fluid life of the lead-tin alloy does not vary appreciably over the range of different metal heads employed. This behavior indicates that solidification occurred near the flow channel entrance by a mechanism similar to that of Fig. 10.

By employing the heat flow relation, equation (1), and the fluid life from Fig. 17, it was calculated that only 25 per cent of the cross-section at the flow channel entrance was solid at the end of the flow. Because this solid was in the form of a network of dendrites this small amount was able to effect flow stoppage.

The data for fluidity vs. fluid life of aluminum-5 per cent tin alloy are included in Fig. 17 for comparison. Here, fluid life decreases with increasing metal head indicating the solidification occurred by a different mechanism, typical of that illustrated in Fig. 11.

Lead-Tin Alloy Fluidity

Figure 18 shows the fluidity runs of a series of lead-tin alloys. Significant is the difference in character of the curves for the different alloys. Pure tin gradually slows down as the solid metal grows into the liquid. For 5.1 per cent lead, the flow velocity is essentially constant until flow stoppage occurs quite suddenly. The sudden stop indicates a rapid increase in friction near the end of flow due to growth of mesh-like dendrites.

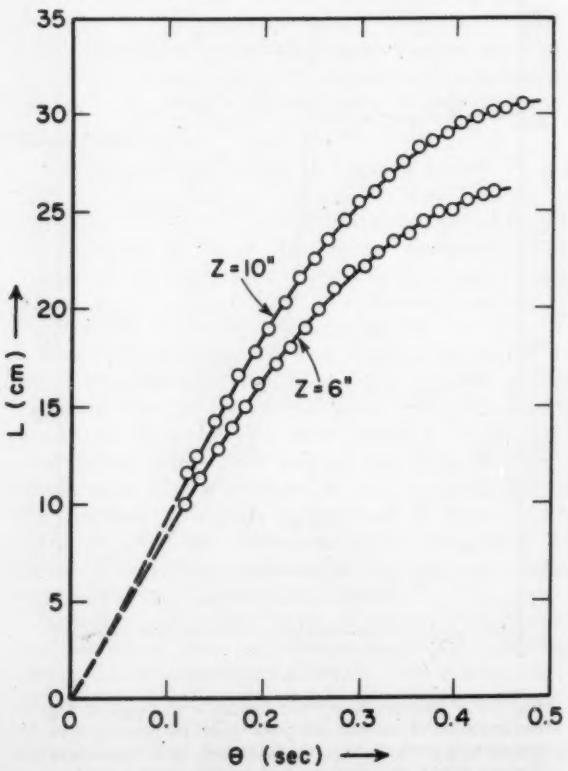


Fig. 19 — Distance vs. time curves for two fluidity tests of low carbon steel at 2837 F (1558 C). Test conducted with glass tubes 0.508 cm inside diameter.

Figure 19 is a similar recording of distance vs. time curves for two fluidity tests on low-carbon steel (0.25%, 0.43% Si, 0.6% Mn, 0.010% P, 0.04% S). The curves gradually slope off indicating a continuous increase of friction such as would be expected in a pure metal or dilute alloy. Even though the metal head was increased from 6 to 10 in., the fluid life changed only from 0.44 to 0.47 sec indicating a mechanism wherein flow stoppage occurs by solid adhering to the mold wall near the entrance to the flow channel (Fig. 10). That this was, in fact, the case was substantiated by radiographs of several of the fluidity test castings, Fig. 20.

The leading tips of the castings exhibited pipes, and the radiographs show that centerline shrinkage extended from the tip back to a location near the flow channel entrance. It was, therefore, at a point near the entrance that choking off of metal occurred; this point was slightly downstream from the entrance because of some superheat. It is of interest to note that the *V* pattern of shrinkage indicates some metal feeding took place (after flow stoppage) in a direction the reverse of original metal flow.

The preceding brief investigation of fluidity of low-carbon steel was conducted primarily to establish that theoretical studies and experiments utilizing low melting alloys yielded information which would facilitate understanding fluidity of higher melting, more widely used alloys. This objective has been realized; for example, the fluidity of low-carbon steel has been shown to be analagous in many ways to fluidity of a tin alloy containing the order of 0.1 per cent lead.

It is the expectation of the authors that the gradual acquisition of fundamental information pertaining to fluidity, such as that contained herein, will someday enable close control of the many factors affecting fluidity and fluid flow. Such control in casting quality should make it possible to produce intricate thin section castings of types that cannot be successfully cast today.

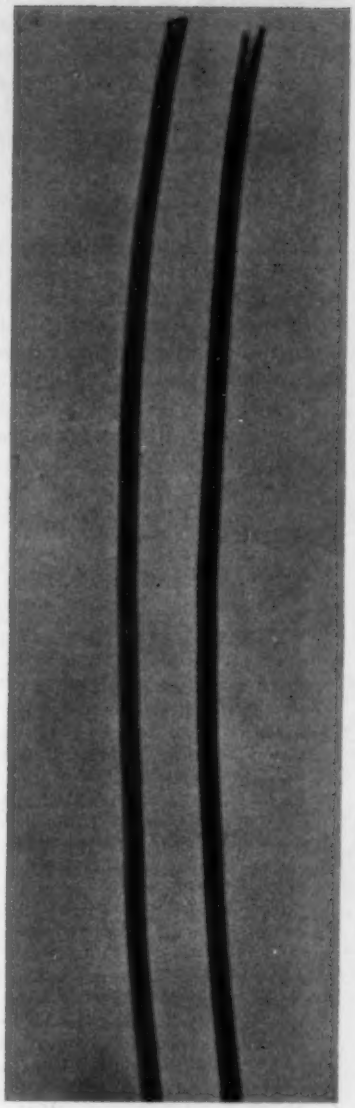


Fig. 20 — Radiographs of two typical low carbon steel fluidity castings, actual size. Note "pipe" at the leading tip (top) of the casting, with centerline shrinkage extending nearly back to the channel entrance.

CONCLUSIONS

The importance of heat flow and fluid flow factors on fluidity of metals was expressed quantitatively; and compared with experimental observations of fluidity of pure tin, tin-lead alloys, aluminum tin alloys and a low-carbon steel.

Theoretical analysis of fluidity of pure metals showed that for sand or glass molds (where heat flow rate decreases with time) the interface between solid metal and the flowing liquid is parabolic in shape.

For pure metals, the interface between solid and flowing liquid is smooth and fluidity can be evaluated quantitatively; for alloys, freezing is dendritic and quantitative analysis of fluidity would be much more difficult.

Small additions of lead to tin (above about 0.07 per cent) decrease fluidity markedly. Fluidity at the liquidus temperature of tin-0.01 per cent lead alloy was approximately 70 cm; fluidity of tin-0.1 per cent lead was 45 cm; and of tin-1.0 per cent lead, 22 cm.

Mode of solidification accounts for most of the differences in fluidity resulting from varying the alloy content of a metal. Pure metals freeze with smooth solid-liquid interfaces, and long times are necessary for the freezing of enough solid to offer substantial resistance to liquid flow; pure metals have high fluidity. Alloys freeze dendritically, and a relatively small amount of solid is able to prevent passage of further liquid; such alloys have lower fluidity. It was found that only 25 per cent solid in the entrance area of a test channel was sufficient to stop the flow of a tin-5.1 per cent lead alloy.

Flow stops in a fluidity test when some point is sufficiently solid to prevent passage of further liquid. This point may be at the entrance to the test channel, within the flowing stream or at the forward tip of the stream, depending (predictably) on alloy analysis and superheat.

Superheat is more effective in improving fluidity of pure metals than in improving fluidity of alloys. Reasons for this were analyzed theoretically, and it

was concluded that superheat caused a greater degree of remelting (during flow) in fluidity spirals of pure metals.

Experiments conducted on aluminum-tin alloys and on a low-carbon steel showed the results of earlier quantitative evaluations were applicable to the fluid flow behavior of these systems.

ACKNOWLEDGMENTS

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Fluidity data reported herein for the aluminum-tin system were abstracted from research performed by Dr. Sebastian Feliu Matas; this work is reported in detail elsewhere.[6,13]

REFERENCES

1. Briggs, C. S., "Fluidity of Metals," *Metals Handbook*, pp. 199-204, 1948.
2. Clark, K. L., "Fluidity Testing of Foundry Alloys," AFA TRANSACTIONS, *vol. 54*, pp. 37-48, 1946.
3. Krynitsky, A. L., "Progress Made in Fluidity Testing of Molten Metals During the Last Ten Years," AFS TRANSACTIONS, *vol. 61*, pp. 399-411, 1953.
4. Ragone, D. V., Adams, C. M. and Taylor, H. F., "Some Factors Affecting Fluidity of Metals," AFS TRANSACTIONS, *vol. 64*, pp. 640-652, 1956.
5. Ragone, D. V., Adams, C. M. and Taylor, H. F., "A New Method for Determining the Effect of Solidification Range on Fluidity," AFS TRANSACTIONS, *vol. 64*, pp. 653-657, 1956.
6. Foundry Section, Metals Processing Div., Dept. of Metallurgy, M.I.T., "Research on Parameters Influencing Fluidity in Aluminum Base Alloys," Final Report, Department of the Army, Ordinance Corps Contract No. DA-19-020-507-ORD-4503, July 1958.
7. Adams, C. M., "Thermal Considerations in Freezing," *Liquid Metals and Solidification*, A.S.M., Cleveland, p. 187, 1958.
8. Rutter, J. W. and Chalmers, B., "A Prismatic Substructure Formed During Solidification of Metals," *Canadian Journal of Physics, vol. 31*, p. 15, 1953.
9. Walker, James L., "Structure of Ingots and Castings," *Liquid Metals and Solidification*, A.S.M., Cleveland, p. 319, 1958.
10. Flemings, M. C., Adams, C. M., Hucke, E. E. and Taylor, H. F., "Metal Solidification in a Flowing Stream," AFS TRANSACTIONS, *vol. 64*, pp. 636-639, 1956.
11. Rightmire, B. G. and Taylor, H. F., "The Fluidity of Molten Steel," *Journal of Iron and Steel Institute, vol. 175*, pp. 167-176, Oct. 1953.
12. Lips, E. and Nipper, H., "Untersuchungen uber den Einfluss der Kristallausscheidung auf die Fliesseigenschaften von Schmelzen," *Die Giesserie, vol. 25*, pp. 369-372, July 1938.
13. Feliu Matas, S., Flemings, M. C. and Taylor, H. F., "Effect of Mode of Solidification on the Fluidity of Aluminum-Tin Alloys," submitted 26 International Foundry Congress, Spain.
14. Niesse, J. E., Flemings, M. C. and Taylor, H. F., "The Fluidity of a Series of Magnesium Alloys," AFS TRANSACTIONS, *vol. 65*, 1957.
15. Floreen, S. and Ragone, D. V., "The Fluidity of Some Aluminum Alloys," AFS TRANSACTIONS, *vol. 65*, pp. 391-393, 1957.
16. Stephens, *Phil. Mag., vol. 14*, p. 897, 1932.
17. DeVries, *Ind. Eng. Chem., vol. 22*, p. 617, 1930.
18. Hildebrand, Duschak, Foster and Beebe, *Journal of American Chemical Society, vol. 39*, p. 2293, 1917.
19. *Handbook of Chemistry and Physics*, 30th Edition, Chemical Rubber Publishing Co., Cleveland, 1946.

APPENDIX A

HEAT FLOW ANALYSIS FOR CYLINDRICAL CASTINGS

This heat flow analysis for cylindrical castings in molds such as sand may be made if one assumes that no resistance to heat flow occurs in the metal. The temperature distribution is represented in Fig. 5. Then, the heat flow rate at the metal-mold interface is given by:

$$q = 2\pi R h (T - T_{sc}) \qquad (A1)$$

Adams' equation[7] for heat flow rate into an "infinite" cylindrical mold is:

$$q = 2\pi R (T_{sc} - T_r) k' \left[\frac{1}{2R} + \frac{1}{\sqrt{\pi \alpha' \theta_h}}\right] \qquad (A2)$$

The unknown temperature, T_{sc}, may be eliminated by rearranging and then adding equations (A1) and (A2):

$$(T - T_r) = q\left[\frac{1}{2\pi R h} + \frac{1}{2\pi R k'\left(\frac{1}{2R} + \frac{1}{\sqrt{\pi \alpha' \theta_h}}\right)}\right] \qquad (A3)$$

Using $\alpha' = \frac{k'}{\rho' C'}$ and rearranging yields:

$$\frac{dQ}{d\theta} = q = 2\pi R h (T - T_r) \left[\frac{\sqrt{\theta_h}}{\left(1 + \frac{2Rh}{k'}\right)\sqrt{\theta_h} + \frac{2R}{\sqrt{\pi \alpha'}}} + \frac{\frac{2R}{\sqrt{\pi \alpha'}}}{\left(1 + \frac{2Rh}{k'}\right)\sqrt{\theta_h} + \frac{2R}{\sqrt{\pi \alpha'}}}\right] \qquad (A4)$$

for simplification let:

$$\left.\begin{aligned} A &\equiv \frac{2R}{\sqrt{\pi \alpha'}} \sec^{\frac{1}{2}} \\ B &\equiv \left(1 + \frac{2Rh}{k'}\right) \end{aligned}\right\} \qquad (A5)$$

Equation (A4) is a differential equation in terms of Q and θ_h. Separating the variables and integrating between the limits O to Q and O to θ_h yields:

$$Q = \frac{2\pi R h (T - T_r)}{B} \left[\theta_h + 2A\left(1 - \frac{1}{B}\right)\sqrt{\theta_h} - \frac{2A^2}{B}\left(1 - \frac{1}{B}\right) \ln\left(1 + \frac{B}{A}\sqrt{\theta_h}\right)\right] \qquad (A6)$$

Equation (A6) defines the heat entering the mold as a function of time. All of the physical properties in equation (A6) may be found from published data, except h which must be determined empirically (A value of $h = 0.18$ cal/cm² C sec was found for tin freezing in glass tubes of inside diameter of 0.386 cm).

APPENDIX B

DEFINITION OF SYMBOLS

$A = 2R/\sqrt{\pi \alpha'}$ $\sec^{1/2}$

$B = \left(1 + \frac{2Rh}{k'}\right)$

c' = Specific heat of mold material cal/gm C

D = Diameter of mold channel, cm

f = Friction factor

g = Acceleration due to gravity cm/sec^2

h = Heat transfer coefficient at metal-mold interface, cal/°C cm^2 sec

H = Heat of fusion of metal, cal/gm

k' = Thermal conductivity of mold, cal/sec cm^2 °C/cm

L = Length along metal stream from mold channel entrance, cm

L_e = Length to end of metal stream, cm

L_f = "Fluidity" or final length of metal stream, cm

q = Heat flow rate, cal/sec

Q = Total heat entering mold material per unit length in time θ_h, cal/cm^2 sec

r = Radius of liquid metal in mold, cm

R = Radius of mold channel, cm

T = Temperature of liquid metal, C

T_i = Initial or pouring temperature, C

T_m = Melting temperature or liquidus temperature, C

T_r = Room temperature (25 C) or initial mold temperature, C

T_{sc} = Temperature of mold at mold-metal interface, C

V = Average metal velocity at any cross-section, cm/sec

V_e = Velocity of end of metal stream, cm/sec

V_i = Initial velocity of metal stream, cm/sec

Z = Height of metal head, cm

d' = Thermal diffusivity of mold, cm^2/sec

θ = Time after metal starts to flow, sec

θ_h = Time after metal-mold contact, sec

θ_f = Final freezing time, sec

ρ = Metal density, gm/cm^3

ρ' = Mold density, gm/cm^3

ϕ_{ent} = Entrance loss coefficient

CAST LOW ALLOY STEELS DUCTILITY AND TOUGHNESS

Reduced phosphorus and sulfur content significance

By John Zotos

ABSTRACT

A developed basic-electric arc furnace melting practice was employed to produce low sulfur and low phosphorus 1040, 4140, high nickel 4325, 4330 and high silicon 4340 type steel castings. The castings produced were chemically analyzed, and the high nickel 4325 and 4330 cast steels were subjected to various heat treatments and physically tested. A decrease in the sulfur content of the high nickel 4325 steel castings resulted in a significant increase in the reduction in area, impact strength and elongation, while mantaining a constant strength level.

A decrease in both the phosphorus and sulfur content of several 4325 and 4330 steel castings also resulted in a significant improvement in their ductility and impact resistance.

INTRODUCTION

The basic-electric steelmaking process permits the refining of steel under oxidizing, neutral and reducing conditions, and with adequate slag control, produces good quality steels.[1,2] In the basic-electric practice, phosphorus is removed during the oxidizing period and sulfur is removed primarily during the reducing period. Furthermore, compared to other steel processes such as the basic-open-hearth, the basic-electric practice can produce cleaner steels of more uniform composition, from poorer raw materials.[3]

Many investigators have studied the reactions occurring between the slag and metal during the production of basic-electric steels.[4-11] Slag characteristics which affect phosphorus and sulfur removal during the basic-electric practice include basicity, oxidizing power and fluidity.[12-14]

The oxidizing period in the basic-electric steel practice commences with the formation of the molten metal. The objectives during this period are to reduce the carbon content of the steel, produce a vigorous boil to intermix the slag and metal and aid in nitrogen and hydrogen removal and oxidize the phosphorus in the steel to calcium phosphate, which is ultimately removed in the deslagging operation. The majority of the elements present in the steel oxidize to varying degrees, and the reactions which occur are discussed in detail in the literature.[1-14]

J. ZOTOS is Asst. Chief, Experimental Foundry Branch, Rodman Laboratory, Watertown, Mass.

The reducing period in the basic-electric steel practice commences with the deoxidation of the molten steel by the controlled addition of deoxidizers such as ferrosilicon, ferromanganese and carbon.[10] The objective during this period is to reduce the sulfur content of the steel by forming a basic reducing slag containing two per cent calcium carbide (CaC_2).[14] The calcium carbide combines chemically with the coke, lime, fluorspar and silicon in the slag, desulfurizes the steel and returns reducible compounds of manganese, chromium, iron, etc., from the slag back into the metal.[7,8] Sulfur removal improves with increased slag basicity and higher bath temperatures.[14]

Other investigators have studied both the effect of deoxidation practice, and the combined effect of phosphorus and sulfur content, melting practice and testing temperature on the physical properties of basic-electric, cast low alloy steels.[15,17] In general, basic-electric steel castings with decreased phosphorus and sulfur contents have the best ductility and impact resistance.[17]

In reviewing the published data, insufficient information was available on the physical properties of cast low alloy steels having phosphorus and sulfur contents in the vicinity of 0.010 per cent. This report presents the results of investigations conducted to evaluate the significance of reduced phosphorus and sulfur contents on the ductility and toughness of cast low alloy steels.

PROCEDURE

The effect of decreased phosphorus and sulfur content on the mechanical properties of low alloy steel castings was studied during the development of an operating procedure for producing low phosphorus and low sulfur steel in a 9 ft diameter, basic-electric arc melting furnace. This melting practice (described in Appendix A) initially produced 4140 and high nickel 4325 steel castings with average sulfur and phosphorus contents of 0.010 per cent, and was later used to evaluate its effectiveness in producing 1040, 4330 and high silicon 4340 type cast steels.

The 1040, 4140 and high silicon 4340 steels were cast into 6 and 8 in. diameter coupons (25 in. long) which were processed and chemically tested.

The high nickel 4325 steel was cast into 90mm (M-41) breech rings and 4 in. test coupons which were shaken out, chemically analyzed and subjected to the following heat treatment:

Normalized at 1900 F, 16 hr and air cooled.

Austenitized at 1650 F, 6 hr; furnace cooled to 1500 F, 2 hr; water quenched, 6 min; air cooled for 10 min; oil quenched 15 min.

Tempered at 600 F, 2 hr; raised to 1130 F, 6 hr; oil quenched for 15 min.

Soaked at 600 F, 12 hr and air cooled.

The 4330 steel was cast into 280mm shell bodies and 4 in. test coupons which were shaken-out, chemically analyzed and subjected to the following heat treatment:

Normalized at 1900 F, 16 hr and air cooled.

Austenitized at 1625 F, 5 hr; furnace cooled to 1500 F, 1½ hr; water quenched, 5 min; air cooled, 5 min; oil quenched, 10 min.

Tempered at 1000 F, 5 hr and water quenched.

Soaked at 600 F, 12 hr and air cooled.

Upon completion of the heat treatment the breech rings, shell bodies and coupons were physically tested in accordance with Military Specification, MIL-S-10029. The 0.357 in. diameter, 3¼ in. long, shoulder-type tensile specimens[19] were removed from specified locations in the castings[18] and tested at room temperature to determine their tensile strength, yield strength at 0.1 per cent offset, per cent elongation and per cent reduction in area. The 0.394 in. square, 2.165 in. long, V-notch Charpy specimens[19] were also obtained from specified locations in the castings[18] and tested at −20 F (−29 C) to determine their impact strength.

RESULTS AND DISCUSSIONS

The chemical analyses of the 14 steel heats produced during this investigation are listed in Table 1.

Figure 1 illustrates the relationship between the sulfur content of the steel and the processing time. Based on the average curve, the meltdown sulfur content of the steels produced is 0.024 per cent and the final sulfur content is 0.008 per cent, indicating that 67 per cent of the sulfur originally contained in the scrap charge was removed using the developed melting practice. Breaking the results down into the various periods of the procedure, the average curve shows that 0.002 per cent sulfur was removed from the steel during the oxidizing period and an additional 0.002 per cent sulfur was eliminated during the wash period.

Thus, the first two periods account for only 25 per cent of the total sulfur removed. During the reducing period, the average sulfur curve drops 0.012 per cent (Fig. 1), indicating that the reducing slag used during this investigation removed 75 per cent of the total sulfur eliminated over the entire heat cycle.

Similarly, Fig. 2 illustrates the variation in the phosphorus content of the steels with increasing process time. Based on the average curve, 0.018 per cent phosphorus was removed from the steel during the oxidizing period, and a constant phosphorus level was main-

TABLE 1 — ANALYSES OF HIGH QUALITY STEELS PRODUCED USING DEVELOPED OPERATING PROCEDURE IN SIX AND ONE-HALF TON BASIC ELECTRIC FURNACE

Heat No.	Type of Steel	STEEL COMPOSITION IN PER CENT							
		C	Mn	Si	Cr	Mo	Ni	S	P
1	1040	0.40	0.92	0.21	—	—	—	0.014	0.013
2	4140	0.43	0.80	0.19	1.07	0.25	—	0.013	0.012
3	4140	0.39	1.33	0.37	1.08	0.21	—	0.009	0.011
4	4140	0.41	0.99	0.35	1.05	0.19	0.42	0.012	0.010
5	HiNi 4325	0.26	1.50	0.45	0.92	0.45	2.45	0.010	0.012
6	HiNi 4325	0.29	1.25	0.32	0.75	0.43	2.30	0.007	0.010
7	4330	0.29	0.96	0.23	0.96	0.37	2.04	0.005	0.007
8	4330	0.31	0.94	0.20	0.88	0.36	2.02	0.012	0.011
9	4330	0.32	1.19	0.46	0.85	0.36	2.02	0.007	0.006
10	4330	0.285	1.59	0.44	1.00	0.40	2.02	0.010	0.021
11	4330	0.26	0.85	0.35	0.88	0.34	1.92	0.011	0.019
12	4330	0.30	1.12	0.45	0.89	0.36	2.07	0.009	0.010
13	4330	0.305	1.22	0.23	0.97	0.33	1.96	0.009	0.010
14	HiSi 4340	0.42	0.82	3.20	1.06	0.29	1.92	0.010	0.011

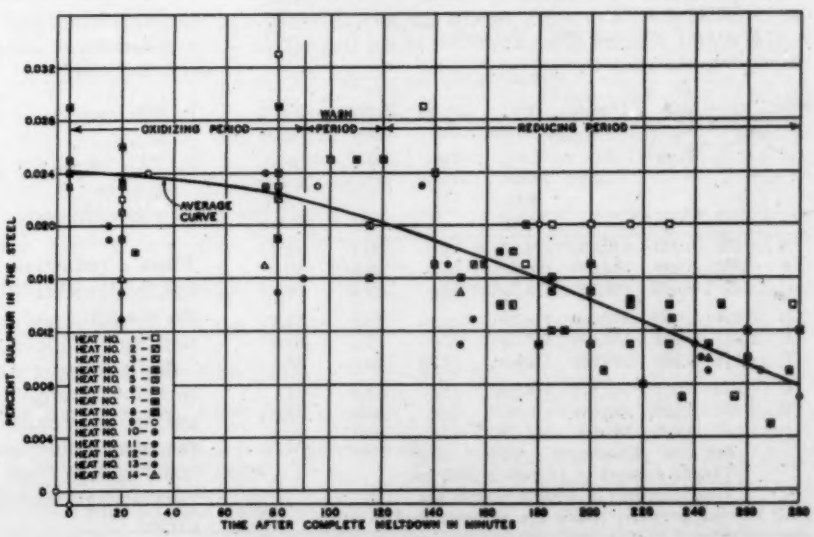


Fig. 1 — Relationship between the sulfur content of the steels and the processing time.

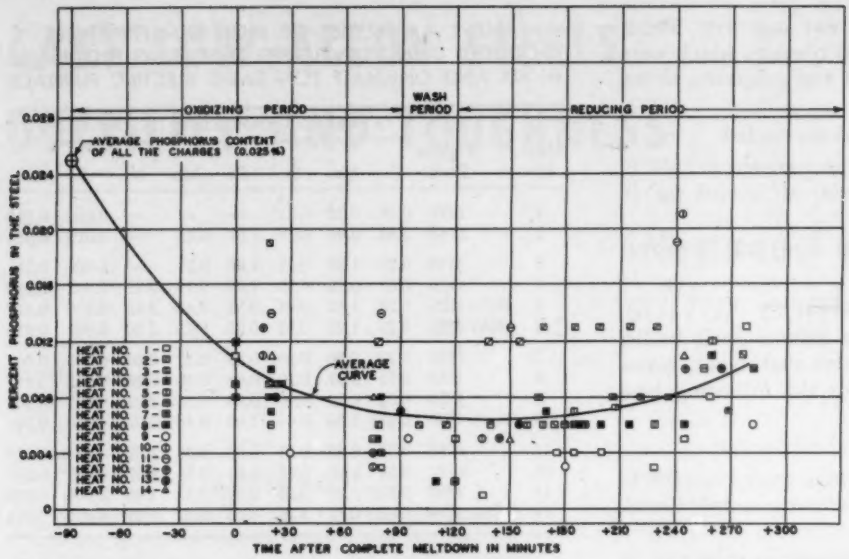


Fig. 2 — Relationship between the phosphorus content of the steels and the processing time.

tained during the wash period. During the reducing period, the average phosphorus curve rose 0.004 per cent to a final value of 0.010 per cent. This rise in phosphorus content can be accounted for as being present in the alloy additions to the steel during the refining period.

These results indicate that the oxidizing slag used during this investigation, removes the bulk of the phosphorus contained in the initial scrap steel charge.

A series of nine basic-electric, high nickel 4325 breech ring steel coupons having a constant low phosphorus content (0.010 ± 0.002 per cent) and varying sulfur content (0.007 to 0.037 per cent) were subjected to the uniform heat treatment previously described and then physically tested. The results, listed in Table 2, indicate that the tensile and yield strength levels remained relatively constant (133,000 ± 8000 psi and 118,000 ± 6000 psi, respectively) while the ductility and toughness improved with decreasing sulfur content.

TABLE 2 — PHOSPHORUS AND SULFUR CONTENT AND PHYSICAL PROPERTIES OF NINE HIGH NICKEL 4325 BASIC-ELECTRIC STEEL HEATS OBTAINED FROM A STANDARD BREECH RING COUPON (4 IN. DIAMETER)

Heat No.	HiNi 4325 Steel Analysis		PHYSICAL PROPERTIES* Tensile Strength	Yield Strength,	Elongation	Reduction	V-Notch Charpy
	S, %	P, %	(Rm. Temp)	0.1% Offset (Rm. Temp.	(Rm. Temp), %	(Rm. Temp), %	Impact, −29 F or −29 C (Ft lb)
A	0.037	0.010	142,250	126,500	12.8	22.4	14.2
B	0.032	0.009	133,000	120,000	13.3	26.8	18.4
C	0.0275	0.010	133,000	118,750	13.6	26.3	23.5
D	0.024	0.008	132,000	118,250	15.4	23.5	34.2
E	0.017	0.012	130,500	117,500	17.1	42.8	35.6
F	0.014	0.009	126,000	112,000	17.9	44.9	36.5
G	0.010	0.012	132,375	116,250	18.6	51.0	34.1
H	0.008	0.009	135,600	120,250	18.6	49.6	38.6
I	0.007	0.010	130,500	115,000	18.6	46.3	38.8

Per cent phosphorus = 0.010 ± 0.002.
Tensile strength = 133,000 ± 8000 psi.
Yield Strength = 118,000 ± 6000 psi.

*All Physical property results are an average of two tests.

Sulfur Content Effect

Figure 3 illustrates this effect of sulfur content on the reduction in area, elongation and impact strength of the high nickel 4325 steel castings. A linear regression analysis (Appendices B, C and D) of the data resulted in the following mathematical relationships:

1) (% R. A) = −975 (%S) + 57.3.
2) (Impact) = −782 (%S) + 45.7.
3) (% Elong) = −215 (%S) + 20.4.

where:

% R.A. = Room temp. per cent reduction in area.
Impact = Charpy V-notch impact strength in ft-lb at −20 F or 29 C.
% Elong. = Room temp. per cent elongation.
%S = the per cent sulfur contained in the steel.

The correlation coefficients resulting from the statistical analyses (Appendices B, C and D) were −0.971, −0.927 and −0.983 for equations (1), (2) and (3), respectively. Using these equations, it is evident that a reduction in sulfur content from 0.030 to 0.010 per cent increases:

1) The reduction in area 70 per cent (from 28.0 to 47.6).
2) The impact resistance 70.7 per cent (from 22.2 to 37.9).
3) The elongation 30 per cent (from 14 to 18.2).

Thus, a reduction in the sulfur content of basic-electric, high-nickel 4325 cast steel significantly improves the ductility and toughness while maintaining a constant strength level.

Extrapolating the per cent sulfur in the high nickel 4325 steel castings (Fig. 3) to zero, and comparing the resulting physical properties to a 0.030 per cent sulfur steel, a maximum increase of 104.6, 105.8 and 45.7 per cent in the reduction in area, impact strength and elongation, respectively, can be predicted.

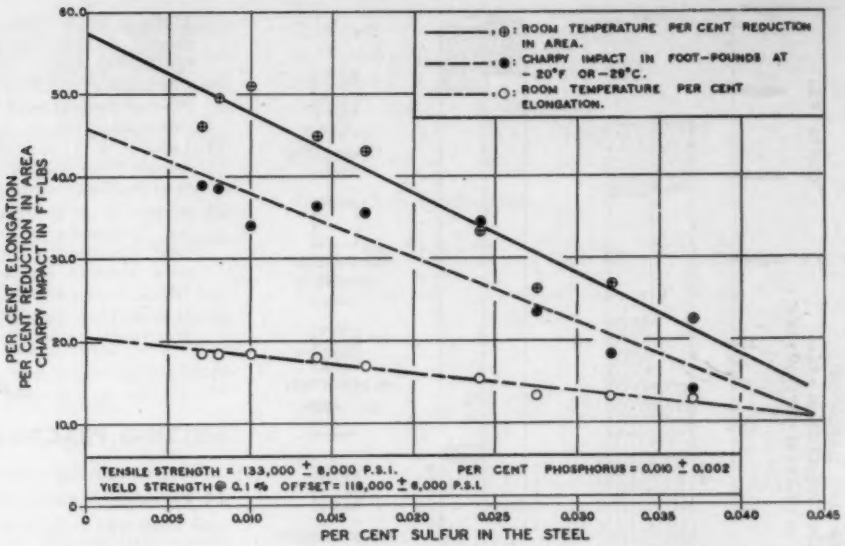


Fig. 3 — Sulfur content effect on the physical properties of cast high nickel 4325 steels.

Several 4325 and 4330 steel castings produced by both basic-electric and basic-induction melting practices and having low and moderate phosphorus and sulfur contents, respectively, were subjected to the heat treatments previously described and then physically tested. The results listed in Table 3 are graphically illustrated in Fig. 4. At relatively constant strength levels (132,000 psi), the 0.008 per cent phosphorus and 0.009 per cent sulfur 4325 basic-electric steel castings have reduction in area, impact and elongation properties which are 16.8, 7.2 and 12.9 per cent greater, respectively, as compared to the 4325 basic-induction steel castings having 0.016 per cent phosphorus and 0.018 per cent sulfur.

In addition, at constant strength levels (156,000 psi), the 0.010 per cent phosphorus and 0.009 per cent sulfur 4330 basic-electric steel castings have reduction in area, impact and elongation properties which are 21.8, 116.7 and 12.4 per cent greater, respectively, as compared to the 4330 basic-induction steel castings having 0.012 per cent phosphorus and 0.025 per cent sulfur. Data available in the literature on a 4330 steel casting subjected to a similar heat treatment (T.S. – 150,000 psi) and containing 0.013 per cent phosphorus and 0.015 per cent sulfur had an impact strength of 15 ft lb at −20 F compared to the low phosphorus and sulfur 4330 steel value of 36.4 ft lb.

These results indicate that reduced phosphorus and sulfur content significantly improves the ductility and impact resistance of 4325 and 4330 steel castings while the strength levels remain constant.

ACKNOWLEDGMENT

The author is deeply appreciative to Mr. K. D. Holmes of the Rodman Laboratory's Experimental Foundry Branch for supplying some of the castings used in this study.

REFERENCES

1. Camp, J. M. and Francis, C. B., *The Making, Shaping and Treating of Steel,* United States Steel Co., Pittsburgh, 5th Edition, p. 481, April 1940.
2. Stagg, H. J., "Fifty Years of Progress in Electric Furnace Steelmaking," *Metal Progress, vol. 69,* no. 4, p. 49, April, 1956.

TABLE 3 — ANALYSES AND PHYSICAL PROPERTIES OF SEVERAL STEEL HEATS

Type of Steel	Melting Practice	STEEL ANALYSIS IN PER CENT								PHYSICAL PROPERTIES				
		C	Mn	Si	Cr	Mo	Ni	S	P	Tensile Strength (Rm. Temp)	Yield Strength, 60.1% Offset (Rm. Temp)	Elongation (Rm. Temp), %	Reduction in Area (Rm. Temp), %	V-Notch Charpy Impact in ft lb (−20 F or −29 C)
HiNi 4325	Basic-Electric	0.245	0.87	0.42	0.74	0.44	2.48	0.008	0.009	132,750	116,000	19.3	52.9	39.4
HiNi 4325	Acid-Induction	0.25	0.79	0.33	0.72	0.45	2.34	0.018	0.016	133,000	117,500	17.1	45.3	36.8
4330	Basic-Electric	0.305	1.22	0.23	0.97	0.33	1.96	0.009	0.010	156,000	141,000	13.6	38.0	36.4
4330	Acid-Induction	0.27	0.80	0.36	0.86	0.35	2.01	0.025	0.013	157,500	145,000	12.1	31.2	16.8
4330*	Basic-Electric	0.32	0.76	0.40	0.69	0.32	1.65	0.015	0.013	150,000	——	—	—	15.0

*Data were obtained from reference 17 and the heat treatment used was:
Normalized at 1650 F, 8 hr and air cooled.
Austenitized at 1600 F, 6 hr and oil quenched.
Tempered at 1020 F, 1 hr and water quenched.

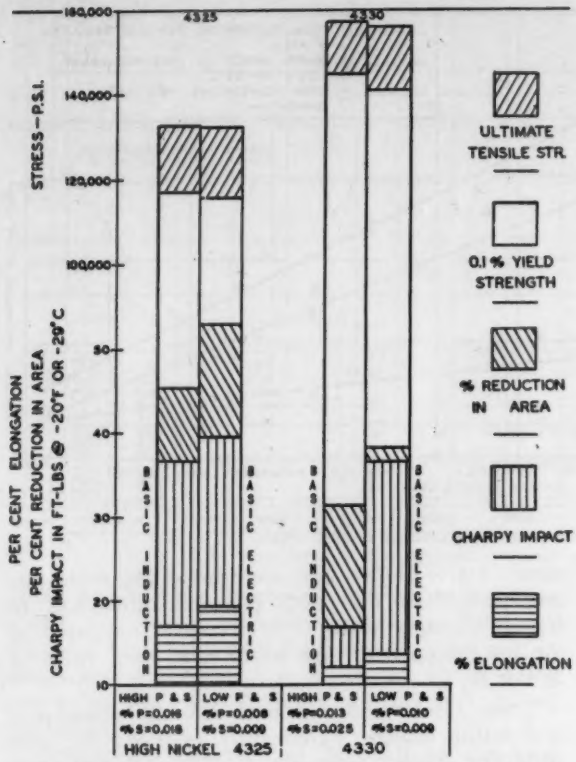


Fig. 4 — Phosphorus and sulfur effect on the physical properties of cast 4325 and 4330 steels.

3. Farley, R. W. and McCurdy, R. J., "Scrap — Availability, Present Physical Condition, and Contamination," *A.I.M.E. Electric Furnace Steel Proceedings, vol. 8,* 1950.
4. Ess, T. J., "The Modern Arc Furnace," *Iron and Steel Engineer,* p. AF-24, Feb. 1944.
5. Shortsleeve, F. J. and Troiano, A. R., "A Literature Survey on the Mechanism of Flake Formation by Case Institute of Technology," *Interim Technical Report No. 1,* p. 95, Jan., 1953.
6. Bever, M. B., Emerick, H. B., Larsen, B. M. and Philbrook, W. O., *Basic Open Hearth Steelmaking,* American Institute of Mining, Metallurgical and Petroleum Engineers, New York, 1951.
7. Briggs, C. W., *The Metallurgy of Steel Castings,* McGraw-Hill Book Company, Inc., New York, 1st Edition, p. 55, 1946.
8. Camp, J. M., and Francis, C. B., *The Making, Shaping and Treating of Steel,* United States Steel Co., Pittsburgh, 6th Edition, p. 513, June, 1951.
9. Carter, S. F., "Chromium Reactions in Acid and Basic Furnaces," *A.I.M.E. Electric Furnace Steel Proceedings, vol. 13,* pp. 134-149, 1956.
10. Bishop, H. L., Chipman, J., Grant, N. J. and Lander, H. N., "Equilibria of Sulphur and Oxygen Between Liquid Iron and Open Hearth Type Slags," *Journal of Metals, vol. 8,* No. 7, pp. 862-868, July, 1956.
11. Zotos, J., "The Development of a Standard Operating Procedure for a Six and One-Half Ton Basic-Electric Steel Furnace Using Optimum Conditions of Desulphurization and Dephosphorization," *Production Preparedness Measure Report No. 10/7,* Rodman Process Laboratory, Watertown Arsenal, October 1956.
12. Tibbets, W., "Comparison of Acid and Basic-Electric Melting Practice for Mild and High Tensile Strength Steel Castings," *A.I.M.E. Electric Furnace Steel Proceedings, vol. 13,* pp. 206-221, 1955.
13. Johnson, H. H. and Fisher, "Quality Control of Melting Practices," *A.I.M.E. Electric Furnace Steel Proceedings, vol. 13,* pp. 150-165, 1955.
14. Loria, E. A., "Sulphur Removal Studies in Basic-Electric Melting," *Blast Furnace and Steel Plant, vol. 43,* no. 1, pp. 86-87, 114-115, Jan., 1955.
15. Milligan, M. F., "Effect of Deoxidation on Mechanical Properties," *A.I.M.E. Electric Furnace Steel Proceedings, vol. 5,* pp. 180-181, 1947.
16. Christopher, C. F., "Deoxidation and Its Effects on the Physical Properties of Steels," *A.I.M.E. Electric Furnace Steel Proceedings, vol. 11,* pp. 101-120, 1953.
17. Jackson, W. J., and Michie, G. M., "The Low-Temperature Impact Properties of Cast Steel," *Journal of the Iron and Steel Institute, vol. 187,* part 2, pp. 104-120, Oct. 1957.
18. Ahearn, P. J. and Belser, "Watertown Arsenal Manufacturing Techniques for Gun Tubes, Breech Rings and Ferrous Sand Castings for the Period 1950-1954 Inclusive," *Watertown Arsenal — Rodman Laboratory Technical Bulletin,* no. W.P.L. 1, April, 1954.
19. Driscoll, D. E. and Dalsy, D. M., "Standard Test Specimens," *Watertown Arsenal Laboratory Report,* Jan., 1952.

APPENDIX A

MELTING PRACTICE

The operating procedure established for producing 14 heats of high quality, low sulfur and phosphorus cast steel was as follows:

1) The charge consisted of 12,000 lb of scrap, 400 (amount of coke varied according to per cent lb of limestone and 50 lb of calcined coke carbon in the scrap; a meltdown carbon of approximately 0.8 per cent was desired). The limestone and coke were loaded into the furnace after the bottom was lined with a thin layer of scrap and the remaining scrap was added.
2) Meltdown of the charge was initiated.
3) A 20-min waiting period followed the meltdown during which the bath was rabbled several times with the aid of a steel skimmer, and a bath temperature of 2850 F to 2950 F was maintained.
4) Metal test specimen *A* was taken and analyzed for carbon, manganese, silicon, chromium, molybdenum, nickel, sulfur and phosphorus.
5) Oxidation commenced with a controlled addition of 480 lb of dry iron ore and 200 lb of limestone, premixed together and added over a 45 min period.
6) One hr after the start of the ore addition, the slag (meltdown slag plus oxidizing slag) was removed through the slag door with the aid of steel skimmers. During this period, the phosphorus in the steel oxidized and reacted with the slag to produce calcium phosphate, which was removed during the deslagging operation.
7) After the first slag-off, metal test specimen *B* was taken for carbon, sulfur and phosphorus analyses.
8) The wash slag consisting of 180 lb of limestone and 42 lb of dry fluorspar was added to the clean steel bath. This slag absorbed any excess phosphates remaining after the oxidizing slag had been removed, and minimized the possibility of phosphorus reverting back into the molten steel.
9) Thirty min later the wash slag was removed. The steel bath was rabbled several times during the wash period to intermix the slag and metal.
10) After the second slag-off, the reducing slag was added in the following sequence:
    a) 90 lb of ferrosilicon, 60 lb of ferromanganese and 10 lb of carborite were added.

b) Upon complete solution of the components added in *a*, the premixed slag consisting of 360 lb of limestone, 60 lb of dry fluorspar and 15 lb of fine ferrosilicon was added.

c) When this slag became molten, powdered coke amounting to 30 lb in total was added to the slag at various intervals and intermixed with the slag by rabbling the bath. This aided in the formation of calcium carbide for favorable desulfurization.

d) Varying amounts of limestone, fluorspar, ferrosilicon and coke were added at intervals to properly shape up the slag. The bath was rabbled several times and the bath temperature was maintained between 2900 F and 2950 F.

During this period, the sulfur in the steel reacted with the slag to produce calcium sulfide, which remained in the slag if a high basicity was maintained.

11) After the reducing slag had shaped up, metal test specimen *C* was taken for a complete analysis (step 4).
12) Upon receipt of the analysis of metal test specimen *C*, the appropriate nickel, chromium, molybdenum and carbon additions were made in the form of ferro-alloys.
13) Upon complete solution of these alloy additions, the power input was raised to bring the bath up to the desired pouring temperature.
14) Ferro-alloy additions of manganese and silicon, based on metal test *C* were made and the molten steel was heated to the tapping temperature.
15) The final metal test specimen *D* was taken for a complete analysis and the heat was tapped into the bottom pour ladle.
16) One-tenth of 1 per cent aluminum was added to the ladle during the tapping operation to assure maximum deoxidation.
17) The various steels produced were cast into 6 and 8 in. billets (1045, 4140, and high silicon 4340), 90mm breech rings and test coupons (high nickel 4325) and 280mm shell bodies and test coupons (4330).
18) The castings were processed and evaluated for their chemical and physical properties.

APPENDIX B

LINEAR REGRESSION ANALYSIS OF THE VARIATION IN THE ROOM TEMPERATURE PER CENT REDUCTION IN AREA WITH DECREASING SULFUR CONTENT IN HIGH NICKEL 4325 STEELS

Y = Rm. temp. per cent reduction in area (Table 2)

$$\bar{Y} = \frac{\Sigma Y}{N} = \frac{343.6}{9} = 38.2$$

x = Per cent sulfur (Table 2)

$$\bar{x} = \frac{\Sigma x}{N} = \frac{0.1765}{9} = 0.0196$$

Summation Results:

1) $N = 9$
2) $\Sigma (x - \bar{x}) = +0.0001$
3) $\Sigma (Y - \bar{Y}) = 0.2$
4) $\Sigma (x - \bar{x})^2 = +0.00096189$
5) $\Sigma (x - \bar{x})(Y - \bar{Y}) = -0.93763$
6) $\Sigma (Y - \bar{Y})^2 = +968.76$

Correlation Coefficient – r:

$$r = \frac{\left(\frac{\Sigma (x-\bar{x})(Y-\bar{Y})}{N}\right) - \left(\frac{\Sigma (x-\bar{x})}{N} \times \frac{\Sigma (Y-\bar{Y})}{N}\right)}{\sigma_x \times \sigma_Y} = -0.971$$

where:

$$\sigma_x = \left[\frac{\Sigma (x-\bar{x})^2}{N} - \frac{\Sigma (x-\bar{x})^2}{N}\right]^{1/2} = \pm 0.01034$$

$$\sigma_Y = \left[\frac{\Sigma (Y-\bar{Y})^2}{N} - \frac{\Sigma (Y-\bar{Y})^2}{N}\right]^{1/2} = \pm 10.38$$

Equation for Best Fitting Straight Line:

$$(Y - \bar{Y}) = \frac{r\sigma_Y}{\sigma_x}(x - \bar{x}) - \frac{r\sigma_Y}{\sigma_x}\left[\frac{\Sigma (x-\bar{x})}{N}\right] + \frac{\Sigma (Y-\bar{Y})}{N}$$

Negligible

$$(Y - 38.2) = -975\,(x - 0.0196)$$
$$Y - 38.2 = -975\,x + 19.1$$
$$Y = -975 \times 57.3$$
$$(\%\ \text{R.A.}) = -975\,(\%S) + 57.3$$

APPENDIX C

LINEAR REGRESSION ANALYSIS OF THE VARIATION IN THE V-NOTCH CHARPY IMPACT (—20 F or —29 C) WITH DECREASING SULFUR CONTENT IN HIGH NICKEL 4325 STEELS

V-Notch Charpy Impact

Y = in ft lb —20 F or —29 C (Table 2)

$$\bar{Y} = \frac{\Sigma Y}{N} = \frac{273.9}{9} = 30.4$$

x = Per cent sulfur (Table 2)

$$\bar{x} = \frac{\Sigma x}{N} = \frac{0.1765}{9} = 0.0196$$

Summation Results:

1) $N = 9$
2) $\Sigma (x - \bar{x}) = +0.0001$
3) $\Sigma (Y - \bar{Y}) = +0.3$
4) $\Sigma (x - \bar{x})^2 = +0.00096189$
5) $\Sigma (x - \bar{x})(Y - \bar{Y}) = -0.75263$
6) $\Sigma (Y - \bar{Y})^2 = +684.23$

Correlation Coefficient – r:

$$r = \frac{\left(\frac{\Sigma (x-\bar{x})(Y-\bar{Y})}{N}\right) - \left(\frac{\Sigma (x-\bar{x})}{N} \times \frac{\Sigma (Y-\bar{Y})}{N}\right)}{\sigma_x \times \sigma_Y} = -0.927$$

where:

$$\sigma_x = \left[\frac{\Sigma (x-\bar{x})^2}{N} - \frac{\Sigma (x-\bar{x})^2}{N}\right]^{1/2} = \pm 0.01034$$

$$\sigma_Y = \left[\frac{\Sigma(Y-\bar{Y})^2}{N} - \frac{\Sigma(Y-\bar{Y})^2}{N}\right] \frac{1}{2} = \pm 8.72$$

Equation for Best Fitting Straight Line:

$$(Y-\bar{Y}) = \frac{r\sigma_Y}{\sigma_X}(X-\bar{X}) - \frac{r\sigma_Y}{\sigma_X}\left[\frac{\Sigma(X-\bar{X})}{N}\right] + \frac{\Sigma(Y-\bar{Y})}{N}$$

Negligible

$$(Y-30.4) = 783\,(x-0.0196)$$
$$Y-30.4 = -782\,x + 15.3$$
$$Y = -782\,x + 45.7$$
$$\text{(Impact in ft-lbs)} = -782\,(\%S) + 45.7$$

APPENDIX D

LINEAR REGRESSION ANALYSIS OF VARIATION IN ROOM TEMPERATURE PER CENT ELONGATION WITH DECREASING SULFUR CONTENT IN HIGH NICKEL 4325 STEELS

Rm. temp. per cent
Y = Elongation (Table 2)

$$\bar{Y} = \frac{\Sigma Y}{N} = \frac{145.9}{9} = 16.2$$

x = Per cent sulfur (Table 2)

$$\bar{X} = \frac{\Sigma x}{N} = \frac{0.1765}{9} = 0.0196$$

Summation Results:

1) $N = 9$
2) $\Sigma(x-\bar{x}) = +0.0001$
3) $\Sigma(Y-\bar{Y}) = +0.1$
4) $\Sigma(x-\bar{x})^2 = +0.00096189$
5) $\Sigma(x-\bar{x})(Y-\bar{Y}) = -0.21216$
6) $\Sigma(Y-\bar{Y})^2 = +48.35$

Correlation Coefficient – r:

$$r = \frac{\left(\frac{\Sigma(x-\bar{x})(Y-\bar{Y})}{N}\right) - \left(\frac{\Sigma(x-\bar{x})}{N} \times \frac{\Sigma(Y-\bar{Y})}{N}\right)}{\sigma_x \times \sigma_Y} = -0.983$$

where:

$$\sigma_x = \left[\frac{\Sigma(x-\bar{x})^2}{N} - \frac{\Sigma(x-\bar{x})^2}{N}\right] \frac{1}{2} = \pm 0.01034$$

$$\sigma_Y = \left[\frac{\Sigma(Y-\bar{Y})^2}{N} - \frac{\Sigma(Y-\bar{Y})^2}{N}\right] \frac{1}{2} = \pm 2.32$$

Equation for Best Fitting Straight Line:

$$(Y-\bar{Y}) = \frac{r\sigma_Y}{\sigma_X}(X-\bar{X}) - \frac{r\sigma_Y}{\sigma_X}\left[\frac{\Sigma(X-\bar{X})}{N}\right] + \frac{\Sigma(Y-\bar{Y})}{N}$$

Negligible

$$(Y-16.2) = -215\,(x-0.0196)$$
$$Y-16.2 = -215\,x + 4.2$$
$$Y = -215\,x + 20.4$$
$$(\%\ \text{Elongation}) = -215\,(\%S) + 20.4$$

SIGNIFICANCE OF REPORTED CHEMICAL ANALYSIS OF CAST IRON

By D. E. Krause

ABSTRACT

Chemical analysis data, reported by more than 30 laboratories for 40 samples of cast iron over a period of 15 years, indicate that limits for accuracy and reproducibility are greater than commonly assumed. Good reproducibility of analyses by a laboratory does not necessarily imply good accuracy. Accuracy of analysis can be best determined by submitting the same sample of iron to at least six laboratories for analysis. The range for reported analyses from a large group of laboratories will be about three times as great as that obtained by one laboratory for the same sample of iron.

The existence of this somewhat broad range for reported analysis of cast iron in the literature must be recognized when evaluating the relationship of chemical composition of cast iron to its mechanical and physical properties. Causes for some of the variations in accuracy of analyses are explained.

INTRODUCTION

Although cast iron is purchased principally on the basis of its mechanical properties, the chemical analysis of cast iron is important since so many of its properties are dependent on its composition. For many applications, experience has shown that in addition to certain mechanical properties, it is desirable to also specify composition limits for some of the elements in cast iron.

For example, Specification A-278 (A.S.T.M.) "Gray Iron Castings for Pressure-Containing Parts for Temperatures up to 650 F," places limits of 3.8 per cent for carbon equivalent, 0.25 per cent for phosphorus and 0.12 per cent for sulfur. Many other specifications for cast iron often contain definite composition limits.

In addition to chemical composition being part of a purchase specification, it is important to the foundryman and metallurgist as a means of controlling the properties of cast iron. There are of course other factors governing the mechanical properties of the cast iron such as melting practice, inoculation and foundry practice which may not be evaluated on the basis of composition. Chemical analysis of cast iron is therefore an important quality control tool.

Not only is chemical analysis useful for estimating the mechanical properties of the iron in the casting, but also for evaluating the operation of the melting equipment and the quality of the raw materials being charged into the cupola or furnace. The need for accurate chemical analyses with good reproducibility should be obvious. Although cast iron is one of the lowest priced cast ferrous materials it is metallurgically a rather complex material, which adds materially to the difficulty of obtaining accurate chemical analyses.

Analytical data from a particular laboratory may indicate a high degree of accuracy because of the precision or reproducibility of results. The high degree of precision or reproducibility may be misleading, since it will be shown that such results can be quite incorrect. Evaluation of an analytical method and the skill of the analyst, by means of National Bureau of Standards cast iron samples, does not necessarily assure accuracy of analysis of all samples of cast iron determined by that method.

Purpose of Study

It is the purpose of this paper to evaluate the accuracy and reproducibility of routine chemical analysis of cast iron. Differences in reported analyses for a particular sample of cast iron when submitted to several laboratories are often puzzling.

The accuracy of analytical data for cast iron appearing in technical publications on melting and the properties of cast iron must be known if the results of different investigators are to be compared. Since the carbon equivalent value has such a marked effect on the properties of cast iron, it is important that the significance of reported values be recognized.

There have been many publications showing the relationship between tensile strength and carbon equivalent. For A.S.T.M. Class 25 to class 50 cast irons (1.2-in. diameter bar), a difference of 0.10 per cent CE will result in a change of 2500 to 3000 psi in tensile strength.

Since the metallurgist is interested in the true composition of the iron, sampling and analytical techniques should be adopted which approach this ideal. The sample preparation procedure used by the National Bureau of Standards for example is designed to yield a high degree of uniformity in the prepared sample for standardization purposes, rather than to represent the true analysis of the casting or the melt.

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The carbon content of such a sample may be 0.10 to 0.40 per cent lower than the actual carbon content of the casting from which the sample was obtained.

ACCURACY, REPRODUCIBILITY AND PRECISION

The terms accuracy, precision and reproducibility are often carelessly used. Precision and reproducibility can be used interchangeably, and are used to designate the closeness with which the results of a series of determinations agree. Since the term precision is too often confused with accuracy, the term reproducibility will be used in this paper in preference to the term precision. A chemical analysis method may give a high degree of reproducibility, and yet be inaccurate insofar as reflecting the true value for the concentration of the element in the iron.

In molybdenum containing cast iron the evolution method for sulfur may give values with a high degree of reproducibility of say 0.08 per cent, while the gravimetric or combustion method will give a result of possibly 0.11 per cent with about the same degree of reproducibility but much better accuracy. Therefore, a high degree of reproducibility reported for a certain determination may have no relation to accuracy. Accuracy is used to indicate how close the reported value is to the true value of the amount of element present.

It may be difficult to arrive at the true value, but if a sufficient number of determinations are carried out using methods known to have a high degree of accuracy, and by several analysts, the true value might be determined within rather narrow limits. The significance of the terms, accuracy and reproducibility as applied to chemical analysis of cast iron has been rather clearly discussed by Clarke.[1]

Sampling Technique Effect

The accuracy and reproducibility of chemical analysis are of course influenced by the sampling technique. Unless a representative sample is obtained, a high degree of accuracy of the chemical analysis is of limited value. Correct processing of the sample for analysis is likewise important if the final results are to be reliable.

In regard to errors affecting the accuracy of chemical analysis, Clarke[1] has conveniently grouped them into two divisions namely:

1) Determinate errors.
2) Indeterminate errors.

The determinate errors are defined as those which can be recognized and defined. These can be divided into three classes as follows.

a) Errors due to faulty apparatus or impure reagents.
b) Personal errors.
c) Errors due to the method employed.

These three factors can be controlled, and their effect minimized by proper selection of apparatus and reagents, training of the analysts and use of a method known to give reliable results.

Whereas the causes for the determinate errors can be defined and corrective measures taken, the cause for indeterminate errors is generally not known. Using one method and making successive determinations, the results obtained by any one analyst will show certain variations. These variations generally become smaller as the skill of the analyst increases. If a sufficient number of determinations are made, the results will produce a typical probability curve as found in statistical analysis.

CHEMICAL ANALYSIS DATA FOR AN ALLOY CAST IRON

Over a period of 15 years, 35 samples of cast iron in various forms have been submitted to approximately 30 laboratories for chemical analysis. During this period, the composition of the samples has varied over the following limits;

Element	Per Cent
Total carbon	2.75 - 3.80
Silicon	1.64 - 3.07
Sulfur	0.05 - 0.14
Manganese	0.35 - 0.96
Phosphorus	0.06 - 0.78

Although the reproducibility of analyses varies with the composition of the iron and the nature of the sample, the frequency distribution of the reported analytical data has a certain degree of similarity. The data reported for an alloy iron sample (No. 568) has been selected for discussion rather than an unalloyed iron because of the increasing production of alloy cast irons.

Castings in the form of plates 5½-in. x 2¼-in. and ½-in. thick were cast from alloy iron and sent to 25 laboratories. With the exception of a specially prepared sample of turnings for carbon determination, the laboratories prepared their own samples. Spectrographic analyses were reported by some laboratories using both the photographic film and the direct reader types of equipment. The results reported for this alloy iron sample are summarized in Table 1.

The desired limits of accuracy are based on a combination of reproducibility values published in Lundell, Huffman and Bright,[2] Westwood and Mayer[3] and Clarke,[1] for recommended methods for analysis for cast iron. In addition, the results of certain lab-

TABLE 1 — SUMMARY OF ANALYSES FOR ALLOY IRON SAMPLE 568

Element	Values, %				Limits ± %	Analyses in Limits, %
	Probable	Min.	Max.	Range		
Carbon, (SpT)*	3.11	3.04	3.28	0.24	0.04	77
Carbon, (R)**	3.12	2.96	3.24	0.28	0.04	73
Silicon	1.93	1.80	2.08	0.28	0.03	60
Sulfur	0.110	0.051	0.120	0.079	0.005	61
Manganese	0.60	0.54	0.64	0.10	0.02	68
Phosphorus	0.060	0.040	0.123	0.083	0.01	63
Chromium	0.35	0.27	0.39	0.12	0.02	76
Nickel	0.91	0.83	1.00	0.17	0.02	62
Vanadium	0.11	0.05	0.16	0.11	0.01	47
Copper	0.28	0.21	0.34	0.13	0.02	63
Molybdenum	0.56	0.49	0.64	0.15	0.02	62

*Specially prepared turnings.
**Sample prepared by analyst.

oratories participating in these surveys who have in their employ highly skilled chemists are also used to establish these limits. The limits are further modified to take into account the relative importance of the various elements to the metallurgist.

Actually, with the exception of the results for vanadium, the per cent of analyses within the limits shown is quite close to one standard deviation in a statistical analysis. At least half of the values for vanadium were reported by laboratories not ordinarily analyzing for vanadium which may account for low percentage of values within the limits shown. Experience has shown that these limits are not unreasonable.

The frequency distribution plots of the analyses reported for alloy iron sample 568 are shown in Figs. 1 and 2 together with the methods used for obtaining the values. The relationship of the methods used to the values obtained will be briefly commented on in regard to this particular sample. Factors affecting the determination of carbon, silicon and sulfur will be discussed in greater detail later on in this paper.

Analyses Results

Carbon. In regard to carbon analyses shown in Fig. 1, the results obtained with the gasometric method using either an induction heater or resistance heated furnace do not show as good a reproducibility as the gravimetric methods. The same type of distribution by method is characteristic for carbon analyses for previous surveys. The exact cause for the lesser accuracy for the gasometric method is not known. Green[4] reports that the larger number of joints in the gasometric equipment, and the greater possibility of personal error in reading a burrette, are contributing factors.

Silicon. The results for silicon shown in Fig. 1 indicate somewhat better reproducibility for the perchloric acid method than for the sulfuric acid method. Although photometric methods for silicon analysis are sometimes satisfactory for plant control work at the lower silicon levels, the presence of a considerable amount of alloying elements evidently seriously affects the accuracy and reproducibility of these methods. Although it is generally recognized that a chill cast sample is more desirable than a gray iron sample of cast iron for spectrographic analysis, the two spectrographic values for silicon using a gray iron sample are quite good.

Sulfur results obtained with the evolution method for this sample are generally quite low, as shown in Fig. 1, as five of the values are off the left end of the chart. The presence of molybdenum in the sample is known to seriously interfere with the reliability of the evolution method for sulfur. With the exception of one low value at 0.098 per cent sulfur, the combustion method using a resistance heated furnace shows better reproducibility than the high frequency induction heated furnace. This is not always true as some of the previous surveys for unalloyed samples

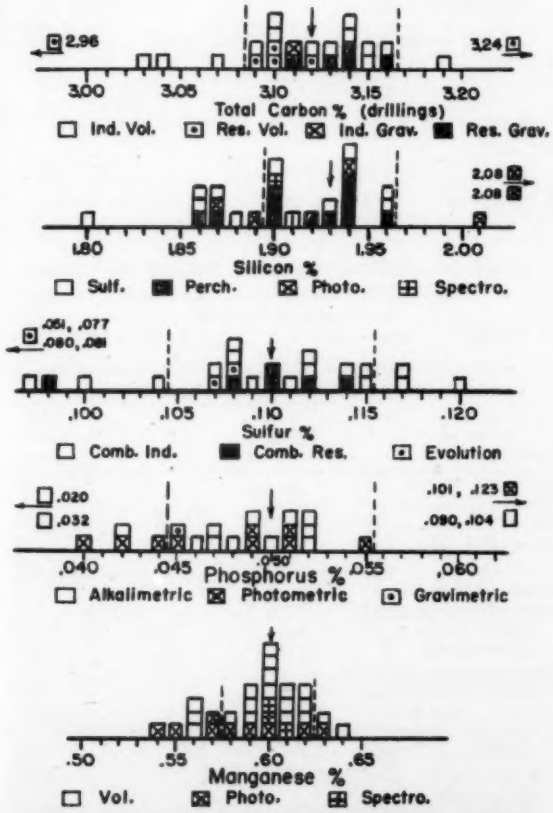


Fig. 1 — Distribution of chemical analyses, sample 568.

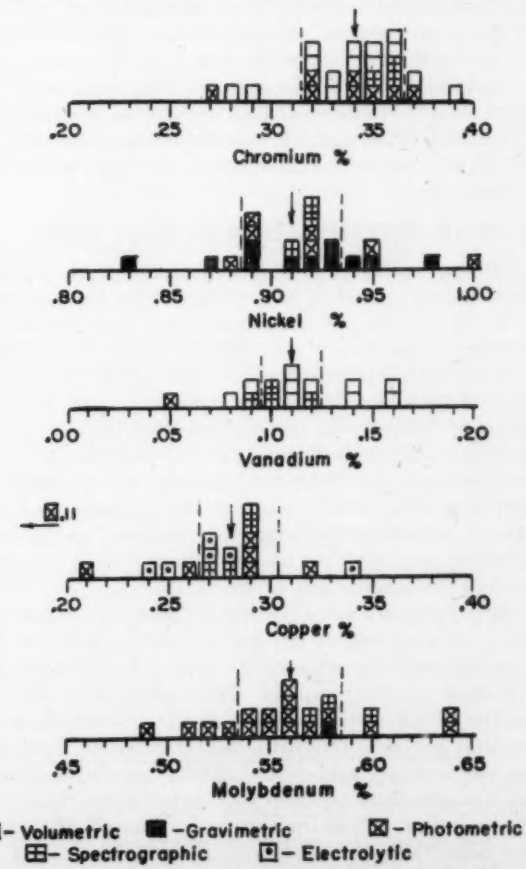


Fig. 2 — Distribution of chemical analyses, sample 568.

indicated excellent accuracy for the combustion-induction method.

Phosphorus results for sample 568, shown in Fig. 1, exhibit a wide range. The phosphorus content of the iron is fairly low, and this factor together with the presence of vanadium may be responsible for some of the extreme values. Phosphorus determinations were made primarily with just two methods, namely the alkalimetric and photometric. One method appears no better than the other.

Manganese results, as shown in Fig. 1, are characteristic of those reported for other samples in the past. The volumetric method (persulfate-arsenite) appears to be better than the photometric method, and the three spectrographic results are good. In general, manganese analyses are within a narrow range and show good reproducibility and since the effect of a variation of ± 0.02 per cent manganese on the metallurgy of the iron is hardly detectable, the limits could well be wider than shown and still represent control.

Alloying elements. The results reported for the various alloying elements are shown in Fig. 2. The data are largely self explanatory, although attention is directed to the good results obtained with spectrographic methods. Inherently, the direct reading spectrograph should be capable of considerably greater accuracy and reproducibility than conventional chemical analysis methods, since the sample weighing source of error is eliminated as well as some of the other variations in the skills of analysts.

The direct reader should also be superior in accuracy to the film photographic type of instrument, as variations in processing of the film from development to measurement of line density on the film are eliminated. The principal deterrent to wider use of the direct reading spectrograph is the high initial installation cost.

CARBON DETERMINATION IN CAST IRON

Of the various elements contributing to the mechanical and physical properties of cast iron, carbon is easily one of the most important. Unfortunately, determination for the true carbon content of cast iron is beset with many difficulties. The major portion of the carbon in cast iron exists as graphite and, because of its low density in comparison with the matrix, it occupies a much larger volume in the iron than the analysis indicates. The low strength of the graphite and lack of adhesion between the graphite and the matrix further complicates the problem, and a considerable and variable loss of carbon during preparation of the sample can occur.

The difficulty in obtaining an accurate carbon analysis of cast iron is dependent on the form of the graphite and the amount present. The size and distribution of the graphite flakes may often have a greater effect on the accuracy of the analysis than the amount present. The influence of graphite flake size on the accuracy and reproducibility of carbon analyses has also been reported by Clarke[5] and others.[6,7,8] If the section size of the casting is thin and the graphite flakes small, the carbon content determined on drillings is similar to that obtained for solid samples.

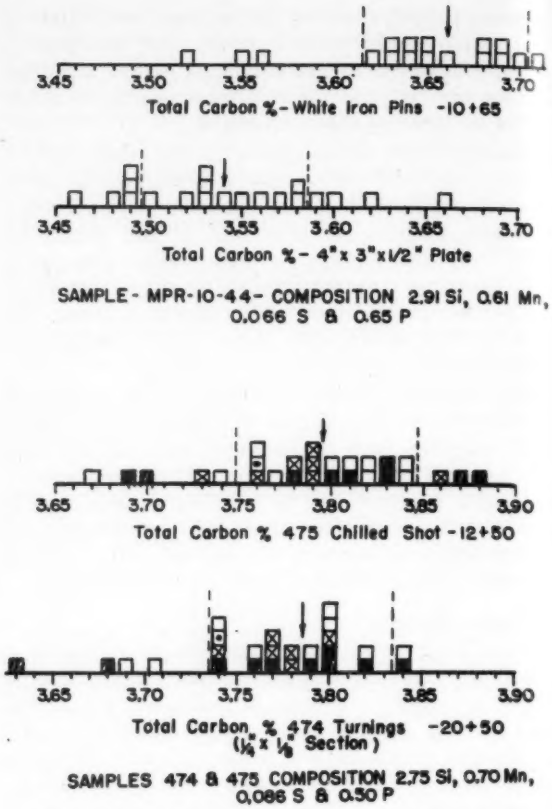


Fig. 3 — Influence of section size and chill cast on total carbon analyses.

On the other hand, for a 7-in. thick section of an ingot mold[5] solid samples gave a carbon value of 3.93 per cent, the +30 sieve fraction of the drillings gave 2.78 per cent while the −30 sieve fraction gave 6.60 per cent.

The influence of cooling rate on accuracy and reproducibility of carbon analyses is further shown in Fig. 3, samples MPR-10-44 and 474-475. Sample MPR-

TABLE 2 — DISTRIBUTION OF CARBON IN 1.2-IN. DIAMETER TRANSVERSE TEST BARS

Sample EA 10-45

Probable analysis — 3.50% T.C., 2.43% Si, 0.69% Mn, 0.096% S, 0.27% P.

Description of Sample	Laboratory BMI T.C.% ave.	CLC T.C.% ave.	UPR T.C.% ave.
Drillings from hole through diameter of bar		3.42	3.48
Faced off sample, whole section			3.46
Slugs cut from thin slices of entire section			
Edge of bar	3.51	3.40	3.46
Interior of bar	3.58	3.50	3.51
Layer analyses (turnings)			
Outside (1.2 in D. to 0.95-in D.)	3.51		3.47
half radius approx. (0.73-in D to 0.48-in D.)	3.50		3.50
center of bar (0.25-in D.)	3.49		3.48

TABLE 3 — RELATIONSHIP OF SAMPLE LOCATION AND PREPARATION ON CARBON ANALYSES

Sample No. 3-49

Probable analysis — 3.41% T.C., 3.07% Si, 0.62% Mn, 0.093% S, 0.58% P (Hypereutectic).

Casting in form of plate 12-in. x 3-in. x ½-in. thick cast horizontally.

Description of Sample	Laboratory — Carbon, %				
	CLC	LR	LL	GR	UPR
Drillings from entire cross section	3.37	3.33	3.35		3.41
Drillings with 5/32-in. drill, 7/32-in. deep hole					
Cope surface	3.39		3.30	3.50	3.46
Drag surface	3.09		3.26	3.43	3.33
Solid pieces, cope	3.45				
Solid pieces, drag	3.39				
Solid pieces, entire ½-in. section	3.40				3.43

10-44 was cast in the form of chilled pins in iron molds and as plates 4-in. x 3-in. x ½-in. thick in green sand molds. The probable value for the chill cast pins is shown as 3.66 per cent carbon while the probable value for drillings obtained from the ½-in. thick plate is 3.54 per cent carbon or a loss of 0.12 per cent carbon.

On the other hand, there was comparatively little difference in carbon content between chilled shot and sand cast rings 2⅞-in. diameter with a ¼-in. x ⅛-in. cross-section, as shown in Fig. 3 for samples 474 and 475. Although both sets of samples were cast in the same foundry, they were obtained at different times.

Although the possibility of carbon segregation in large castings particularly if the carbon content is high is generally recognized, segregation may also occur in lighter castings. There can be a difference in carbon content between the outside and inside of a 1.2-in. diameter test bar, as shown by the results in Table 2. The difference is more marked when solid samples are used for analysis instead of turnings.

It is of course entirely possible that some carbon was lost in preparation of the turnings. If the iron is of hypereutectic composition, a difference in carbon content between cope and drag surfaces might be found in even a ½-in. thick plate. Data from five laboratories for sample 3-49 are shown in Table 3. Drillings obtained with a 5/32-in. drill showed differences as great as 0.30 per cent carbon. Care must therefore be exercised in obtaining samples from castings of a hypereutectic iron.

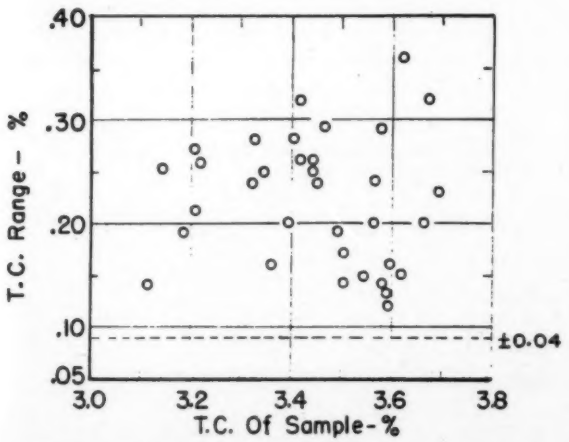


Fig. 4 — Relationship of carbon content to range of reported analyses, 35 samples.

Carbon Analyses Accuracy

That carbon content in itself has little influence on the reproducibility, and accuracy of carbon analyses can be seen from inspection of Fig. 4. Data in Fig. 4 were obtained from analysis of 35 samples of iron by 18 to 24 laboratories. In some cases, the range in carbon values was less for a sample containing 3.60 per cent carbon than for a sample containing 3.20 per cent carbon.

Accuracy and reproducibility of carbon analyses are related to the laboratory making the determinations. In Fig. 5 the relationship of the maximum deviation and average deviation from the probable carbon value of 19 to 35 samples of cast iron of various types for each of 18 laboratories to the per cent of carbon analyses within ±0.04 per cent carbon of the probable value is shown. The relation shown could be anticipated from a statistical standpoint, but the important point is that laboratories with a good accuracy record are not apt to be far off on any one particular sample.

That the personal factor enters into the problem of reproducibility and accuracy of chemical analysis is partly illustrated in the relation of range in silicon values to range in carbon values reported for repeated analysis of the same sample. Such a relation is shown in Fig. 6. Data for Fig. 6 were obtained by repeated analysis at daily intervals of the same sample of iron or several samples which were renumbered each time to mask their identity.

This was done in 20 laboratories for a period of a month. Since the carbon and silicon determinations are entirely separate from each other, it is difficult to attribute the relationship shown in Fig. 6 to any particular method of analysis. The graph is based entirely on iron samples from the plant with which the laboratory is associated.

Standardization Problem

In discussing the problem of standardization of the equipment and method for carbon analysis, analysts frequently felt that a higher carbon iron than available from the National Bureau of Standards would be

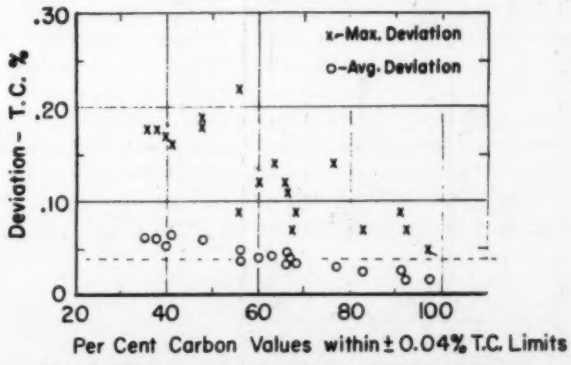


Fig. 5 — Relationship of maximum and average deviation to per cent of carbon analyses within ± 0.04 per cent T.C. limits for each laboratory.

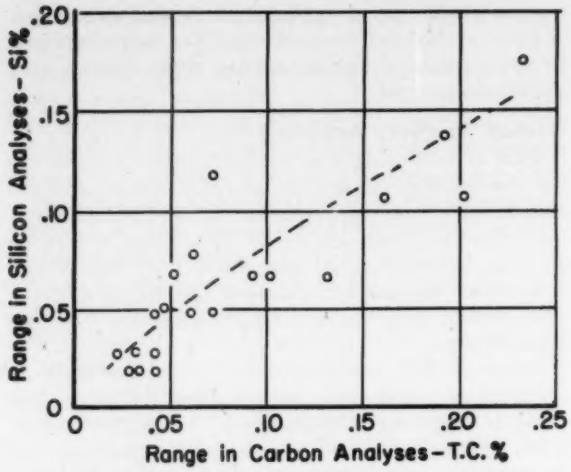


Fig. 6 — Relationship of range in carbon analyses to range in silicon analyses for plant reproducibility samples.

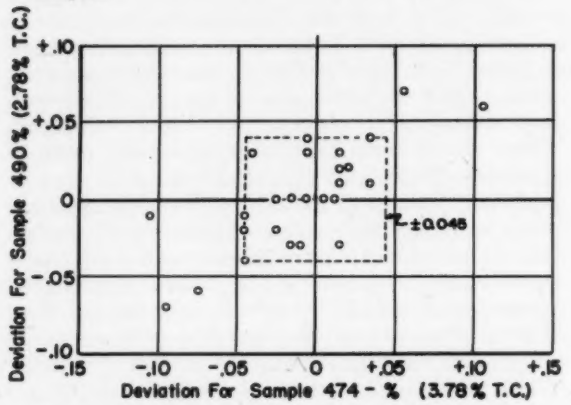


Fig. 7 — Relationship of average deviations for samples 474 and 490.

helpful. A reply to a questionnaire sent to 32 laboratories indicated almost complete agreement on the desire to have available a high carbon iron for a standard. A high carbon gray iron sample was therefore prepared.

Since it was already known that a small graphite flake size was essential to avoid losses in carbon during preparation of the sample, a fast cooling rate would be necessary. A high carbon iron was poured into rings 2⅞-in. O.D. with a ¼-in. x ⅛-in. section and a special turned and mixed sample prepared having a probable carbon content of 3.78 per cent, as shown in Fig. 3. To determine the reproducibility for this high carbon sample, the laboratories were asked to make one carbon determination per day for a period of about 10 days, and at the same time also repeat the procedure with a low carbon iron (Sample 490-2.78 per cent carbon).

Altogether approximately 650 carbon determinations were made during this investigation. The average deviation for each laboratory for the two samples was calculated, and the values for the high carbon sample plotted against the values for the low carbon sample, as shown in Fig. 7.

It should be observed that the average deviation in reported values by a laboratory for the 3.78 per cent carbon iron was nearly always reported in the same direction as the deviation for the 2.78 per cent carbon sample. Some of the individual plant analyses distributions for sample 474 (high carbon) and sample 490 (low carbon) shown in Fig. 8 were selected to illustrate reproducibility for low and high average values, as well as for values close to the probable carbon content of the iron.

The results from laboratory "BL" in Fig. 8 are somewhat unusual in that the average of a rather wide range in values is close to the probable value.

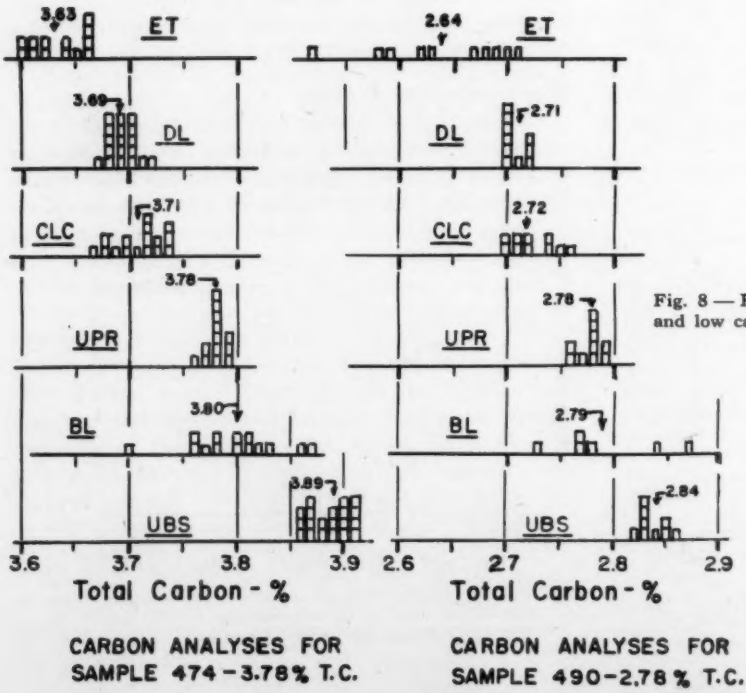


Fig. 8 — Reproducibility data for high and low carbon iron samples.

Generally, high reproducibility is associated with high accuracy, as shown by the results of laboratory "UPR." After studying the data, it was decided that the carbon level of the currently available National Bureau of Standard Samples was adequate for standardization purposes. If satisfactory values are obtained with a standard having a carbon content of 3.20 per cent, the procedure will also be satisfactory for higher carbon irons. Uniformity of the sample is more important than actual carbon content.

On the basis of the data reported here, and more detailed examination of procedures involved in carrying out the determinations, the following suggestions for good carbon analyses are made:

1. Sample
   a) Chilled iron pin should be used whenever possible.
   b) If a casting must be sampled, a solid piece is preferred, particularly for heavy sections.
   c) Thin section castings may with reasonable care be sampled by drilling.
2. Procedure
   a) Replace oxygen purifying train and sulfur removing trap at regular intervals. Erratic carbon results will be obtained long before the reagents in use are spent.
   b) Reanalyze routine samples under a new identification number at least several times a week to detect lack of accuracy of results.
   c) Carefully control oxygen flow. Rew[8] and Green[4] state that control of oxygen flow rate to within 1 per cent of 250 ml per min is essential for accurate results.

SILICON DETERMINATION IN CAST IRON

After carbon, silicon is probably next in importance. In contrast to carbon, silicon is virtually all present in the form of a solid solution in the matrix of the iron. There are substantiated examples of some of the silicon reported by chemical analysis to be in the form of silicates. The proportion of the silicon present as silicates is generally quite small, and should not influence the accuracy of the determination by chemical means. Silicon is important as the amount present has an appreciable effect on the solubility of carbon in iron, and is also one of the factors controlling the graphitizing tendencies of the iron.

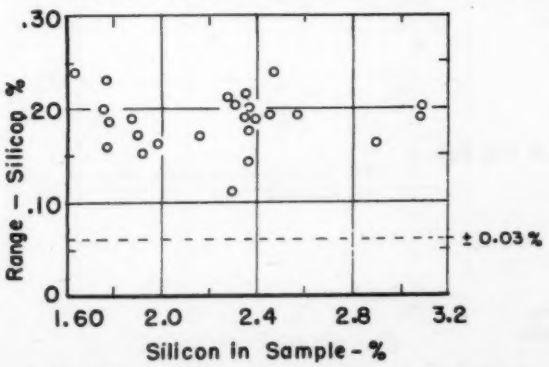


Fig. 9 — Relationship of silicon content of iron to range in reported silicon analyses.

The silicon content is primarily determined by four methods. The perchloric acid method generally yields more accurate results than the sulfuric acid method. Review of plant methods using nitro-sulfuric acid often discloses that those laboratories reporting inaccurate results are not following the recommendations of A.S.T.M.[9] and Lundell, Hoffman and Bright.[2] The ratio of sulfuric acid to sample weight may be wrong or taking the dehydrated sample into solution for filtration is carelessly done.

The data shown in Fig. 1 are representative of other samples. Some laboratories consistently obtain satisfactory results with the nitro-sulfuric acid method and others do not. The perchloric acid method is not entirely foolproof, but does yield more accurate results when data from many laboratories are compared. It is also the preferred method if the filtrate is to be used for determinations of other elements.[9] Silicon can be determined photometrically, but experience thus far indicates that the composition of the sample in regard to alloying elements present must be taken into account.

The accuracy of silicon analyses obtained by the photometric method shown in Fig. 1 are not as good as generally obtained, and might be attributed to the high alloy content of the sample. Timing of the analytical procedure is important. Although the accuracy of silicon determinations by the photometric method is better for unalloyed samples with a silicon content of under 2.00 per cent, it is still not as dependable as the perchloric acid method. Too little data are available to evaluate the spectrographic method, but if the results in Fig. 1 are representative the method seems quite satisfactory.

There is no apparent relationship between the silicon content of a sample of iron and the deviation of reported analyses from the probable value as shown in Fig. 9. In addition to the samples covering a silicon range of 1.62 per cent to 3.07 per cent, the samples varied as to the amounts of other elements present and the nature of the casting from which the samples were taken.

SULFUR DETERMINATION IN CAST IRON

The sulfur content of cast iron is important to the metallurgist for several reasons. On the basis of service experience, purchasers of gray iron castings have found it desirable to place a maximum limit on sulfur content, and this limit is often 0.12 per cent as found in A.S.T.M. specifications A 278, A 126, A 319 and many others. When the sulfur content of the iron at the spout approaches the 0.12 per cent limit, it is necessary to know the sulfur content accurately as an incorrectly reported low value might result in rejection of the castings, and an incorrectly reported high value might involve unnecessary desulfurizing treatments.

In the production of heavy castings, too high a sulfur content may give rise to an increase in cope surface defects, usually blowholes of various types. In the production of ductile iron, the actual sulfur content is important as much of the control of the properties of ductile iron depend on knowing accurately the

initial sulfur content. The importance of knowing the sulfur content accurately is one reason for placing the limits for analytical data for sulfur at ±0.005 per cent, as shown in Table 1.

Most cast irons contain sufficient manganese and are low enough in sulfur content that the sulfur exists primarily as manganese sulfide. When molybdenum, zirconium, titanium or aluminum is present, the composition of the sulfides might be somewhat more complex. The variable recovery of sulfur from cast iron using the evolution method might be an indication of differences in composition of the sulfides and the ease with which they are attacked by the hydrochloric acid.

Determination Methods

Sulfur is generally determined by either the evolution method or the combustion method. For soft, annealed or certain unalloyed irons, reasonably good results can sometimes be obtained with the evolution method. The analysis for sulfur by the evolution method is seriously affected by the presence of molybdenum, as shown by the sulfur analyses distribution in Fig. 1. Although the combustion sulfur method is not invulnerable to error, it nevertheless is inherently a satisfactory method if recommendations of A.S.T.M.[10] are followed. Furthermore, it can be used with a chilled iron sample. The British Cast Iron Research Association has published information on this method.[11]

In the evolution method for sulfur, some laboratories attempt to correct for incomplete evolution of the sulfur from the sample by applying a correction factor depending on how much sulfur is present. This can be quite unsatisfactory. The accuracy of the evolution method depends on how consistently and how completely the sulfur is evolved from the sample and not necessarily on the amount present. In the evolution method, it is common practice to establish an empirical factor for the titrating solution based on Bureau of Standards cast iron samples which are usually pearlitic.

If only 75 per cent or less of the sulfur is evolved, and the analyst assumes that 95 per cent or more of the sulfur from the Standard sample was evolved, the factor assigned to the titrating solution will be in error. The titrating solution factor will only be correct when other samples of iron being analyzed evolve exactly the same proportion of their sulfur content as the Standard sample. This will be illustrated by two examples.

In Fig. 10, sulfur analyses reported for two samples of cast iron with the method used are shown. The probable analyses for these two samples are as follows:

Sample	Composition, %					Condition of sample
	T.C.	Si	Mn	S	P	
474	3.79	2.75	0.70	0.086	0.505	as-cast
413	3.52	2.63	0.78	0.107	0.28	annealed

Inspection of Fig. 10 shows that the sulfur values reported for sample 474 using the evolution method were generally lower than those obtained with the combustion method. Sample 474 was cast in green sand in the form of rings with a 1/4-in. x 1/8-in. cross-section with pearlitic matrix. In contrast with the above results, the sulfur values for sample 413 using the evolution method were generally higher than those obtained by the combustion method. Sample 413 was cast in a permanent mold and subsequently annealed. The annealing treatment obviously greatly improved the evolution of sulfur from the sample. The titrating solution was obviously incorrectly standardized.

There is a rather good relationship between the sulfur content of the iron and range of reported values, as shown by the data for 21 samples plotted in Fig. 11. Since many of the early data were not identified as to method used, it is impossible to comment on the effect of method on the relation in Fig. 11.

Ladle or Casting Sampling

In sampling iron in a ladle or from a casting, it is necessary to observe certain precautions otherwise nonrepresentative sulfur analyses will be obtained. If samples for analysis are taken from near the cope sur-

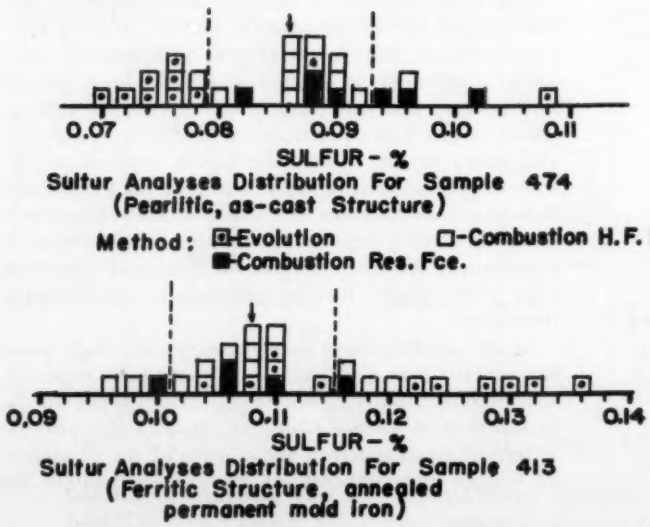


Fig. 10 — Microstructure effect on sulfur analyses.

face of large castings, the sulfur content might be appreciably higher than obtained from samples farther in from the surface. The segregation of manganese sulfide is marked in centrifugal castings. Sulfur can segregate to the top of the surface of a ladle of as small a capacity as one ton.

One of the samples used during the analysis survey was cast in the form of 12 plates, 12-in. x 5½-in. x ½-in. thick. The iron in the one ton ladle was well stirred at the furnace and then transported to the molding floor. The molds were green sand and poured with a 100 lb capacity hand ladle. Ten molds were poured from the first 100 lb ladle, and two molds from the second 100 lb ladle of iron. The one ton ladle was held on the monorail while pouring the first ten molds. Reported analyses for sulfur indicated that there might be a difference between the ten plates from the first ladle and the two plates from the second ladle.

Wherever possible, samples were taken from all the plates and the sulfur content determined by two laboratories having a good record of analytical accuracy. The analyses thus obtained are shown in Fig. 12. It is apparent that the sulfur content is lower for the last two plates than for the first ten. The only reasonable explanation that can be offered is that the first ladle must have contained manganese sulfide inclusions which rose to the top when the iron was transported from the furnace to the molding floor.

CONCLUSIONS AND RECOMMENDATIONS

The implied accuracy of chemical analyses for cast iron by reporting to a larger number of significant figures than justified is misleading. Good reproducibility for a particular determination by a laboratory does not necessarily imply a high degree of accuracy. The range of reported analyses for a sample of iron from a large number of laboratories will be from two to three times as great and at least twice as great as would be expected using the same methods by only one laboratory.

Four analysts using the same equipment in a laboratory may obtain results for a given sample over a wider range than obtained by only one analyst. Because of the wider than expected range, chemical analysis data for cast iron reported in the literature must be evaluated with the wide range of reproducibility kept in mind.

It is a waste of time to perform chemical analysis on an improperly taken and prepared sample. Since the cause of determinate errors is known, every effort should be made to reduce such errors to a minimum.

It is worth while to provide adequate room and facilities for a laboratory. Too small a laboratory will be subjected to greater extremes in temperature than a larger space. Extremes in temperature and inadequate removal of fumes affect the useful life and accuracy of the equipment and the efficiency and skill of the analyst. Balances should be checked at regular intervals for sensitivity and accuracy. Preventive maintenance is just as important to accurate analyses as it is to efficient operation of foundry equipment. Too often, new equipment is installed and then virtually forgotten insofar as keeping it in first class working condition is concerned.

Skilled Analysts Needed

It is almost impossible to obtain reasonably accurate chemical analyses with improperly trained analysts. Even with proper training, some persons become more highly skilled analysts than others. The analysts should be encouraged to keep abreast of developments in analytical procedures. Because of the nature of chemical analysis and the location of the laboratory particularly in the smaller foundries, the life of an analyst may become somewhat boring and the accuracy of the analyses will suffer. The analyst is human and a little more appreciation of his efforts will be repaid in more accurate analyses.

The economic losses to a foundry resulting from several inaccurate analyses may well be considerably greater than the cost of sending the analyst to a meeting of chemists now and then where he will have the opportunity to discuss his problems with others in similar positions. It is necessary that the analyst be thoroughly interested in his work if he is to do a good job. Providing good reference books and related periodicals will well repay the cost incurred through more interest in the job.

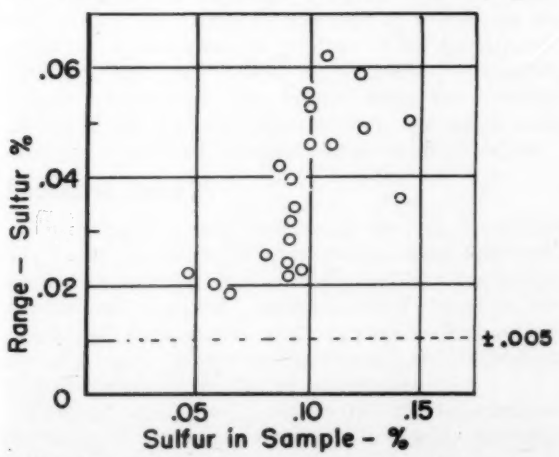


Fig. 11 — Relationship of sulfur content of iron to range in reported sulfur analyses.

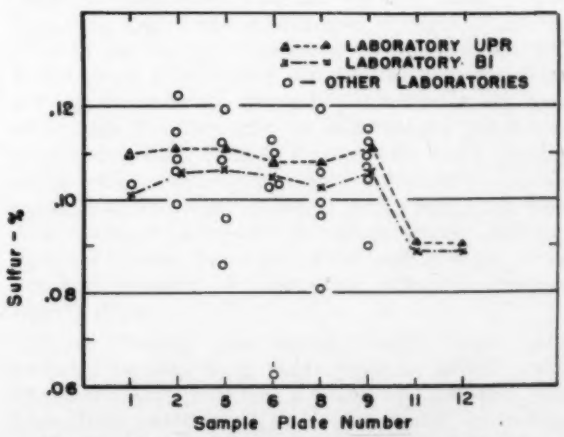


Fig. 12 — Distribution of sulfur analyses in relation to sample plate casting number, sample S-8-49.

Frequent recycling of samples under a different number will be helpful in evaluating the accuracy of the analyst as well as the method. A plant sample prepared in the customary manner will prove more useful than an outside standard for this purpose.

ACKNOWLEDGMENT

The author wishes to acknowledge the large amount of analytical data supplied by members of the Gray Iron Research Institute, Inc., for the chemical analysis surveys. Participation in the reproducibility investigation of high carbon cast iron samples by members of the Task Force on Carbon in Pig Iron of Subcommittee 1 on Pig Iron of A-3 on Cast Iron of A.S.T.M. is gratefully acknowledged. The helpful suggestions and comments of members of the staff of the British Cast Iron Research Association are also appreciated.

REFERENCES

1. W. E. Clarke, "The Significance and Use of Chemical Results," Research Report No. 255, *B.C.I.R.A. Journal of Research and Development, vol 3,* No. 6, pp. 381-386, June 1950.
2. Lundell, Hoffman and Bright, *Chemical Analysis of Iron and Steel,* 1931 edition, John Wiley and Sons, Inc., New York, Fifth Printing 1950.
3. W. Westwood and A. Mayer, *Chemical Analysis of Cast Iron and Foundry Materials,* George Allen and Unwin, Ltd, Ruskin House, London (Currently out of print, a revised edition is in preparation), 1951.
4. H. Green, Private communication relative to accuracy and reproducibility of Chemical Analysis of Cast Iron, Analytical Laboratory, B.C.I.R.A.
5. W. E. Clarke, "The Sampling of Pig and Cast Iron for Carbon Determination," Research Report No. 350, *B.C.I.R.A. Journal of Research and Development, vol. 4,* No. 11, pp. 496-519, April 1953.
6. Extended Abstract of Report of Subcommittee on Analysis of Pig Iron, Carbon and Low Alloy Steels, British Iron and Steel Association, *"Sampling for Analysis," Foundry Trade Journal, 364,* pp. 357-358, Sept. 25, 1952.
7. R. E. Deas and L. T. Conradi, "Determination of Total Carbon in Pig Iron, High-Carbon Iron, and Nodular Cast Iron," *Foundry,* pp. 68, 69, July 1949.
8. R. Rew, "Apparatus for Carbon Content Determination," advance copy of report to be published in *B.C.I.R.A. Journal of Research and Development.*
9. W. B. Sobers, "A Combined Method of Chemical Analysis for Cast Iron and Steel," AMERICAN FOUNDRYMAN, *vol. 6,* No. 1, pp. 2-5, Jan. 1944.
10. *A.S.T.M. Methods for Chemical Analysis of Metals,* American Society for Testing Materials, Philadelphia. 1956 Edition.

WHAT THE CASTING DESIGNER AND BUYER EXPECT OF THE FOUNDRYMAN

By T. O. Kuivinen

ABSTRACT

The need for more knowledge by designers of foundrymen's problems, and by foundrymen of designer's problems is discussed. The author feels that closer communication between the two parties will result in a better cast product, better in design and properties necessary for the application, if there is a better understanding between designer and foundryman. The various needs of the designer as they relate to the foundryman are given.

INTRODUCTION

The relationship between a designer and a foundryman may be difficult to discuss in its entirety. Of necessity this discussion will be limited to the heavy machinery industry where low volume run of a single part is the rule rather than the exception. In the author's company's business, the manufacturing of large engines and compressors, the Engineering Dept. is both the designer and the buyer. The corporation has captive foundries producing several grades of irons, and purchases steel, bronze, aluminum and magnesium castings from outside foundries. In all cases castings are designed, specifications are written and inspection requirements are issued by the same department.

Each person may feel he has his own sphere of operation, but actually the designer and the foundryman have the common objective of furnishing the customer a functioning product. The foundryman measures his success not in shipment of capacity volume of castings, but in providing the ultimate customer, the product user, with a top performing product. The casting designer has the same objective.

PROBLEM AREAS

This topic is a most interesting one. As the author sees it, the problem involves several areas that need discussion; strength in the casting, casting soundness, dimensional control, communication between designer and foundryman and economics. Perhaps all of these areas also need discussion in connection with the desires of the casting designer.

Through long association with the foundryman the designer develops confidence in the foundryman producing strength, soundness and required dimension.

Through the foundry's laboratory facilities the designer obtains data concerning tensile strength, shear strength, endurance strength in fatigue, and other properties of the materials used in various product parts. Such information is quite valuable and is used during design of the casting. Standard procedures as outlined by the A.S.T.M. are widely accepted. But, the A.S.T.M. procedures are directed at a test bar cast attached or separately at the time the casting is made. Usually all the data offered to the designer are determined from such test bars.

INFORMATION NEEDED

The designer needs to have assurance that the strength level in the casting will be at the level used during his design computations. He is entitled to expect the metal in the casting to be at strength levels attainable repetitively in production castings. The A.S.T.M. procedures and specifications relate only to the quality of the metal in the test bars. Strength levels determined from test bars may or may not be representative of the strength in important sections of the casting. Many factors in casting production may result in large differences in strength.

For example, the casting will probably have different cooling rates than the test bar faces. Therefore, one might expect different strength characteristics in the casting than found in the test bar. The foundryman should have suitable facilities for determining the material strength attainable in all types of castings. Yet there are no standardized procedures established. Each foundry's laboratory must develop means for determining and correlating casting strength to the values obtained from control test bars. The author's company produces many different types of parts. Some of these are always being scrapped because of machining errors or various types of flaws.

Being complex and usually fairly large, such scrapped castings have many perfectly sound areas. Therefore, such castings are cut up and test bars from representative sections are made to collect strength data and correlate such data to information obtained from the usual control test bar.

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Methods for doing this for various materials and various types of castings and the findings could be the background for development of standard procedures by organizations such as A.S.T.M. Any such study and correlation must be a continuing effort in order to maintain worthwhile usable data.

DETERMINE STRENGTH LEVELS

The foundryman can instill a high level of confidence by determining strength levels obtainable in castings. Through correlation of this knowledge with test bar data, he gives the designer a feeling of confidence in what he can expect in the castings he has ordered.

A more difficult problem is attaining assurance that all castings are sound. Shrink areas, dross and blow holes are not tolerable in the modern machine. The load demands on current products are high. Careful design for good stress distribution in limited space is essential. Ability to withstand dynamic loads without fatigue failure is absolutely necessary. Stress concentrations caused by design shape and function requirements cannot be avoided entirely. Therefore, stress concentrations caused by shrink or blow holes introduce quite high risk of casting failure in service. Experimental work is required to find means for elimination of shrink and dross. Often the latter can be handled by design alteration so that the dross occurs in an area where it is removed by machining.

GUIDANCE NECESSARY

Service failures due to flaws discourage repeat orders for a product just as well as failures due to poor design. Neither designer nor foundryman is interested in losing business, but both will if castings fail in service because of lack of soundness. An understanding of the problems involved in making sound castings is required of the designer, therefore, the designer expects this knowledge to come from association with the foundryman. The designer needs the foundryman's guidance during the casting planning stage so that foundry problems are minimized. Careful design study from the casting viewpoint alone is just as essential as from the functional viewpoint.

Once a decision is made future change may be difficult, for redesign of the entire machine might be necessary. The designer may expect soundness in castings produced by the foundryman but he does not expect it without consideration of the foundryman's problems while the casting is in the design stage. Quite often a casting design may be arranged for best location of feeder heads and gates. Wall intersections may be studied and modified to avoid hot spots and shrink porosity. Other requirements are considered in like manner.

Dimension is probably more easily controlled than strength and soundness and it is just as important. The designer today expects close adherence to the dimension and shape specified. As previously stated, space limitations in the highly loaded modern machine force design practices that utilize to the limit all sections of the casting. Design skills have increased through the use of stress-measuring equipment such as electric strain gages, brittle lacquer, and photoelasticity. One has to almost run to keep pace with the new developments in loading and stress-measuring techniques.

Research laboratories are also providing new knowledge concerning the behavior of materials under various types of loading. Nearly all machinery is subject to cycling load in a specific manner. Information is now available concerning the capability of material to endure such loading without failure in the lifetime desired for the machine.

The improved designs resulting from using such modern techniques places greater demand on the foundryman to control production closer to specifications. With less "factor of ignorance" in the "safety factor" the designer obtains greater performance from less space and material by squeezing down on the safety factor. This requires careful adherence to specified dimension, and assurance of specified strength in the casting.

NEED FOR COMMUNICATION

How is all this attained? An attempt will not be made to answer the question as such technical matters are not within the scope of this topic nor within the realm of the author's experiences. This current topic concerns what the designer expects to have in the castings produced by the foundryman. There is an area, however, that can be included under this topic which may aid in answering the question. One of the ingredients for success can be the development of better communication between the designer and the foundryman. The need for communication was indicated earlier. Confidence is developed through continuing association.

Know Each Other's Problems

Naturally the designer cannot be an expert foundryman. But, he must understand the manufacturing problems of the foundryman so that he can create designs that can be produced successfully and economically. He needs to consult the foundryman during the designing of a part so that the part is producible as well as functional. Often more than one way is found to design a part. The design layouts are discussed with the foundryman and the patternmaker to resolve the questions of production in order to have sound, flaw-free castings at lowest possible cost.

The foundryman may suggest an alternate approach to make feeding easier, or to minimize cleaning time, and so on. At this stage the foundryman learns of the quantity to be produced, for that may affect the type of pattern to be built or the molding method to be used. Either of these could influence the design shape and dimensions. Many other factors can be clarified, all in the interest of desired performance in the product at minimum manufacturing cost.

In like manner the foundryman is not expected to be an expert designer, but he is entitled to know something about the designer's problems and the function of the casting he is to produce. Effective communication between the two helps to attain the

common objective. Thus, the product is the result of team effort on the part of both foundryman and designer. A successful team must have continuous communication among the several specialists who contribute to the manufacturing of the product that the customer, who is the ultimate user, will buy. The man who deals directly with the foundryman is only his immediate customer. The real customer, the product user, measures the success of the foundryman-designer team by the performance of the casting in the product.

The author's company's experience in the manufacture of large heavy-duty machinery has demonstrated the need for close communication with the foundryman. Through consulting, experimenting and correlating information the foundryman builds the high level of confidence which the designer must have. I believe it can be safely said that some manufacturers have substituted welded fabricated parts where castings had been used as their designers have lost faith in castings because of some poor performance. The company believes the casting process is a reliable economical method for producing machinery components. It is also believed that there is a place for weldments in their machinery.

FABRICATION METHOD SELECTION

The designer makes the decision as to which method of fabrication is used by knowing the service function and the manufacturing costs involved. The art of creating and producing machinery for the best performance at minimum cost must include consideration of forgings, rolled stock, weldments, and castings in the many different types of materials available for all of these processes. Each has its place. The castings' position is a real one and it is growing.

No attempt has been made to discuss specific examples of successes and failures in the use of castings. All of us have seen both in various degrees. It is hoped that I have conveyed the designer's and buyer's messages and desires. I hope this discussion will help you to understand the need for desired strength in the casting and correlation of this information to that obtained from control test bars. Sound castings are essential in modern machinery, and compact design requires dimensional control. These are the things the designer expects of the foundryman. I am sure the foundryman is willing to meet these requirements.

DEVELOPMENT OF LOW ALLOY STEEL COMPOSITIONS SUITABLE FOR HIGH STRENGTH STEEL CASTINGS

By H. R. Larson and F. B. Herlihy

ABSTRACT

In order to produce high integrity castings with the best possible alloys two problems must be solved:

1) Find the best available materials.
2) Produce high integrity castings in which metal quality approaches as closely as possible that obtained in the material investigation.

The results of research into the first of these two problems are given, pertaining to obtaining low alloy steel composition with improved mechanical properties.

INTRODUCTION

It is generally agreed that the casting process has many potential advantages as a method of fabricating structural components of airframes. Many of these components have complicated shapes which are difficult or expensive to produce as forgings, machined shapes, weldments or combinations thereof. For this reason aircraft producers foresee considerable savings in time and manpower in utilizing cast shapes in the medium and ultra high strength range.

Although considerable work has been done on wrought ultra high strength steels, there has been little work on corresponding cast counterparts. Also even less has been done to determine what modifications must be made in the analysis and heat treatment of the wrought grades to develop optimum properties in the cast grades. Because of problems inherent in the casting process, all phases of production including analysis, melting, deoxidation and heat treatment must be studied.

The ultimate goal is to produce high integrity castings with the best possible alloys. This goal can be conveniently subdivided into two problems which are interdependent but which can be attacked separately.

The first is the problem of finding the best materials available. This includes finding the best analysis, melting practice and heat treatment. These factors are evaluated using test bars which are simple castings designed to eliminate casting flaws, particularly microshrinkage. Only in this way can the ultimate properties of the alloys be determined.

The second problem is to produce high integrity castings in which metal quality approaches as closely as possible that obtained in the material investigation.

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Results of an investigation directed towards the first problem, obtaining low alloy steel compositions with improved mechanical properties are presented herein.

PROCEDURE

A study was made of the effect of variations in carbon, nickel, chromium, molybdenum, vanadium and silicon on a modified A.I.S.I. 4300 base analysis of

Mn, %	Ni, %	Si, %	Cr, %	Mo, %
1.00	1.85	0.45	0.90	0.40

These heats were made with conventional or normal purity raw materials.

A number of heats were also made with high purity raw materials. Both tensile and impact tests were performed to evaluate quality of the various compositions. As a result of these experiments a leaner alloy corresponding approximately to the A.I.S.I. 8700 series was developed.

Heat treatment variables were also studied including homogenization at 2280 F and the effects of various austenitizing and tempering treatments.

Melting and Deoxidation Practice

Most of the heats were made in induction furnaces having rated capacities of 100, 200, 650 and 1000 lb. The two smaller furnaces had magnesia crucibles; the larger furnaces had a proprietary rammed (MgO. Al_2O_3) lining. Several heats were made in a 1000 lb electric arc furnace.

Heats of normal purity were made with low carbon steel punchings which were tumbled to remove dirt and rust. Other materials included pig iron, ferro-alloys and electrolytic nickel. Foundry returns up to 50 per cent of the charge was used whenever possible.

A layer of scrap was placed on the bottom of the furnace followed by pig iron, ferro-alloys and nickel. Larger pieces of foundry returns were then added, and, finally, the balance of the returns and steel punchings were added.

After the bath was heated to 3000 F to 3050 F the ferrosilicon and ferromanganese were added. In 3 to 7 min, depending on furnace size, the bath reached the normal tapping temperature of 3200 F. All heats were tapped into basic lined ladles. CaMnSi in the amount

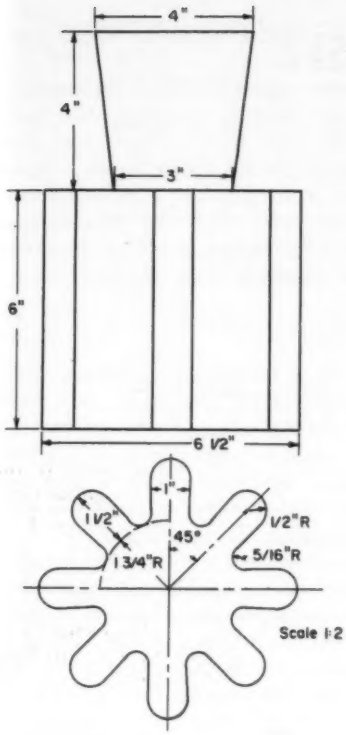


Fig. 1 — Test Coupon design for cast steel.

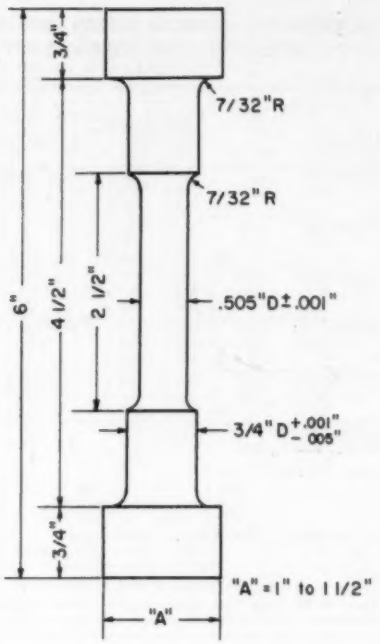


Fig. 2 — Shouldered round tension test specimen.

TABLE 1 — CHARPY V-NOTCH AND IZOD V-NOTCH IMPACT STRENGTH COMPARISON

Heat	Tempering Temp.	Room Temp. Izod	Room Temp. Charpy	−40 F Izod	−40 F Charpy
58-182	400 F-6	14.0, 15.0	14.0, 13.5	15, 17.5	17.5, 17
	1000 F-2	18.5, 18.5	18.0, 15.5	18.0, 17.5	

of 6 lb/ton was added to the bottom of the ladle just prior to tapping. Aluminum wire (0.06 per cent) was wrapped on a steel rod and plunged at about half tap.

For high purity heats either high purity sponge iron briquets or electrolytic iron was used. Graphite was used in place of pig iron, and electrolytic chromium and manganese were used in place of corresponding ferro-alloys. Sponge iron heats formed considerably more slag on melting which was skimmed off prior to the silicon and manganese additions. Deoxidation was the same for high and normal purity heats.

TEST BAR PRACTICE

A drawing of the casting used to obtain test coupons is shown in Fig. 1. This "octabar" casting was designed for use with cast steel and yields eight1x1x6 in. test coupons/casting.

Most of the tensile results were obtained from test bars with a 0.505 in. diameter and a 2 in. gage length. They were shoulder grip rather than treaded specimens. A drawing is shown in Fig. 2. Yield strengths were determined at 0.2 per cent permanent offset from an automatically recorded stress-strain curve.

Impact tests were done on multiple V-notch Izod specimens. They were about 6 in. long and had a 0.394 in. square cross-section. Each had four standard V-notches. Generally two breaks were made at room temperature and two at −40 F.

The Izod strength and Charpy V-notch impact strength are comparable in the region of 15 to 20 ft-lb as shown in Table 1.

RESULTS

The experimental work is described in the following sections. Alloy variations were made on the 4300 base analysis and tested in tension and impact. These heats were made with normal purity charge materials.

Alloy Variations — Carbon

Table 2 and Fig. 3 present data for a series of 4300 alloy steels with carbon contents ranging from 0.17 to 0.45 per cent and heat treated to the ultra high strength level. With the exception of the 0.17 per cent carbon specimen, these test bars were homogenized at 2280 F, normalized from 1850 F, oil quenched from 1650 F and tempered at 400 F. Test bars from the 0.17 per cent carbon heat were normalized from 1850 F, water quenched from 1675 F and tempered at 400 F. The increase in strength and decrease in ductility with increasing carbon follow the expected trend.

The maximum yield strength at about 0.38 per cent carbon and subsequent decline at higher carbon contents are believed to be the result of appreciable quantites of retained austenite.

The data indicate that the 0.17 per cent carbon level is most favorable with respect to impact strength, although the impact strength remains above 15 ft-lb to at least 0.36 per cent carbon.

Table 3 presents data for carbon variations at the 180,000 psi tensile strength level. Tempering temperatures were chosen according to carbon content to obtain this strength level, and with one exception were in the range 850 F to 1050 F. The exception

TABLE 2 — CARBON EFFECT ON PROPERTIES OF MODIFIED 4300 SERIES ALLOY AT HIGH STRENGTH LEVEL

Heat and Bar No.	C, %	Yield Str., ksi	Tensile Str., ksi	El., %	R.A., %	Y.S. T.S.	Izod Impact Strength (ft-lb)* Room Temp.	—40 F
58-326-3A	0.17	150.3	182.5	9.0	40.4	0.82		
-3B	0.17	156.6	189.5	10.5	40.7	0.82		
	Av.	153.4	186.0	9.8	40.6	0.82	25.0	24.5
57-296-5D	0.25	187.5	231.0	9.0	27.8	0.81		
-6D	0.25	183.0	230.0	9.0	26.1	0.79		
	Av.	185.2	230.5	9.0	27.0	0.80	17.6	16.5
57-287-3G	0.30	202.5	254.0	8.5	23.0	0.80		
-3H	0.30	198.6	253.5	8.5	20.6	0.78		
	Av.	200.6	253.8	8.5	21.8	0.79		
57-386-2C	0.35	207.0	263.0	8.5	16.3	0.79		
-2D	0.35	207.0	264.5	8.5	18.4	0.78		
	Av.	207.0	263.8	8.5	17.4	0.78		
57-306-3B	0.36	204.0	268.0	9.0	19.5	0.76		
-4C	0.36	216.0	269.0	7.5	13.0	0.80		
	Av.	210.0	268.5	8.2	16.2	0.78	16.0	15.5
57-310-3B	0.41	210.0	281.5	6.0	11.1	0.75		
57-317-3B	0.45	197.4	288.0	4.5	7.7	0.69		

Heat Treatment: 2280 F — 3 hr — air cool
1850 F — 3 hr — air cool
1650 F — 2 hr — oil quench
400 F — 6 hr — air cool

Except Heat 58-326: 1850 F — 3 hr — air cool
1675 F — 2 hr — water quench
400 F — 6 hr — air cool

*Ave. of at least two breaks at each temperature.

Heat 58-326 used high purity melting stock; other heats normal purity.

was the previously mentioned water quenched 0.17 per cent carbon heat which had a tensile strength of about 180,000 psi after tempering at 400 F. Although the impact strength is quite high at 0.17 per cent carbon, the yield:tensile ratio is low because of the low tempering temperature.

The impact strength is lowest with 0.25 per cent carbon, lower in fact than the 0.25 per cent carbon test bar tempered at 400 F, as may be seen by comparing data in Tables 2 and 3 for heat 57-296. This is explained by the fact that impact strength passes through a minimum between tempering temperatures of 400 F and 900 F, as shown in Fig. 17. Thus, the relatively low tempering temperature of 850 F used to obtain a 180,000 psi tensile strength at 0.25 per cent carbon results in low impact strength. Impact strength remains above 15 ft-lb for 0.35-0.39 per cent

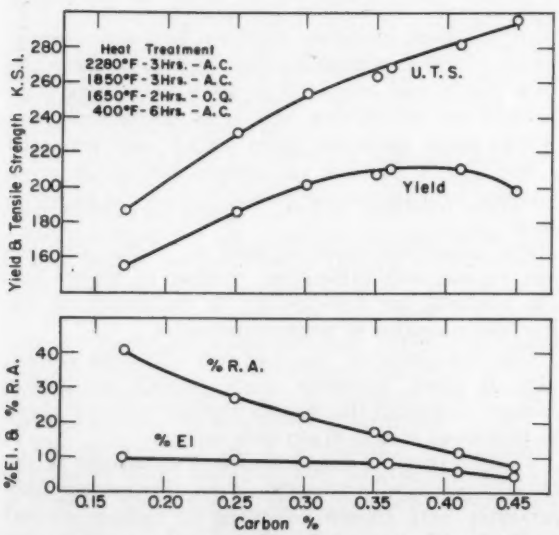


Fig. 3 — Carbon effect on properties of modified 4330 series alloy at high strength level.

carbon, and elongation remains at about 10 per cent through 0.45 per cent carbon.

In summary, it seems apparent that a carbon content of 33 to 38 per cent is optimum for both strength levels. For the ultra high strength level this carbon content yields maximum tensile strength without a drop-off in yield strength, and adequate impact strength and ductility. For the 180,000 psi strength level, this carbon content yields an optimum combination of yield strength, ductility and impact strength.

Nickel

Figure 4 presents data for nickel variations from zero to 2.46 per cent. Tensile specimens were heat treated to both the 180,000 psi tensile strength level and the ultra high strength level. Since the carbon contents of these heats varied from 0.33 to 0.38 per cent, the data at the higher strength level was corrected to 0.34 per cent carbon by using Fig. 3.

Nickel has only a minor effect on tensile strength or ductility, although there may be a slight tendency at the higher strength level for the tensile strength to decrease with increasing nickel. There is a rather definite tendency at both strength levels for the yield strength to decrease with increasing nickel.

Chromium

Data for chromium variations are shown in Table 4, although nickel and manganese variations are also present. It is apparent that chromium variations between 0.56 per cent and 1.51 per cent do not have any marked effect on tensile properties in the section sizes involved. At the lower strength level (tempering at 975 F) the higher chromium taps have a slightly higher strength with a correspondingly lower ductility.

At the higher strength level, the data are conflicting, with one heat showing a pronounced decrease

TABLE 3 — CARBON EFFECT ON PROPERTIES OF MODIFIED 4300 SERIES ALLOY TEMPERED TO 180,000 PSI STRENGTH LEVEL

Heat and Bar No.	C, %	Temper, Temp., F	Yield Str., ksi	Tensile Str., ksi	El., %	R.A., %	Y.S. T.S.	Izod Impact Strength (ft-lb)* Room Temp.	—40 F
58-326-3A	0.17	400	150.3	182.5	9.0	40.4	0.82		
-3B	0.17	400	156.6	189.5	10.5	40.7	0.82		
		Av.	153.4	186.0	9.8	40.6	0.82	25.0	24.5
57-296-5G	0.25	850	168.6	179.5	7.0	17.7	0.94	15.0	12.5
57-287-2E	0.30	900	171.0	189.0	9.0	29.8	0.91		
-2F	0.30	900	177.3	195.0	8.0	24.4	0.91		
		Av.	174.2	192.0	8.5	27.1	0.91		
57-386-2A	0.35	975	164.1	186.0	10.0	29.2	0.88		
-2B	0.35	975	164.1	187.0	11.0	30.4	0.88		
		Av.	164.1	186.5	10.5	29.8	0.88	18.0	17.5
57-306-4F	0.36	975	169.8	185.0	11.0	28.5	0.92		
-5D	0.36	975	174.6	191.0	9.0	24.1	0.91		
		Av.	172.2	188.0	10.0	26.3	0.92	17.5	20.0
58-033-5A	0.39	1000	164.7	184.5	11.0	25.2	0.89	28.0	18.0
57-310-3E	0.41	1000	173.7	186.0	10.5	28.5	0.93		
57-317-2C	0.45	1050	167.7	186.0	10.0	23.7	0.90		

Heat Treatment: 2280 F — 3 hr — air cool
1850 F — 3 hr — air cool
1650 F — 2 hr — oil quench
Temper 2 hr as shown

Except: 57-317-2C — No 2280 F treatment
57-386-2A & 2B — No 2280 F treatment
58-033-5A — No 2280 F treatment
58-326-3A & 3B — No 2280 F treatment and water quench from 1675 F instead of oil quench; temper at 400 F for 6 hr.

*Ave. of at least two breaks at each temperature.

Heat 58-326 used high purity melting stock; other heats normal purity.

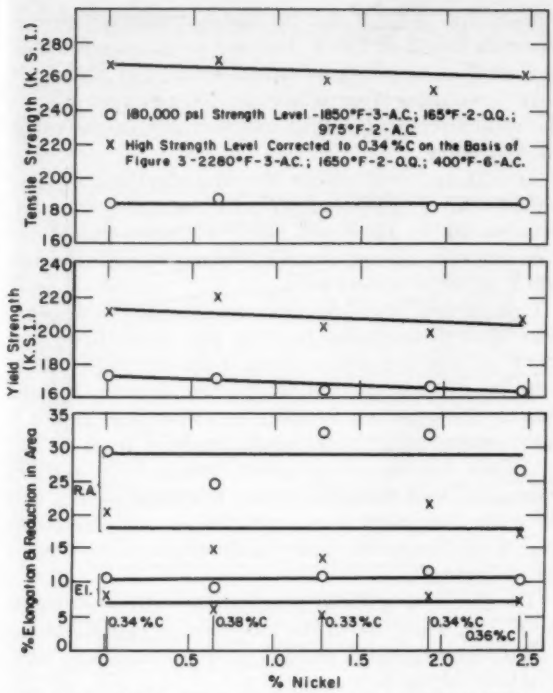


Fig. 4 — Nickel effect on properties of modified 4335 base analysis.

in yield and tensile strength with increasing chromium and the other heat showing almost no effect.

Molybdenum

When tempered at 975 F the high molybdenum tap (0.65 per cent) exhibits appreciably higher yield and tensile strengths than the lower molybdenum tap (0.28 per cent). However, this improvement is obtained at the expense of a decrease in ductility, as may be seen from Table 5. If the higher molybdenum tap were tempered at a higher temperature to give the same tensile strength as the lower molybdenum tap, the properties would probably be quite similar.

At the high strength level no effect from molybdenum is apparent from the limited data.

Vanadium

Figures 5 and 6 summarize data for vanadium variations from zero to 0.35 per cent. It is apparent that with a 975 F temper vanadium additions increase the yield and tensile strength but at the expense of decreased ductility. At the high strength level ductility is reduced with no appreciable gain in strength.

The effect of various tempering temperatures on the vanadium series is illustrated in Fig. 6. The alloy with 0.35 per cent vanadium may be tempered at 1150 F to approximately the same strength and ductility as the alloy with no vanadium tempered at 950 F. The Izod impact strengths are also comparable. However, with the higher tempering temperature the vanadium containing alloy could be used at a higher operating temperature in service.

Silicon

Two heats of normal purity 4340 were poured with four taps having silicon contents of 0.46, 1.56, 2.29 and 3.06 per cent. The data are plotted in Fig. 7. It is evident that silicon is detrimental when tempered at high temperatures (975 F), and that high silicon alloys have no potentialities at the 180,000 psi strength level.

With tempering at 400 F and 650 F the yield and tensile strength increase with increasing silicon, and the ductility shows a corresponding decrease. At the higher silicon contents 650 F tempering is superior, whereas at lower silicon contents 400 F tempering is superior. This is true of both yield strength and ductility. Although the yield strength behavior was expected, the difference in ductility appears too large at low silicon contents. This difference might be less if repetitive tests were made to eliminate the possibility of one or two slightly defective bars having an unduly large influence on the results.

Tempering Temperature

It is evident that the optimum tempering temperature is a function of silicon content. To determine this temperature, test bars from a number of high purity 8740 and 4340 heats were tempered in the range 400 F to 650 F. For convenience the data have

TABLE 4 — CHROMIUM EFFECT IN LOW ALLOY STEEL

Heat and Bar No.	Composition, % C	Mn	Cr	Ni	Yield Str., ksi	Tensile Str., ksi	El., %	R.A., %	Y.S. T.S.
High Strength Level*									
58-077-2D	0.34	1.00	0.56	0.67	219.9	270.0	4.5	9.3	0.81
-5D	0.34	0.99	1.46	0.67	202.8	256.6	4.5	9.6	0.79
58-080-2D	0.35	1.51	0.86	0.0	207.9	263.0	6.25	13.4	0.79
-5D	0.34	1.53	1.51	0.0	204.6	262.0	5.5	9.2	0.78
*Heat Treatment: 2280 F — 3 hr — air cool 1650 F — 2 hr — oil quench 400 F — 6 hr — air cool									
180,000 psi Strength Level†									
58-077-2A	0.34	1.00	0.56	0.67	171.0	181.5	9.5	26.8	0.94
-2B	0.34	1.00	0.56	0.67	168.0	178.5	10.25	28.5	0.94
					Av. 169.5	180.0	9.88	27.65	
58-077-5A	0.34	0.99	1.46	0.67	170.1	188.5	9.5	27.5	0.90
-5B	0.34	0.99	1.46	0.67	169.5	189.0	9.5	26.8	0.90
					Av. 169.8	188.75	9.5	27.15	
58-080-2A	0.35	1.51	0.86	0.0	160.2	177.0	10.5	28.8	0.91
-2B	0.35	1.51	0.86	0.0	164.1	178.5	11.0	30.5	0.92
					Av. 162.15	177.75	10.75	29.65	
58-080-5A	0.34	1.53	1.51	0.0	164.1	182.5	9.5	28.2	0.90
-5B	0.34	1.53	1.51	0.0	165.9	184.5	10.0	25.4	0.90
					Av. 165.0	183.5	9.75	26.8	
†Heat Treatment: 1850 F — 3 hr — air cool 1650 F — 2 hr — oil quench 975 F — 2 hr — air cool									

	Chemical Analysis, % P	S	Si	Mo	Al
58-077-tap 1	0.014	0.024	0.48	0.40	0.020
tap 2	0.015	0.026	0.42	0.40	0.013
58-080-tap 1	0.016	0.023	0.51	0.40	0.035
tap 2	0.019	0.023	0.47	0.40	0.062

TABLE 5 — MOLYBDENUM EFFECT IN MODIFIED 4335

Heat & Bar	C, %	Mo, %	Yield Str., ksi	Tensile Str., ksi	El., %	R.A., %	Y.S. T.S.
High Strength Level*							
58-082-2D	0.33	0.28	201.6	255.0	6.25	13.4	0.79
-5D	0.33	0.65	196.2	251.0	5.5	15.2	0.78
*Heat Treatment: 2280 F — 3 hr — air cool 1650 F — 2 hr — oil quench 400 F — 6 hr — air cool							
180,000 psi Strength Level†							
58-082-2A	0.33	0.28	155.1	168.0	11.75	34.1	0.92
-2B	0.33	0.28	159.3	171.0	11.75	32.5	0.93
-5A	0.33	0.65	172.5	193.0	10.5	27.6	0.89
-5B	0.33	0.65	174.6	195.0	10.5	26.1	0.90
†Heat Treatment: 1850 F — 3 hr — air cool 1650 F — 2 hr — oil quench 975 F — 2 hr — air cool							

	Chemical Analysis, % C	Mn	P	S	Si	Ni	Cr	Mo	Al
58-082-tap 1	0.33	0.99	0.016	0.026	0.44	1.91	0.93	0.28	0.028
tap 2	0.33	1.00	0.016	0.025	0.45	1.93	0.93	0.65	0.035

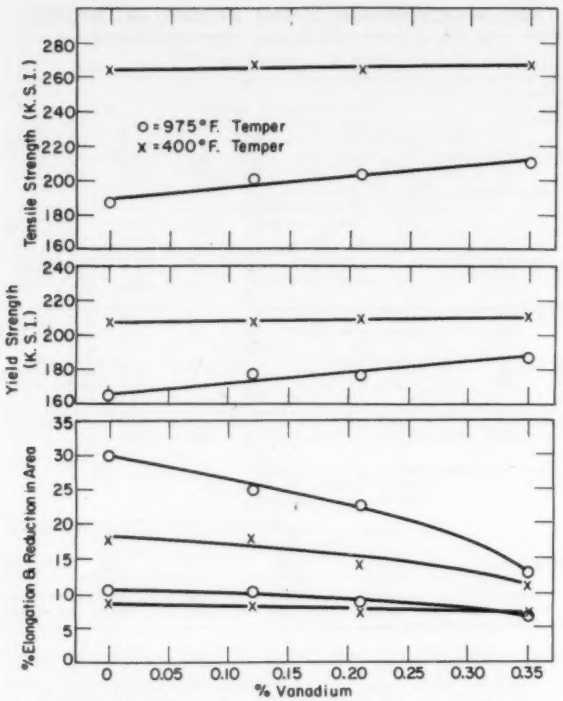


Fig. 5 — Vanadium effect on properties of modified 4335. Bars tempered at 975 F had 1850 F-3 hr-air cool; 1650 F-2 hr-oil quench. Those tempered at 400 F had a prior 2280 F-3 hr-air cool.

been plotted in Fig. 8 as three groups, with silicon contents of approximately 0.50, 1.25 and 1.90 per cent.

For the low silicon heats the yield strength reaches a maximum at 475 F. At 1.25 per cent silicon the maximum is raised to 525 F, but the yield strength does not decrease appreciably up to 650 F. For the three heats containing approximately 1.90 per cent silicon, the yield strength increases continuously in the range 400 F to 650 F.

At 1.25 per cent silicon the best ductility was obtained with a 525 F temper. However, the variation between 400 F, 525 F and 650 F was not marked. At 1.90 per cent silicon, ductility increases slightly from 400 F to 550 F to 650 F.

Considering both strength and ductility it appears that 525 F is optimum for 1.25 per cent silicon and 650 F at 1.90 per cent silicon. Data for high purity 4340 heats (not plotted in Fig. 8) showed approximately the same relationships. It should also be noted that austenitizing temperatures ranged from 1575 F to 1650 F for this series which would have a slight effect on yield strength. These optimum values and data for four high silicon 4340 heats are plotted in Fig. 9 as elongation and yield: tensile ratio versus tensile strength.

To obtain a reliable comparison with low silicon heats, Fig. 9 also shows the yield:tensile ratios and the range of elongations established for low silicon 4340 and 8740 in Fig. 11.

The ductility of the high silicon heats is on the low side of the scatter band established for low silicon heats. However, the yield:tensile ratio is increased from about 0.80 to 0.85 at a tensile strength of 280,000. This represents a change of about 14,000 psi in the yield strength. When the data are compared on the basis of equivalent yield strength rather than equiv-

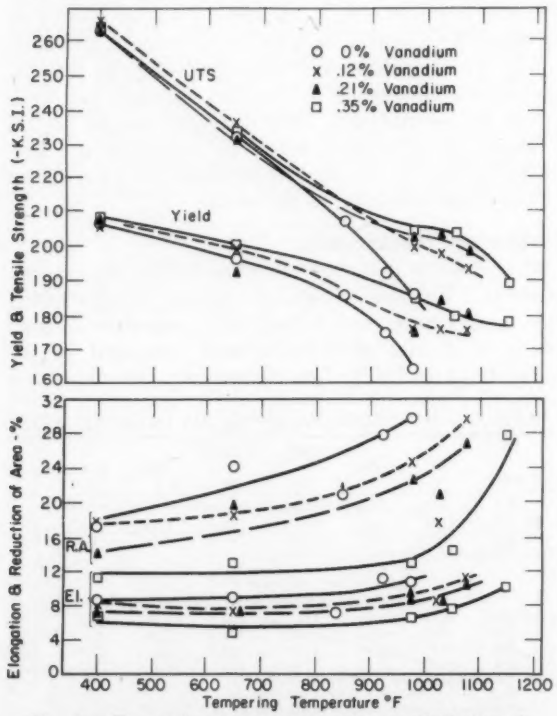


Fig. 6 — Tempering temperature effect on 4335 alloyed with vanadium. Bars tempered above 650 F had 1850 F-3 hr-air cool; 1650 F-2 hr-oil quench. Those tempered at 400 F or 650 F had prior 2280 F-3 hr-A.C.

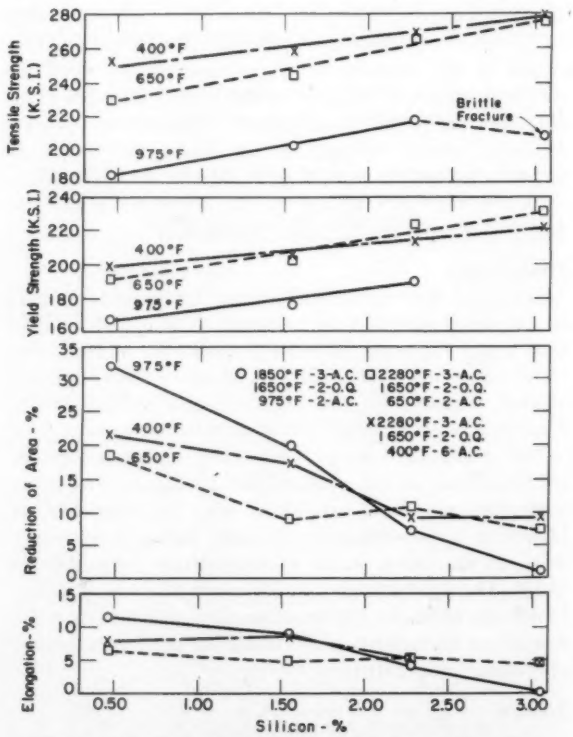


Fig. 7 — Silicon and tempering temperature effect on tensile properties of 4340.

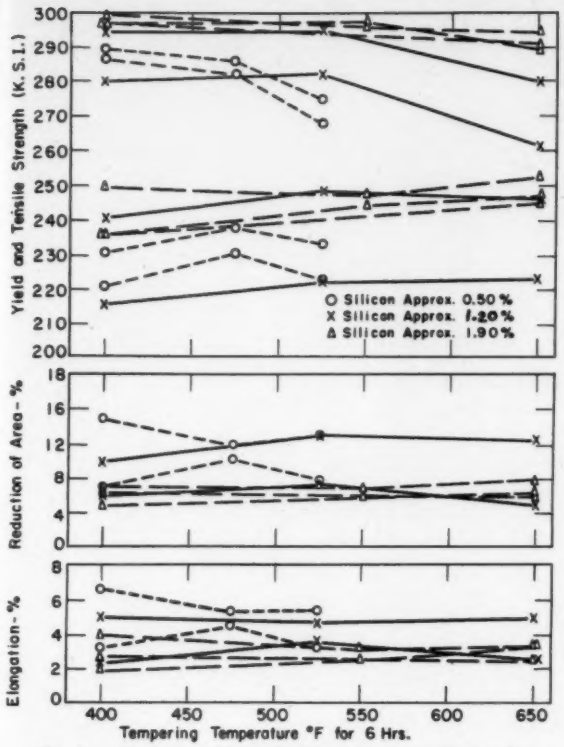


Fig. 8 — Tempering temperature effect on properties of high silicon, high purity 8740.

alent tensile strength, the ductility of the higher silicon heats is equal or better than that of the low silicon heats.

Modified 8740

Data developed during the study of alloy variations suggested the use of a leaner alloy mixture. Figure 10 indicates that the yield strength of 4340 reaches a maximum between 0.35 and 0.40 per cent carbon, and decreases thereafter. Also nickel shows a tendency to reduce the yield strength. It is believed that the combination of higher carbon and nickel promotes the retention of austenite during quenching. This austenite is not completely transformed by tempering at 400 F, and may transform to undesirable products when tempered at higher temperatures.

A series of heats was made with 0.50 per cent nickel corresponding approximately to A.I.S.I. 8740. Although such a composition could not be used in sections as large as would be possible with 4340, hardenability data shown in Fig. 16 indicate that the modified 8740 would harden through at least a 2 in. plate with an oil quench. This is adequate for many applications.

These 8740 heats were made with high purity charge materials, either electrolytic iron or sponge iron briquets.

Figure 10 shows the properties of these high purity 8740 heats as a function of carbon content, and a comparable set of properties for high purity 4340. It is apparent that at any particular carbon content the 8740 has a higher yield and tensile strength and lower ductility. For example at 0.40 per cent carbon the following are the typical values:

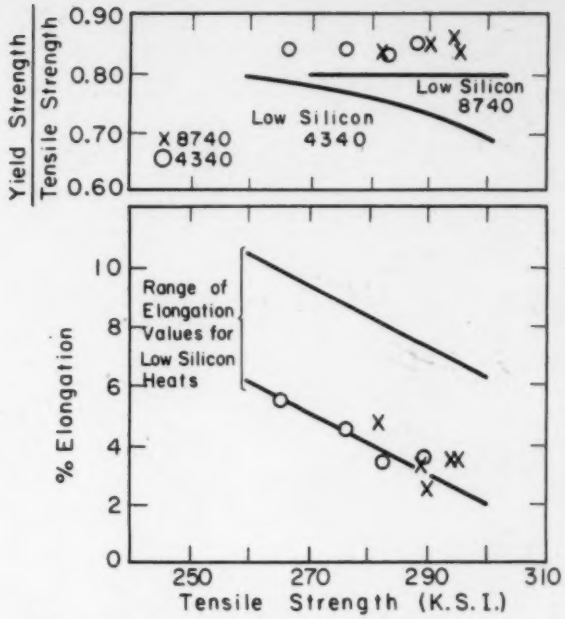


Fig. 9 — Comparison of high and low silicon heats of high purity 8740 and 4340.

	4340	8740
Yield Strength, psi	215,000	228,000
Tensile Strength, psi	280,000	288,000
Elongation, %	7	5
Reduction of Area, %	13.5	10

The yield strength of the 4340 decreases at higher carbon contents, as indicated earlier in Fig. 3. The

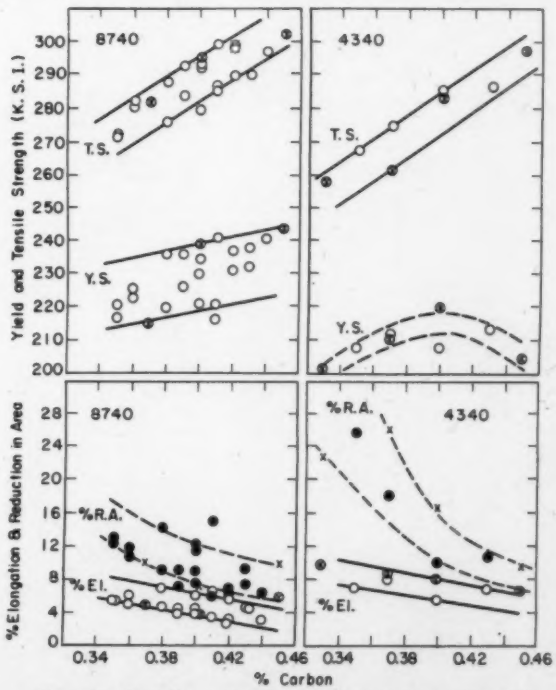


Fig. 10 — Tensile properties of high purity modified 8740 and 4340 as a function of carbon content. Heat treatment consisted of 1850 F-3 hr-air cool and 1525 F to 1650 F-2 hr-oil quench, followed by tempering at 400 F for 4 to 6 hr. *X* heats made with electrolytic iron, others made with sponge iron briquets.

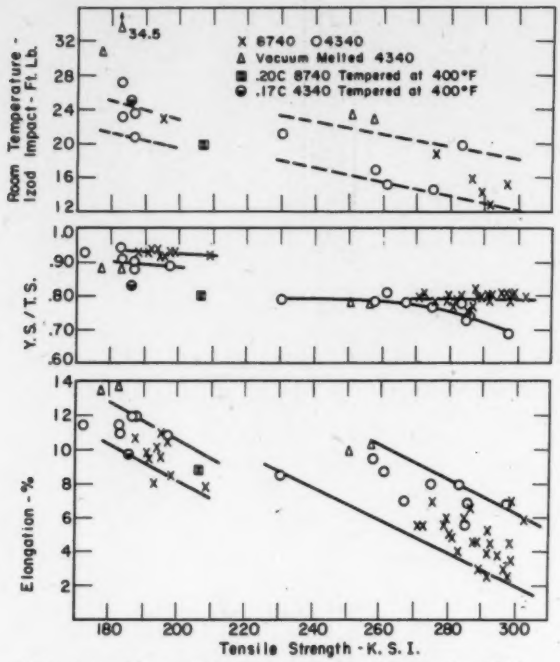


Fig. 11 — Mechanical properties of high purity 8740 and 4340 and vacuum melted 4340. Heat treatment as in Fig. 10.

yield strength of the 8740 continues to increase with increasing carbon content at least up to 0.45 per cent. It should be noted that the 4340 heats were treated before the austenitizing temperature was lowered to 1575 F. However, as Table 11 shows this does not change the yield strength enough to eliminate this difference in the two base alloys.

The data are presented in a different fashion in Fig. 11 where impact strength, yield:tensile ratio and elongation are plotted as a function of tensile strength.

TABLE 6 — RETAINED AUSTENITE DETERMINATIONS BY X-RAY DIFFRACTION

Heat and Bar No.	Temper	Yield Strength, psi	Tensile Strength, psi	El., %	R.A., %	Retained Austenite, % Based on (200) line	Retained Austenite, % Based on (220) line
58-181-3A	Single	205,500	287,000	6.0	10.4	10.4	10.3
58-181-3C	Double	210,300	284,500	5.75	9.2	7.7	7.8
58-185-2A	Single	220,200	286,500	6.0	13.8	3.8	4.7
58-185-2C	Double	226,800	291,500	4.25	7.3	4.3	4.3

NOTE: Heat Treatment for all samples as follows: 1850 F-3 hr-air cool; 1650 F-2 hr-oil quench; 400 F-4 hr-air cool; additional 400 F-4 hr-air cool for double tempered bars only. X-ray samples cut from shoulders of tensile bars.

Heat No.	Composition, % C	Mn	P	S	Si	Ni	Cr	Mo	Al
58-181	0.37	0.96	0.007	0.012	0.60	1.86	1.00	0.40	0.045
58-185	0.41	0.79	0.010	0.010	0.41	0.96	1.15	0.40	0.044

It is apparent that the 8740 heat (58-185) exhibits significantly less retained austenite than the 4340 heat (58-181). Double tempering is helpful in reducing the amount of retained austenite in 4335. It should be noted that the nickel content of heat 58-185 was higher than intended (0.96 per cent Ni rather than 0.50 per cent Ni).

The apparent ductility advantage of 4340 is less noticeable when the data are compared at the same strength level. It is also obvious that the yield: tensile ratio of 8740 remains constant at the high strength level whereas that of 4340 decreases. The leaner 8740 alloy also has a definitely higher yield:tensile ratio at the 180,000 psi level.

There is no noticeable difference in the Izod Impact strength.

Retained austenite determinations were made on samples of both 8740 and 4340 using an x-ray diffraction technique.* Results of these determinations are summarized in Table 6.

Izod Impact Strength

Izod Impact data at −40 F for normal purity 4340 and its alloy variations are compiled in Table 7 and plotted in Fig. 12. These heats are shown as the circled points, and the code number was added for convenience in checking composition in Table 7. All these heats were homogenized at 2280 F. The square and triangular points, referring to high purity 4340 and 8740, respectively, were for the most part given a homogenization treatment at 1850 F only.

Although there is considerable scatter in the data, bands have been drawn in indicate the high and low limits of the 4340 (and variations) data. There is a trend for impact strength to decrease with increasing carbon content.

*This technique is similar to that described by Triplett, Hauser, Wells and Mehl in W.A.D.C. Technical Report 53-518; "Determination of Retained Austenite by a Geiger Counter X-ray Technique;" Feb. 1954.

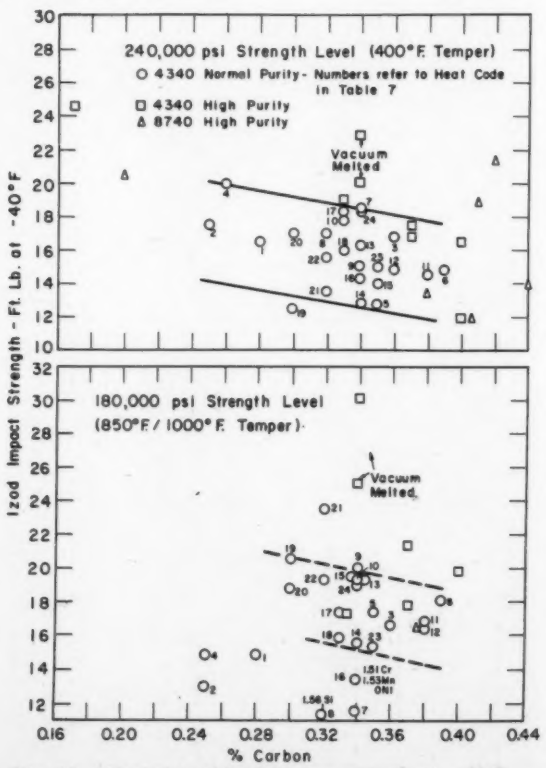


Fig. 12 — Variation of Izod impact strength at −40 F, with carbon content.

Generally the high purity heats are on the high side of the scatter band although most of them were homogenized at 1850 F rather than 2280 F. Also the vacuum melted heats are obviously quite superior.

Several heats tended to be on the low side of the scatter bands at both strength levels (nos. 14, 16 and 5). These heats contained 1.56% chromium plus 0.67% nickel, 1.5% chromium plus 1.53% manganese and no nickel and one standard heat, respectively. Numbers 9, 10 and 13 with 0.01% nickel, 1.29% nickel and 0.56% chromium plus 0.67% nickel, respectively, tended to be on the high side of the scatter band.

Three heats with carbon contents of 0.25 per cent

TABLE 7 — IZOD IMPACT STRENGTH OF NORMAL PURITY 4340*

Heat & Tap	Heat Code No.	C, %	P, %	S, %	Variations from Standard 4335	Heat** Treat.	Ft-lb† R.T.	Ft-lb† —40 F
57-296-1	1	0.28	0.007	0.028	—	B	16	16.5 sand mold
						D-850	17	14.8 sand
57-296-2	2	0.25	0.028	0.030	—	B	16.3	17.5 sand
						D-850	15.0	13.0 sand
57-306	3	0.36	0.013	0.027	—	B	17.3	20.5
						B	16.5	19.0
						B	15.0	14.0
						B	15.5	13.8
						D-975	17.0	19.5
						D	17.8	20.5
						D	16.3	14.0
						D	15.3	16.0
						D	15.0	13.5
57-355	4	0.25	0.022	0.016	—	B	19.0	20.0
						B	19.0	20.0
						D-850	17.5	14.5
						D	19.0	15.0
57-386	5	0.35	0.012	0.024	—	C	11.5	12.0
						C	11.8	13.3
						D-975	17.3	16.0
						D	19.5	18.8
58-033-2	6	0.39	0.020	0.028	—	D-1000	23.5	19.8
						A	17.3	14.3
						C	16.0	15.3
						D-1000	28.0	18.0
58-064-1	7	0.34	0.014	0.025	—	C	12.3	18.5
						D-975	18.0	11.5
58-064-2	8	0.32	0.018	0.025	1.56 Si	C	14.5	17.0
						D-975	12.5	11.3
58-069-1	9	0.34	0.016	0.029	0.01 Ni	C	13.5	15.0
						D-975	22.3	20.0
58-069-2	10	0.33	0.028	0.027	1.29 Ni	D-975	21.0	19.3
						C	16.3	17.8
58-071-1	11	0.38	0.020	0.024	0.65 Ni	C	21.5	14.5
						D	19.3	16.8
58-071-2	12	0.36	0.022	0.028	2.46 Ni	C	13.8	14.8
						D-975	17.3	16.5
58-077-1	13	0.34	0.014	0.024	0.56 Cr 0.67 Ni	C	14.8	16.3
						D-975	19.3	19.3
58-077-2	14	0.34	0.015	0.026	1.56 Cr 0.67 Ni	C	12.8	12.8
						D-975	18.3	15.5
58-080-1	15	0.35	0.016	0.023	0.86 Cr 0 Ni 1.51 Mn	C	12.8	14.0
						D-975	20.35	19.5
58-080-2	16	0.34	0.019	0.023	1.51 Cr 0 Ni 1.53 Mn	D	15.3	15.3
						D-975	14.8	13.3
58-082-1	17	0.33	0.016	0.026	0.28 Mo	C	17	18.3
						D-975	19	17.3
58-082-2	18	0.33	0.016	0.025	0.65 Mo	C	17.5	16.0
						D-975	16.3	15.8
58-084-1	19	0.30	0.028	0.027	0 Ni 0.0008B added	C	16.0	12.5
						D-975	24.8	20.5
58-084-2	20	0.30	0.026	0.027	0 Ni 0.0020B added	C	17.0	17.0
						D-975	24.5	18.8
58-093-1	21	0.32	0.015	0.028	0 Ni 0.0008B added	C	11.5	13.5
						D-975	24	23.5
58-093-2	22	0.32	0.024	0.028	0 Ni 0.0020B added	C	14.3	15.3
						D-975	22.5	19.3
58-049-2	23	0.35	0.025	0.030	—	C	12.8	15.0
						D-975	15.3	15.3
58-Arc-114	24	0.34	0.013	0.010	—	C	17.8	18.3
						D-975	20.0	19.0

*A) 1850-3-air cool
1650-2-oil quench
400-6-air cool

B) 2280-3-air cool
1850-3-air cool
1650-2-oil quench
400-6-air cool

C) 2280-3-air cool
1650-2-oil quench
400-6-air cool

D) 1850-3-air cool
1650-2-oil quench
Tempering as indicated for 2 hr

**Each entry is the average of two values

†Multiple breaks, two at room temperature and two at —40 F for each test bar.

TABLE 8 — PHOSPHORUS AND SULFUR EFFECT ON MECHANICAL PROPERTIES OF MODIFIED 4330

Heat & Tap	C, %	P, %	S, %	Heat Treat.	Yield Str., ksi	Tensile Str., ksi	El., %	R.A., %	Izod Impact R.T.	Izod Impact −40 F
180,000 psi Strength Level										
57-293	0.25	0.005	0.008	A	168.2	177.3	12.8	41.7	26.3	23.0
57-355	0.25	0.022	0.016	A	170.3	181.8	12.3	36.0	18.2	14.8
57-296 Tap 1	0.28	0.007	0.028	A	161.1	172.0	11.5	38.6	18.7	14.5
57-296 Tap 2	0.25	0.028	0.030	A	167.0	177.8	10.8	27.2	15.0	12.5
57-293	0.25	0.005	0.008	B	169.9	179.3	12.2	36.0		
57-355	0.25	0.022	0.016	B	165.5	178.8	11.5	36.6		
57-296 Tap 1	0.28	0.007	0.028	B	165.0	176.5	12.3	34.8		
57-296 Tap 2	0.25	0.028	0.030	B	169.1	178.0	11.0	32.2		
230,000 psi Strength Level										
57-293	0.25	0.005	0.008	C	182.7	230.5	8.5	23.7		
57-355	0.25	0.022	0.016	C	187.4	235.5	9.8	25.3		
57-296 Tap 1	0.28	0.007	0.028	C	181.0	227.0	8.0	24.6		
57-296 Tap 2	0.25	0.028	0.030	C	173.4	223.8	8.75	24.5		

Heat Treatments

A	B	C
1850 F-3-air cool	2280 F-3-air cool	1850 F-3-air cool
1650 F-2-oil quench	1850 F-2-air cool	1650 F-2-oil quench
900 F-2-air cool for tensiles	1650 F-2-oil quench	400 F-6-air cool
850 F-2-air cool for impacts	900 F-2-air cool	

and 0.28 per cent had low impact values when tempered at 850 F to obtain a tensile strength of 180,000 psi. From Fig. 17 it may be seen that impact strength after tempering at 850 F is less than that obtained at 400 F. For optimum properties, considering both tensile and impact strength, the carbon should be approximately 0.35 per cent, since at this carbon the tempering temperature is about 950 F.

MELTING

There are a number of somewhat intangible factors which influence mechanical properties which cannot be explained by variations in normally analyzed elements or heat treatment. Among these factors are inclusion type and distribution, gas content, trace impurities and probably many others. They make up what is often called the heat variable, i.e., apparently identical heats made with the same practice have a range of properties, sometimes from good to bad.

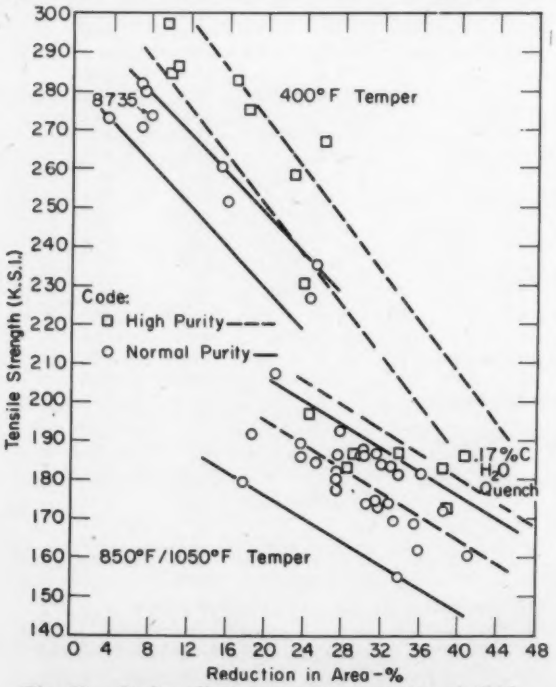


Fig. 13 — Purity effect on tensile properties of 4335. Heat treatment: 1850 F-3 hr-air cool; 1650 F-2 hr-oil quench; temper.

Sulfur and Phosphorus

Although these elements are analyzed on a routine basis, it seemed important to determine what effect variations had on mechanical properties. Accordingly three heats, one with split taps, were made using electrolytic iron with varying amounts of phosphorus and sulfur added as alloying elements. The results are shown in Table 8. At the 180,000 psi level and with homogenization at 1850 F only, the ductility drops off with increasing phosphorous and sulfur. The Izod impact strength drops almost 50 per cent.

The tensile specimens were tempered at 900 F, but the impact specimens were tempered at 850 F before it was realized that this was not the most desirable temperature. With 2280 F homogenization the decrease in ductility is quite small, and it is difficult to ascertain whether the change is significant.

At the 230,000 psi level, obtained by tempering at 400 F, there is no trend whatsoever for ductility to decrease with increasing phosphorous and sulfur. Unfortunately, there were not sufficient test bars to perform impact tests at this strength level. It is probable that with more data, some trend towards lower ductility at higher sulfur is bound to appear. It should also be noted that all of these heats would be considered high purity heats with respect to trace elements aside from phosphorous and sulfur.

Purity Effect

Since the heats made with electrolytic iron generally had good properties even when sulfur and phosphorous were added back, it would appear that purity of the charge has some other effect when merely being low with respect to these two elements. Accordingly a number of high purity 4340 heats (electrolytic iron or special sponge iron briquets) were specially prepared, and in addition some data were available from other programs being carried out in the laboratory. Data comparing high and low purity heats are plotted in Figs. 13 and 14 as tensile strength vs. reduction of area. Figure 13 includes the data for test bars homogenized at 1850 F, and Fig. 14 for bars homogenized at 2280 F; all followed by oil quenching from 1650 F. Tempering was done at 400 F for the high strength level, and in the range 850 F to 1050 F for the 180,000 psi level.

Since there is the usual scatter in the data, bands have been drawn to more readily distinguish the high and low purity groups. It is apparent at each strength level and with all heat treatments, that high purity increases the ductility for any given tensile strength.

It can also be seen that homogenization at 2280 F is beneficial at high strength levels, but the combination of a high purity and 1850 F is superior to normal purity and 2280 F. This is an important fact since heat treatment at 2280 F is quite expensive.

Figure 10 shows all the tensile data for high purity

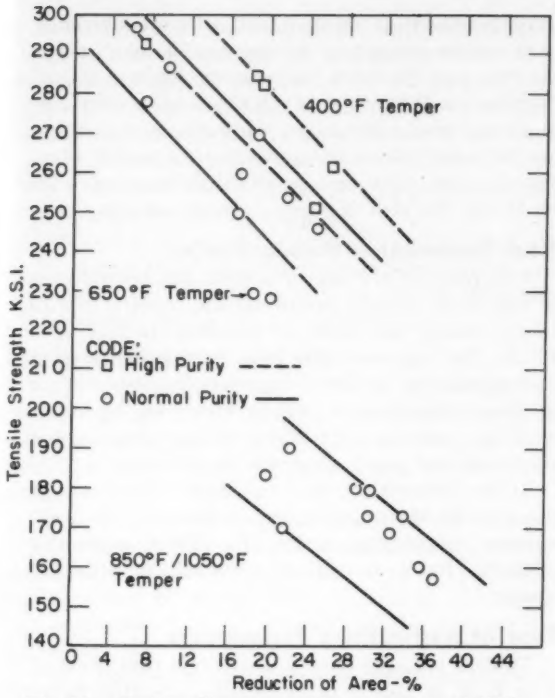


Fig. 14 — Purity effect on tensile properties of 4335. Heat treatment: 2280 F-3 hr-air cool; 1650 F-2 hr-oil quench; temper.

heats of both 4340 and 8740 plotted as a function of carbon content. Heats made with electrolytic iron are distinguished from those made with sponge iron. Of the four 4340 heats made with electrolytic iron, three have ductilities on the high side of the scatter band and one heat is about average. For high purity 8740, one is on the high side and two on the low side.

Arc vs. Induction Furnaces

Four arc furnace heats were melted. One heat made with conventional charge materials had ductility on the high side of the scatter band for normal purity heats. Of three high purity arc heats, two tended to have ductility on the low side and one on the high side. Thus, with this limited amount of data there were no appreciable differences between the two melting furnaces.

Vacuum Melted Heats

Two vacuum melted 4335 heats were melted. Tensile and impact data are plotted in Fig. 11. It is quickly apparent that these heats have superior tensile properties and particularly good impact properties.

For the first vacuum heat no effort was made to add hot top to the riser as is the practice with air melted heats. Samples were examined and some scattered microshrinkage was evident. For the second heat two test molds were poured. An attempt was made to hot top the first with a disc of exothermic material and the second by an electric arc. Neither attempt was a complete success. However, tensile and impact properties are equally good for both heats. Thus, both sets of bars are probably equally sound or unsound.

Inclusion Distribution

Micrographs are shown in Fig. 15, representing heats made by vacuum melting with a high purity charge and with a normal or low purity charge. In

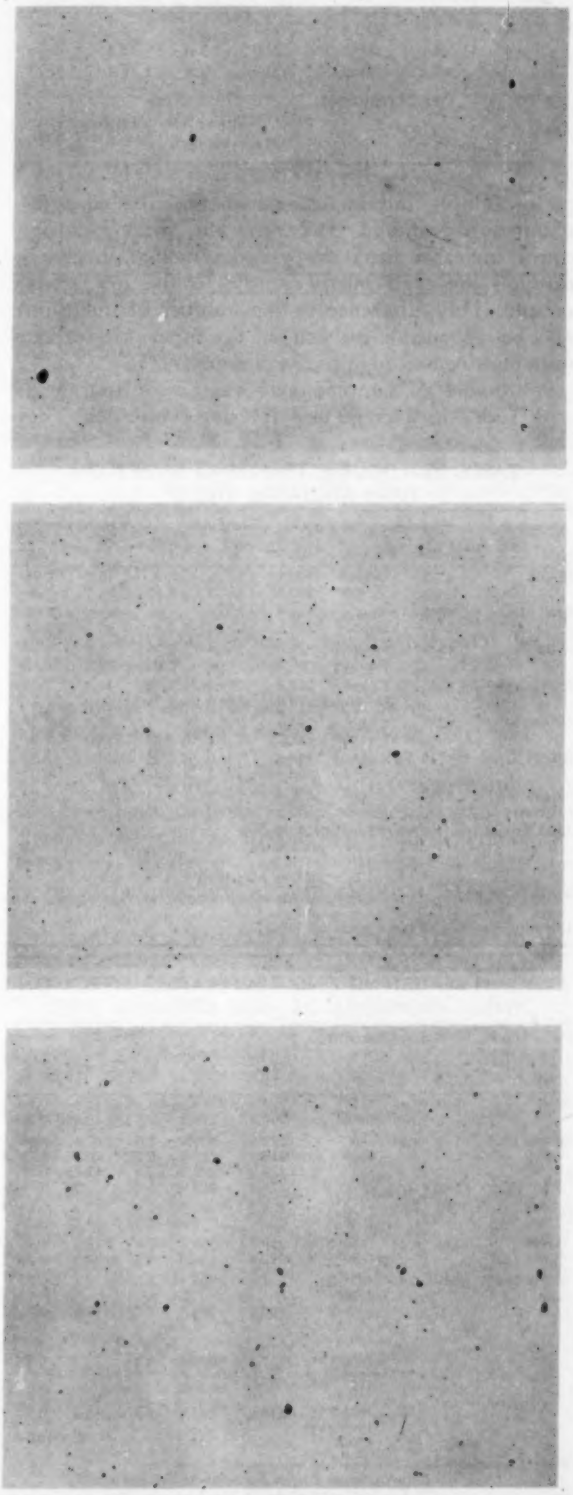


Fig. 15 — Inclusion distribution in 4340 of different purity levels. Top — vacuum heat, center — high purity 4340, bottom — normal purity 4340. 100×.

TABLE 9 — EFFECT OF TIME AT AUSTENITIZING TEMPERATURE

Heat and Bar	Time at 1650 F-hr	Yield Strength, ksi	Tensile Strength, ksi	El., %	R.A., %	Y.S. T.S.
58-306-5A	2	210.0	269.5	8.0	18.8	0.78
-5B	4	209.1	269.0	7.5	17.4	0.78
-5C	6	210.6	269.5	7.5	17.4	0.78

Heat Treatment: 2280 F-3 hr-air cool
1650 F-2, 4 or 6 hr-oil quench
400 F-6-air cool

each example the inclusions are globular and distributed at random. However, the total inclusion count increases from the vacuum melted specimen, through the high purity sample to the low purity sample. This difference in the number of inclusions may be responsible for some of the superiority of vacuum melting and high purity charges.

A number of samples were examined from heats with superior, average and inferior properties. They were examined at random and without reference to their tensile properties. It was possible to distinguish the best and the worst heats on the basis of inclusion distribution. The worst heat had a tendency for type 2, or intergranular inclusions. However, the correlation on the intermediate heats was not too good. Thus, it appears that inclusion type and distribution is a quality factor, but that it is not the only one operating.

TABLE 10 — DOUBLE TEMPERING EFFECT ON HIGH STRENGTH STEELS

Heat	Grade	Temper	Yield Str., ksi	Tensile Str., ksi	Elong., %	Red. of Area, %	Y.S. T.S.
58-064	4335	Single	202.4	206.5	8.0	15.2	0.78
		Double	200.0	252.3	8.63	21.1	0.79
58-089	4335	Single	*198.6	260.0	9.0	21.9	0.76
		Double	*200.1	256.0	9.0	19.2	0.78
58-126	4335	Single	213.3	286.3	6.75	10.6	0.75
		Double	207.8	278.8	6.0	10.4	0.75
58-Arc-181	4335	Single	207.8	285.3	5.63	10.1	0.75
		Double	209.9	284.3	5.75	10.0	0.74
58-Arc-182	4335	Single	211.7	275.0	8.0	18.1	0.77
		Double	211.7	273.8	7.25	16.5	0.77
58-185	8740	Single	220.2	286.8	6.63	14.9	0.77
		Double	225.9	288.8	4.88	9.8	0.79
58-186	8740	Single	230.4	289.8	3.25	7.0	0.80
		Double	228.5	289.0	4.13	9.4	0.79

Heat Treatment: 1850 F — 3 or 4 hr — air cool
1650 F — 2 hr — oil quench
400 F — 4 hr or
500 F — 4 hr — air cool plus 400 F — 4 hr — air cool

*Results shown are for single bars. All others are the average of two test bars.
Chemical analysis of these heats in Table 14.

TABLE 11 — AUSTENITIZING TEMPERATURE EFFECT ON TENSILE PROPERTIES OF HIGH STRENGTH STEEL

Heat	Grade	C, %	Austenitizing Temp., F	Yield Str., ksi	Tensile Str., ksi	El., %	Red. of Area, %	Y.S. T.S.
58-126	4340	0.43	1575	213.6	280.8	6.25	11.0	0.76
			1650	207.8	278.8	6.0	10.35	0.75
58-188	8740	0.36	1500	222.6	279.0	3.5	7.55	0.80
			1550	216.5	279.5	5.38	10.0	0.78
			1575	229.9	279.8	6.0	11.65	0.80
			1600	216.0	278.0	5.0	9.8	0.78
58-193	8740	0.36	1550	223.2	282.3	6.25	12.2	0.79
			1650	221.1	280.8	3.5	9.25	0.79
58-194	8740	0.40	1550	233.0	292.0	2.63	6.6	0.80
			1650	225.8	292.0	5.25	8.35	0.77
58-arc-181	4340	0.40	1650-2* 400-4	207.8	285.3	5.63	10.1	0.75
			1650-2* 400-4 400-4	209.9	284.3	5.75	10.0	0.74
			1650-1** 1520-1½* 410-4	217.5	288.0	4.4	7.7	0.76
			1650-1** 1500-1½* 410-4	216.2	289.0	6.5	9.6	0.75

Heat Treatment: 1850 F — 3 or 4 hr — air cool
oil quench after 2 hr at temperature indicated
400 F — 3 hr — air cool, 400 F — 3 hr — air cool

All results are the average of two tests
*Oil quench
**Air cool

Heat Treatment — Homogenization

This phase of the heat treatment has been discussed in a previous section describing the benefits of a high purity charge. The data are presented in Figs. 13 and 14. At the high strength level (tempering at 400 F), homogenization at 2280 F improves ductility compared to homogenization at 1850 F. However, high purity with heat treatment at 1850 F is equivalent or better than a normal purity charge with treatment at 2280 F.

At the 180,000 psi level the 2280 F treatment does not provide any improvement in ductility. Actually it appears detrimental unless the 2280 F treatment is followed by a normalize from 1850 F (data not shown).

Time at Austenitizing Temperature

Table 9 presents data that indicate that times of 2, 4 or 6 hr at 1650 F prior to oil quenching have no effect on the tensile properties of normal purity 4340. Unfortunately no more bars were available from this heat to investigate still shorter times.

Double Tempering

Tensile specimens from five heats of 4335 and two heats of 8740 were subjected to both single and double tempering treatments at 400 F for 4 hr. The data are presented in Table 10. Although there were slight differences within any one heat, there was no tendency for either single or double tempering to be superior to the other.

Variations in Austenitizing Temperature

Tensile bars from two heats of 4340, and three heats of 8740, were oil quenched from various temperatures in the range 1500 F to 1650 F. The data are presented in Table 11.

There is a tendency for lower temperatures to be superior, particularly with regard to yield strength. The effect was more pronounced for the higher carbon heats.

There was no apparent effect on ductility or tensile strength.

Since there is evidence of improved yield strength, and because quench cracking should be decreased at the lower temperature, standard practice was altered to include 1575 F for alloy steels with carbon contents in the vicinity of 0.35 per cent.

Two Stage Austenitizing Treatments

It has been reported in the literature* that two step austenizing treatments are beneficial from the standpoint of physical properties and in preventing quench cracking. These treatments involve holding

*William, J., Tyler, A. and Ahearn, P. A., "Investigation of Heat Treatments & Mechanical Properties of high Strength Steel Castings," Watertown Arsenal Report No. RPL 11/5.

at about 1600 F for an hour or so, furnace cooling to 1325 F to 1500 F, holding for a short time and then quenching. Table 12 shows data for six heats of high purity 8740 and one heat of 4340. The 8740 heats were oil quenched and the single 4340 heat was water quenched. Although a variety of temperatures and times were used, none of them seemed to provide any improvement in tensile properties.

Although there is no apparent benefit with regard to properties, there was also no detrimental effect; and these heat treatments might still be used to advantage where quench cracking is a problem.

Refrigeration Effect on 8740

Tensile specimens from two heats with high and low silicon taps were cooled to −105 F after being oil quenched from 1650 F. A comparison with unrefrigerated specimens from the same heats are shown in Table 13.

There was a pronounced increase in yield strength, a lesser increase in tensile strength and little change in ductility. In three of the four taps yield strength was increased by about 7000 psi, and in the fourth tap, containing 1.90 per cent silicon, the increase was 17,700 psi.

The yield:tensile ratio increased from 0.78 to 0.80 to 0.81 to 0.83 with refrigeration.

Hardenability

Jominy End Quench curves are shown in Fig. 16 for one heat of 4335 and two heats of 8740. The hardenability of the 8740 is quite high. It should be noted that modifications of 8740 and 4340 both contain higher chromium and molybdenum than the standard A.I.S.I. wrought grades. The A.I.S.I. designations are used, however, for convenience.

The hardness at the quenched end of the 0.36 per cent carbon heat of 8740 is higher than the 0.40 per cent carbon heat. This is unexpected, and in all probability is an experimental error.

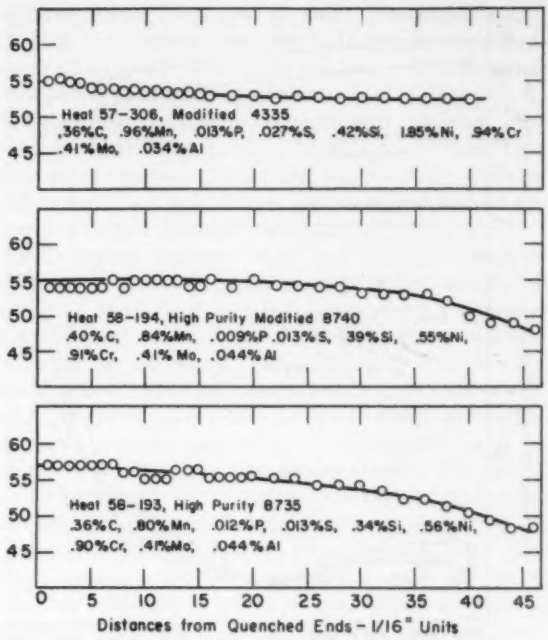


Fig. 16 — Jominy end quench hardenability of modified 4340 and 8740.

Tempering Temperature Effect on Hardness and Izod Impact Strength

Test bars from high purity heats of 8740 and 4340 were tempered for 6 hr each at 100 degree intervals in the range 300 F to 1200 F. Hardness and impact strength are plotted in Fig. 17. The difference between the hardness curves is unexpected since the carbon contents are almost identical. Also tensile

TABLE 12 — TWO STAGE AUSTENITIZING TREATMENTS EFFECT ON TENSILE PROPERTIES OF HIGH STRENGTH STEEL

Heat	C, %	Aust. Treat.	Yield Str., ksi	Tensile Str., ksi	El., %	R.A., %	Y.S. T.S.
**58-185-Tap 1	0.41	1650 F-2* 1650 F-2-Fce. Cool to 1500 F-1*	225.9 227.4	288.8 286.8	4.88 5.13	9.8 9.95	0.78 0.79
58-186-Tap 1	0.42	1650 F-2* 1650 F-2-Fce. Cool to 1500 F-1*	228.5 233.1	289.0 291.0	4.13 4.88	9.4 8.45	0.79 0.80
58-187-Tap 1	0.44	1575 F-2* 1650 F-2-Fce. Cool to 1475 F-1* 1650 F-2-Fce. Cool to 1350 F-½*	240.2 240.0 235.5	297.0 294.5 296.3	3.0 3.5 3.25	6.4 6.55 6.15	0.82 0.82 0.80
58-188-Tap 1	0.36	1575 F-2* 1650 F-2-Fce. Cool to 1475 F-1* 1650 F-2-Fce. Cool to 1350 F-½*	224.9 222.2 221.3	279.8 275.5 275.5	6.0 5.25 6.0	11.65 11.35 13.75	0.84 0.81 0.80
58-193-Tap 1	0.36	1550 F-2* 1650 F-2-Fce. Cool to 1475 F-1*	223.2 218.1	282.3 280.3	6.25 6.0	12.2 10.35	0.79 0.78
58-194-Tap 1	0.40	1550 F-2* 1650 F-2-Fce. Cool to 1475 F-1*	233.0 220.2	292.0 288.3	2.63 4.38	6.6 7.15	0.80 0.76
58-196	0.35	1575 F-2* 1650 F-2-Fce. Cool to 1350 F-½†	207.7 206.4	267.5 255.5	7.0 8.0	26.7 25.4	0.78 0.81

Heat Treatment: 1850 F — 3 hr — air cool
Austenitized and quenched as indicated
double tempered 400 F — 3 or 4 hr — air cool

Results shown for each heat treatment are the ave. of two tensile tests.
*Oil quench
†Water quench
**All heats of 8740 composition except heat 58-196 which is 4340.

TABLE 13 — REFRIGERATION EFFECT ON 8740

Specimen	Heat Treat.	Yield Str., ksi	Tensile Str., ksi	Elong., %	Red. of Area, %	Y.S. T.S.
58-185-2C-Tap 1	A	226,800	291,500	4.25	7.3	0.78
58-185-2D-Tap 1		225,000	286,000	5.5	12.3	0.79
58-185-2G-Tap 1	B	238,800	291,000	4.5	11.1	0.82
58-185-2H-Tap 1		232,200	282,500	4.5	11.9	0.82
58-185-5C-Tap 2	A	230,700	294,500	3.0	6.2	0.78
58-185-5D-Tap 2		234,900	295,000	3.0	5.8	0.80
58-185-5G-Tap 2	B	252,900	306,000	3.0	5.4	0.82
58-185-5H-Tap 2		248,100	308,000	2.5	5.8	0.81
58-186-2C-Tap 1	A	228,000	288,500	4.0	9.6	0.79
58-186-2D-Tap 1		228,900	289,500	4.25	9.2	0.79
58-186-2G Tap 1	B	240,000	292,000	4.0	8.5	0.82
58-186-2H-Tap 1		234,000	290,000	3.0	6.6	0.81
58-186-5C-Tap 2	A	242,700	302,500	4.0	7.0	0.80
58-186-5D-Tap 2		240,900	304,000	3.75	7.0	0.79
58-186-5G-Tap 2	B	249,900	303,000	2.5	5.8	0.83
58-186-5H-Tap 2		249,900	209,000	3.75	7.4	0.81

Heat Treatment

A — 1850 F — 3 hr — air cool
1650 F — 2 hr — oil quench
400 F — 4 hr — air cool
400 F — 4 hr — air cool

B — 1850 F — 3 hr — air cool
1650 F — 2 hr — oil quench
−105 F — ½ hr
400 F — 4 hr — air cool
400 F — 4 hr — air cool

Analysis

	Composition, %								
	C	Mn	P	S	Si	Ni	Cr	Mo	Al
58-185 tap 1	0.41	0.79	0.010	0.010	0.41	0.96	1.15	0.40	0.044
tap 2	0.42	0.78	0.009	0.010	1.90	0.96	1.11	0.40	0.046
58-186 tap 1	0.42	0.82	0.007	0.008	0.49	0.58	0.89	0.41	0.035
tap 2	0.41	0.82	0.008	0.009	1.98	0.58	0.86	0.41	0.042

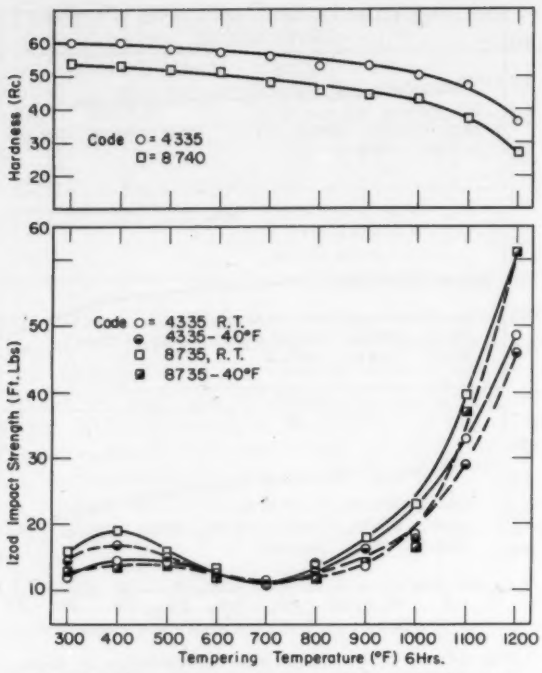


Fig. 17 — Tempering temperature effect on Izod impact strength and Rockwell C hardness of high purity modified 4335 and 8740. Heat treatment: 1850 F-3 hr-air cool; 1575 F to 1600 F-2 hr-oil quench. Temper as indicated.

specimens from these heats tempered at 950 F had almost identical tensile strengths.

There is a marked maximum in the impact strength at 400 F and a minimum at 700 F. The impact strength does not exceed the value obtained at 400 F until a tempering temperature of 950 F is reached. From this point on it increases rapidly.

The 8735 has slightly greater impact strength than the 4335, but this may be related to the difference in hardness.

DISCUSSION

Many investigations have shown that the best combination of tensile and impact properties is obtained when steel is heat treated to give a tempered martensite structure. The main function of alloying elements in steel is to permit obtaining this desired structure in heavier sections; the heavier the section the more alloy that is needed. However, this addition of alloys has the disadvantage of lowering the M_s and M_f temperatures, the temperatures of the start and finish of the martensite transformation. As a result, highly alloyed steels have a greater tendency to have retained austenite present in the final structure. Since this retained austenite may transform in service to produce untempered martensite, it may contribute to premature failure. Even if untransformed, it has an effect on tensile properties if present in sufficient amounts.

The series of experimental heats involving alloy variations based on 4340 indicated that overalloying was detrimental, particularly at the highest strength level. Although tensile strength and ductility were relatively unaffected, there was a definite effect on yield strength and possibly some effect on impact strength.

X-ray diffraction studies showed that 4340 had more retained austenite than the lower alloyed 8740

TABLE 14 — CHEMICAL ANALYSES OF EXPERIMENTAL HEATS

Heat No.	Tap No.	Composition, % C	Mn	P	S	Si	Ni	Cr	Mo	V	Al
57-287		0.30	0.97	0.011	0.027	0.61	2.00	1.02	0.38	—	0.015
57-293		0.25	0.79	0.005	0.008	0.40	1.89	0.89	0.41	—	0.050
57-296	1	0.28	0.86	0.007	0.028	0.40	1.87	0.88	0.40	—	0.022
57-296	2	0.25	0.82	0.028	0.030	0.38	1.90	0.88	0.39	—	0.023
57-306		0.36	0.96	0.013	0.027	0.42	1.85	0.94	0.41	—	0.034
57-310		0.41	0.90	0.013	0.027	0.37	1.89	0.91	0.41	—	0.058
57-317		0.45	0.95	0.012	0.026	0.52	1.89	0.97	0.41	—	0.064
57-355	2	0.25	0.99	0.022	0.016	0.61	1.87	0.86	0.38	—	0.019
57-373	1	0.31	0.94	0.024	0.025	0.51	1.91	0.92	0.40	—	0.036
57-386	1	0.35	0.96	0.015	0.025	0.54	1.92	0.91	0.41	—	0.042
57-386	2	0.35	0.95	0.012	0.024	0.53	1.91	0.92	0.41	—	0.049
57-393	1	0.35	0.95	0.013	0.024	0.50	1.89	0.90	0.41	0.12	0.030
57-393	2	0.35	0.96	0.013	0.028	0.54	1.89	0.92	0.41	0.35	0.058
58-033	1	0.40	1.06	0.006	0.007	0.57	1.88	0.89	0.40	—	0.029
58-033	2	0.39	1.06	0.020	0.028	0.58	1.89	0.89	0.40	—	0.035
58-049	1	0.37	1.01	0.015	0.007	0.49	1.93	0.97	0.41	—	0.024
58-049	2	0.35	1.03	0.025	0.030	0.52	1.93	0.95	0.39	—	0.035
58-064	1	0.34	1.03	0.014	0.025	0.47	1.93	0.92	0.40	—	0.020
58-064	2	0.32	1.01	0.018	0.025	1.56	1.92	0.91	0.40	—	0.003
58-067	1	0.34	1.07	0.018	0.022	2.29	1.97	0.96	0.40	—	0.020
58-067	2	0.34	1.06	0.015	0.022	3.06	1.95	0.93	0.38	—	0.020
58-069	1	0.34	1.01	0.016	0.029	0.47	0.01	0.92	0.40	—	0.029
58-069	2	0.33	1.00	0.028	0.027	0.45	1.29	0.92	0.45	—	0.026
58-071	1	0.38	0.99	0.020	0.024	0.47	0.65	0.93	0.40	—	0.029
58-071	2	0.36	0.97	0.022	0.028	0.47	2.46	0.91	0.40	—	0.029
58-077	1	0.34	1.00	0.014	0.024	0.48	0.67	0.56	0.40	—	0.020
58-077	2	0.34	0.99	0.015	0.026	0.42	0.67	1.56	0.40	—	0.013
58-080	1	0.35	1.51	0.016	0.023	0.51	—	0.86	0.40	—	0.035
58-080	2	0.34	1.53	0.019	0.023	0.47	—	1.51	0.40	—	0.062
58-082	1	0.33	0.99	0.016	0.026	0.44	1.91	0.93	0.28	—	0.028
58-082	2	0.33	1.00	0.016	0.025	0.45	1.93	0.93	0.65	—	0.033
58-084	1	0.30	0.99	0.028	0.027	0.47	—	0.92	0.44	—	0.024
58-084	2	0.30	0.99	0.026	0.027	0.48	—	0.93	0.44	—	0.029
58-089		0.33	1.15	0.009	0.007	0.61	1.93	0.94	0.39	—	0.031
58-093	1	0.32	0.98	0.015	0.028	0.46	—	0.91	0.41	—	0.031
58-093	2	0.32	0.98	0.024	0.028	0.43	—	0.91	0.41	—	0.031
58-111		0.45	1.08	0.004	0.008	0.46	1.91	0.88	0.40	—	0.018
58-Arc-114		0.34	0.93	0.013	0.010	0.28	1.84	0.92	0.40	—	0.011
58-126		0.43	0.76	0.015	0.014	0.26	1.86	0.92	0.42	—	0.011
58-167	1	0.40	1.18	0.010	0.006	1.36	1.93	1.16	0.40	—	0.021
58-167	2	0.40	1.13	0.012	0.007	2.38	1.94	1.09	0.40	—	0.019

Heat No.	Tap No.	Composition, % C	Mn	P	S	Si	Ni	Cr	Mo	V	Al
58-Arc-181											
	1	0.40	1.08	0.008	0.017	0.64	1.91	1.03	0.43	—	0.038
58-Arc-182											
	1	0.37	0.96	0.007	0.012	0.60	1.86	1.00	0.40	—	0.045
58-185	1	0.41	0.79	0.010	0.010	0.41	0.96	1.15	0.40	—	0.044
58-185	2	0.42	0.78	0.009	0.010	1.90	0.96	1.11	0.40	—	0.044
58-186	1	0.42	0.82	0.007	0.008	0.49	0.58	0.89	0.41	—	0.035
58-186	2	0.40	0.82	0.008	0.009	1.98	0.58	0.86	0.41	—	0.042
58-187	1	0.44	0.80	0.010	0.010	0.36	0.56	0.91	0.40	—	0.040
58-187	2	0.42	0.81	0.009	0.010	1.82	0.56	0.88	0.40	—	0.042
58-188	1	0.36	0.83	0.007	0.010	0.44	0.57	0.91	0.40	—	0.046
58-188	2	0.36	0.84	0.008	0.009	2.03	0.57	0.86	0.40	—	0.055
58-190		0.41	0.76	0.006	0.010	2.05	1.92	0.91	0.41	—	0.038
58-193		0.36	0.80	0.012	0.013	0.34	0.56	0.90	0.41	—	0.044
58-194		0.40	0.85	0.009	0.013	0.39	0.55	0.91	0.41	—	0.044
58-196		0.35	0.70	0.013	0.013	0.28	1.88	0.89	0.40	—	0.029
58-207		0.42	0.85	0.008	0.013	1.19	0.57	0.87	0.45	—	0.048
58-208		0.36	0.92	0.008	0.013	1.21	0.58	0.92	0.41	—	0.040
58-Arc-210		0.38	1.02	0.014	0.008	0.37	0.54	0.99	0.39	—	0.040
58-291		0.35	0.75	0.013	0.028	0.49	0.67	0.95	0.40	—	0.051
58-294		0.38	0.60	0.010	0.011	0.32	0.66	0.87	0.35	—	0.049
58-298		0.41	0.68	0.008	0.015	0.36	0.56	0.90	0.38	—	0.020
58-318		0.41	0.92	0.009	0.014	0.45	0.55	0.85	0.41	—	0.038
58-326		0.17	0.53	0.010	0.015	0.25	1.86	0.86	0.40	—	0.031
58-327		0.39	0.68	0.009	0.013	0.36	0.59	0.85	0.40	—	0.031
58-332		0.43	0.81	0.019	0.016	0.40	0.60	0.97	0.39	—	0.035
58-333		0.42	0.77	0.015	0.011	0.38	0.60	0.89	0.40	—	0.046
58-339		0.40	0.67	0.010	0.013	0.30	0.59	0.90	0.40	—	0.050
58-349		0.36	0.59	0.010	0.013	0.30	0.59	0.82	0.41	—	0.072
58-351		0.20	0.64	0.009	0.015	0.27	0.59	0.73	0.41	—	0.046
58-358		0.37	1.01	0.008	0.008	0.65	0.57	0.89	0.39	—	0.050
58-370		0.38	0.76	0.010	0.013	0.40	0.55	0.95	0.40	—	0.031
58-374		0.45	0.86	0.005	0.007	0.47	0.56	0.87	0.40	—	0.033
58-376		0.35	0.72	0.007	0.013	0.42	0.53	0.85	0.41	—	0.049
58-379		0.40	0.81	0.008	0.006	0.47	0.53	0.90	0.40	—	0.065
58-381		0.35	0.68	0.011	0.013	0.42	0.52	0.88	0.41	—	0.013
58-383		0.39	0.74	0.008	0.013	0.45	0.53	0.87	0.39	—	0.033
58-395		0.43	0.88	0.010	0.013	0.55	0.53	0.87	0.40	—	0.029
58-396		0.40	0.76	0.004	0.012	0.42	0.53	0.88	0.41	—	0.040
Utica Metals 2-882 (Vacuum Melt)		0.34	0.97	0.005	0.006	0.47	1.90	0.87	0.39		
Utica Metals 2-964 (Vacuum Melt)		0.32	0.85	0.005		0.53	2.05	0.94	0.48		

after final heat treatment. It is probable that the observed effect on yield strength is related to this difference in retained austenite.

The rather large amounts of retained austenite found in 4340 after tempering were rather surprising at first. However, because of normal segregation in castings the interdendritic areas will be enriched in carbon and alloying elements. Although carbon will diffuse rather rapidly during heat treatment, the alloying elements will not. Thus, there will be islands of highly alloyed material with a lower M_f temperature than the nominal analysis for 4340. These areas probably contain most of the retained austenite.

Of course, when larger sections are involved, the problem of adequate hardenability necessitates additional alloying elements. However, the hardenability of 8740 is adequate for many of the smaller aircraft castings, and nothing is gained by adding more alloy than is actually necessary to produce a tempered martensite structure throughout the section.

Ductility Increase

The comparison of high and normal purity heats indicate a definite advantage is gained in ductility and impact strength by using high purity charge materials. The high purity heats had lower sulfur and phosphorous contents which is probably one reason for this gain in ductility. However, the data from the high purity heats to which sulfur and phosphorous were added is somewhat conflicting. At the high strength level there was little difference in tensile properties.

Heats made with either electrolytic iron or the special high purity sponge iron briquets are probably also lower in other impurity or tramp elements although these have not been identified as yet. However, this investigation is underway.

At first, the cost of the high purity charges may seem prohibitive. The electrolytic iron costs about 30c/lb and the sponge iron briquets about 15c/lb. However, it must be kept in mind that we are considering castings for aircraft and missiles where the utmost in quality is required. Although not negligible, the cost of the metal mix is not the largest item of expense. Also, it should be noted that air melting with high purity charges is cheaper than vacuum melting of the same materials, and the latter is considered feasible for many applications.

Although not a part of this investigation, it has been observed in the pilot foundry on commercial orders that high purity decreases discount from tears and cracks. This in itself probably pays for the material.

Homogenization Effect

It was shown that high temperature homogenization at 2280 F has a decided beneficial effect on ductility at strength levels in the vicinity of 250,000 psi or higher. This treatment must be carried out in a controlled atmosphere and is quite expensive. In addition many foundries or their customers would not have the proper equipment. However, with high purity charges the same level of ductility is obtained without the 2280 F heat treatment.

Other heat treatment experiments showed that lower austenitizing temperatures and refrigeration after quenching were beneficial, particularly with regard to yield strength. These factors would also be expected to reduce retained austenite.

Izod impact strength does not increase continuously with increasing tempering temperatures. A maximum is reached at 400 F followed by a decrease to a minimum at 700 F. Above this minimum the impact strength increases with increasing temperature, and at about 900 F it again reaches the value obtained at 400 F. This behavior results from the microstructural changes occurring in martensite as it is tempered at the different temperatures. The minimum corresponds to the first appearance of cementite with precipitates in grain boundaries.

Selection of the optimum carbon content for a particular tensile strength specification must be made with this behavior in mind. For example, to meet a 180,000 psi minimum tensile strength, an alloy steel containing 0.35 per cent carbon must be tempered at about 1000 F. This is safely above the minimum impact value and also above that obtained at 400 F. In contrast a 0.25 per cent carbon steel would be tempered at 850 F for the same tensile strength. The impact strength, however, would be less than that of the 0.35 per cent carbon steel tempered at the higher temperature.

CONCLUSIONS

1) High purity charge materials improve ductility at high strength levels.
2) Homogenization at 2280 F also improves ductility at the ultra high strength level but not at a tensile strength of 180,000 psi.
3) Alloying elements should be added only to the extent needed to obtain adequate hardenability. Excess alloy has a tendency to promote retention of austenite.
4) Izod impact strength varies with tempering temperature. It has a maximum at 400 F and minimum at 700 F. For a given tensile strength specification, carbon content should be adjusted to avoid tempering between 500 F and 850 F.

ACKNOWLEDGMENT

The authors wish to express thanks to Mr. Ronald C. Campbell for supervision of the x-ray diffraction studies.

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Light Metals Division

Presiding — W. E. SICHA, Aluminum Co. of America, Cleveland. S. LIPSON, Frankford Arsenal, Philadelphia.

Secretary — G. P. MESSENGER, Ordnance Tank-Automotive Command, Centerline, Mich.

Starch Content Effect on a Rammed Graphitic Mold Material for Casting Titanium, H. W. ANTES and R. E. ADELMAN, Frankford Arsenal, Philadelphia.

J. G. KURA:[1] What about reclamation process to reduce material cost?

R. E. EDELMAN: The used molding material is crushed, mainly to reclaim the graphitic material, which reduces the material cost to 15 cents/lb where the original cost is 20 cents/lb.

J. G. KURA: Since prompt use of molds is undesirable, could molds be sealed in plastic bags with a drying agent?

R. E. EDELMAN: This could work, however, it was never tried.

Some Principles for Producing Sound Al-7Mg Alloy Castings, W. H. JOHNSON and J. G. KURA, Battelle Memorial Institute, Columbus, Ohio.

M. C. FLEMINGS and H. F. TAYLOR[3] *(written discussion)*: The authors have conducted an extensive research program in an important area of foundry process development. As they correctly point out, little work has heretofore been directed at determining practical rules for risering non-ferrous castings. Yet such rules, when they are developed, certainly will prove to be important and useful to aluminum foundries engaged in high quality work. The lesson to be learned from the steel foundry industry is a worthwhile one; here, results of risering researches conducted by the Steel Founders' Society (and later by the U.S. Naval Research Laboratory) are used daily in shops throughout the country. Practical use of the research data unquestionably results in increased casting quality and yield, and a decrease in the number of castings that must be scrapped.

However, we are of the opinion that in developing rules for risering non-ferrous castings, care must be taken not to attempt to extend steel thinking too far. The mode of solidification and mechanism of feeding of most aluminum and magnesium alloys are vastly different from those of steel; rules for risering must be based on a different set of fundamental variables, and these rules may ultimately prove to be of quite a different character than those developed for steel.

RESEARCH PROGRAM

In the past eight years there has been an extensive research program in progress in the M.I.T. foundry laboratory supported by Army Ordnance through Frankford Arsenal. This program has been concerned primarily with the fundamental variables involved in solidification of aluminum alloy castings. All publications of the work to date are referenced in a recent summary paper.[1] Most are readily available in the technical literature. Much of the early work was concerned with the effect of chills on microporosity, structure and properties; later, rules were developed regarding spacing of chills and risers to achieve optimum soundness in castings (of several different aluminum alloys). It seems to us this fundamental and practical information should provide a sound basis for further risering studies of aluminum alloys; these references should be examined by the authors.

Based upon studies at M.I.T. we would like to offer specific comments on this paper. The authors relate soundness in their plate castings to thermal gradients at the "solidus" (1030 F). However, diffusion (in the solid) during solidification of aluminum alloy castings is usually quite limited; as a result, these alloys show marked coring and are never completely solid at the equilibrium solidus. Application of a nonequilibrium solidification equation[1] indicates that at 1030 F, Al-7 per cent Mg alloy castings may contain as much as 15-16 per cent liquid metal. It is therefore questionable whether thermal gradients at this temperature (as determined by the authors) have much real significance in determining riser behavior.

AVAILABLE DATA

The authors state that "information on the effect of conventional chills on the thermal gradients that are present in non-ferrous castings is . . . scarce." Actually, there is an abundance of data available on the effect on thermal gradients (and mechanical properties) of plain chills, locking chills, water cooled chills and tapered chills;[2,3] also, the effect of chills in combination with exothermic[2] and insulating[2,4] materials has been reported. The authors imply that steel chills are standard for aluminum foundry use, whereas we feel if any material could be considered standard it is gray iron. However, more and more foundries are beginning to recognize the advantages of aluminum chills (both for technical and economic reasons).

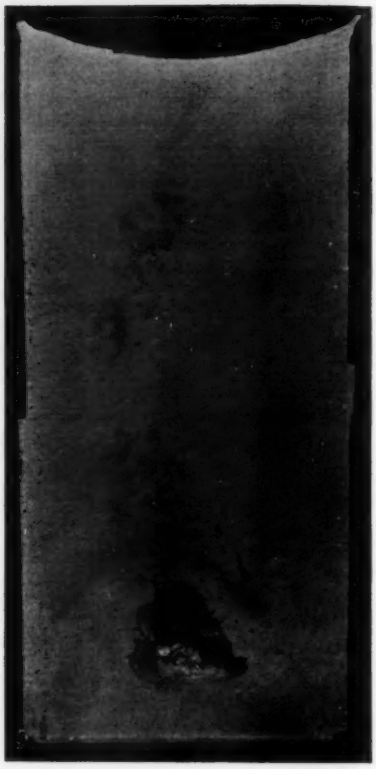


Fig. 1 — Section of top risered cube of aluminum — 20 per cent copper alloy. 5 in. cube has riser 5 in. diameter x 6 in. high. Riser was insulated with a top covering of hot gypsum.

The authors agree with earlier works by the undersigned that thermal gradients are a prime variable affecting soundness (and riser requirements) in aluminum alloy castings. However, other factors are also important and must be considered in developing practical risering parameters. Among these factors is macrosegregation (bulk movement of either liquid or solid during solidification). Figure 1 is an extreme case illustrating how movement of solid affects soundness of an aluminum alloy casting.

The casting shown is of aluminum—20 per cent copper alloy, grain refined; it is a top risered cube, sectioned down the center for examination of the shrinkage. Here the shrinkage pipe is at the bottom of the casting because the low-copper solid grains that formed during the early part of solidification were lighter than the high copper liquid bath; as a result they floated to the top of the casting. The bottom of the casting, rich in low melting eutectic, then froze after the riser. While Fig. 1 represents an unusually severe case, movement of solid (either by floating or settling) appears to be an important factor in determining riser adequacy of many non-ferrous alloys.

FEEDING DISTANCE

The concept of feeding distance in non-ferrous alloys is one that we feel must be treated with caution. Our work has indicated that when thermal gradients fall below about 100 F/in. fine microporosity begins to appear in aluminum castings (of alloys such as 195 or 356); this is shown in Fig. 2.

Since thermal gradients as steep as the above are obtained in sand castings only by severe chilling, unchilled sand castings of such alloys always possess microporosity, regardless of riser size or location. Our experience has been that as thermal gradients are reduced from the high gradients necessary to produce sound metal microporosity gradually increases. There is no distinct thermal gradient (such as the 5 F reported in this investigation) that makes the difference between sound and unsound metal. Effective feeding distance of a riser depends on the amount of porosity that can be tolerated and on the sensitivity of the method used to detect this porosity.

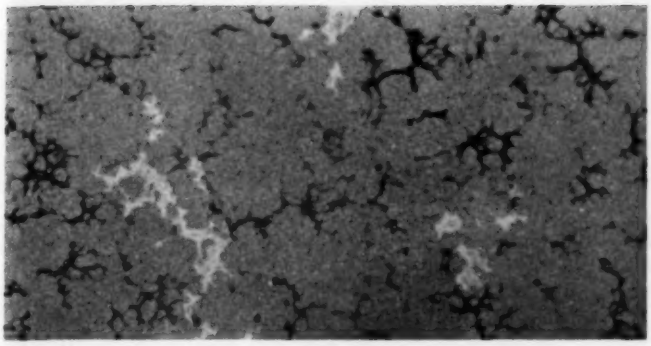


Fig. 2 — Microradiograph of a section from an end-chilled plate casting of 195 alloy. Dark areas are aluminum-copper eutectic; light areas are micro-porosity. Section was taken approximately 5.4 in. from the chill. Microporosity is less severe nearer the chill.[7]

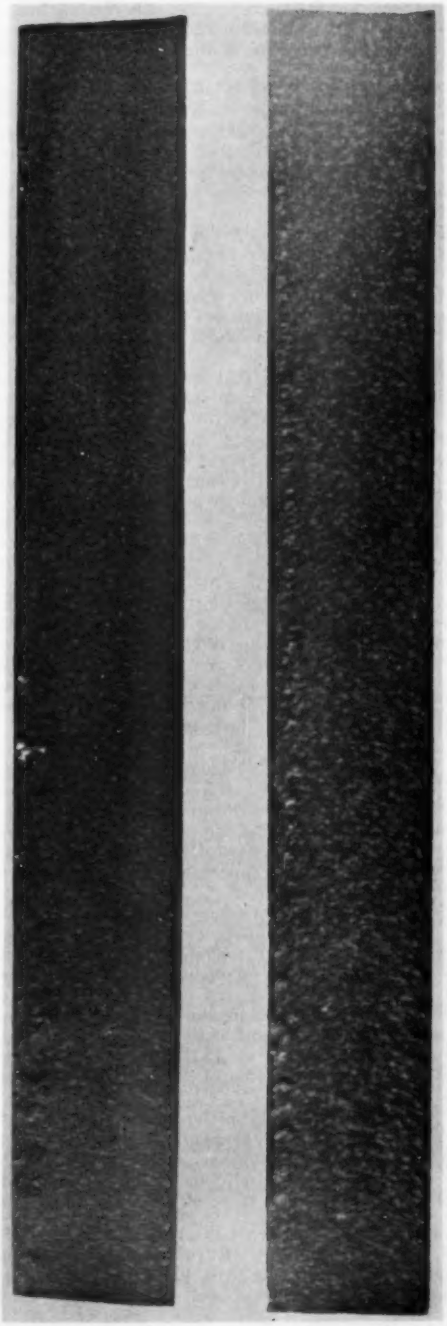


Fig. 3 (Left) — X-rays of sections cut from plate castings of 195 alloy showing how increased riser size reduces shrinkage porosity.
Left: x-ray of a ⅛-in. section from a plate casting end risered with a riser 5 in. diameter x 5-in. high.
Right: Same as left except riser is 2½ in. diameter by 5 in. high.
Both plates were 5 x 8 x 1 in. thick, cast horizontally (see Fig. 3 of paper to which this discussion refers). Riser was at bottom. X-ray sections were cut vertically in the center of the plate and are 8 x 1 x ⅛-in. thick. Light areas are porosity.

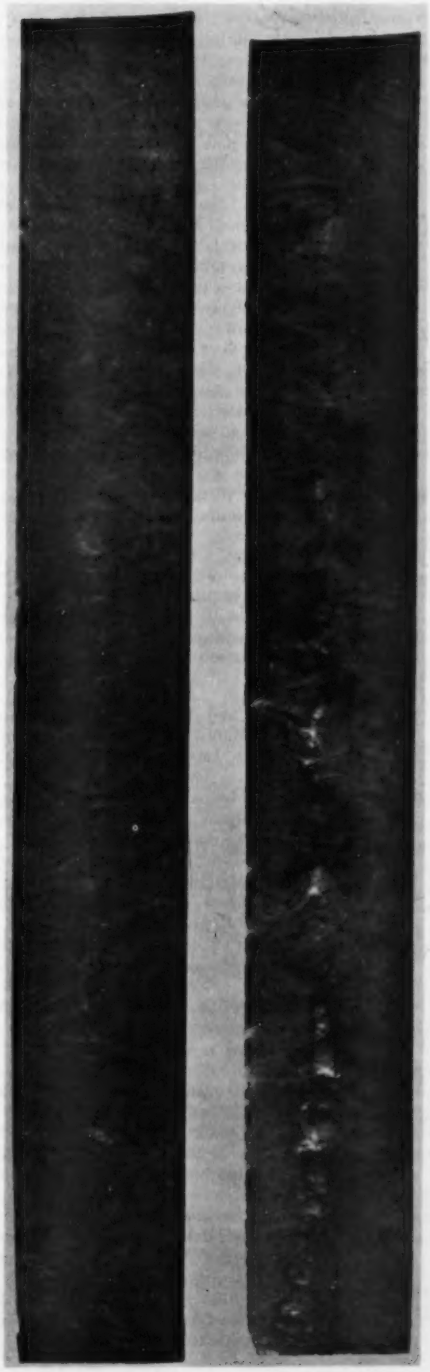


Fig. 4 (Right) — X-rays of sections cut from plate castings of Al — 7 per cent Mg alloy, showing how increased riser size reduces shrinkage porosity.
Left: x-ray of a ⅛-in. section from a plate casting end risered with a riser 5 in. diameter x 5 in. high.
Right: Same as left except riser is 2½ in. diameter x 5 in. high.
Both plates were 5 x 8 x 1 in. thick, cast horizontally (see Fig. 3 of paper to which this discussion refers). Riser was at bottom. X-ray sections were cut vertically in the center of the plate and are 8 x 1 x ⅛-in. thick. Light areas are porosity.

While steep thermal gradients are necessary to make 195 alloy castings perfectly sound, (as measured by careful laboratory techniques) much lower gradients are adequate to make these castings sound to the 2 per cent definition ordinarily achieved on x-ray film. Properly designed risers (in the absence of chills) can produce gradients required to make extensive plate sections appear sound by ordinary x-ray methods.

RISER SIZE NEEDED

However, because of the extreme mushy nature of solidification of the alloy, and because of the high heat diffusivity of aluminum, such risers must be 1) large (by steel casting standards), and 2) thermally insulated from the casting. The latter requirement means that the best risering practice is to attach a large riser to a casting with a smaller riser pad. Top risers should never be used on aluminum castings unless they are necked down at the juncture between casting and riser. When top risers are not necked down, under riser shrinkage almost always occurs, regardless of riser size. This is shown by the authors in their top risered plate casting.

The effect of riser size on soundness of an end risered plate casting (of 195 alloy) is shown in Fig. 3. The left photograph in Fig. 3 is an x-ray of a ⅛-in. section taken from the center of a plate casting; the plate was similar to one employed by the authors' (5 x 8 x 1 in. thick) risered on the end, as was the authors, by a 2½ in. diameter x 5 in. high riser. Microporosity is apparent throughout. For comparison, the right photograph in Fig. 3 is an identical x-ray from a similar plate, except that in this case a much larger riser was used (5 in. diameter x 5 in. high). Increasing the riser size improves feeding; the plate is nearly free of porosity to the limit of x-ray definition.

Figure 4 shows similar x-rays for plates cast of the alloy studied by the authors (Al - 7 per cent Mg). Here, severe shrinkage resulted when the 2½ in. diameter riser was used; this agrees with results reported by the authors. However, doubling the riser diameter essentially eliminates the shrinkage. Thus, it appears the quantitative results of the authors would have been quite different had they investigated (in greater detail than they did) the effect of riser size on feeding. Compared with steel castings effective feeding distances in non-ferrous castings are amazingly long when adequate risering is employed.

CONCLUSION

In conclusion, we would like to re-emphasize our opinions that studies on risering requirements of non-ferrous alloys are of great potential benefit to the non-ferrous foundry industry; but that these studies, to be useful in practice, must be based upon fundamental solidification and heat flow variables.

REFERENCES

1. Flemings, M. C. and Taylor, H. F., "Adapting Theory to Practice in the Manufacture of Light Alloy Castings," Exchange Paper, 25th International Foundry Congress, Brussels, 1958.
2. Foundry Section, Metals Processing Div., Dept. of Met., M.I.T., "Study of Techniques for Refinement of Cast Structures in Aluminum," Final Report, Dept. of the Army, Ordnance Corps Contract No. DA-19-020-ORD-2004, July 1, 1954.
3. Foundry Section, Metals Processing Div., Dept. of Met., M.I.T., "Study of Techniques for Refinement of Cast Structures in Aluminum," Final Report, Dept. of the Army, Ordnance Corps Contract No. DA-19-020-ORD-2004, July 1, 1955.
4. Foundry Section, Metals Processing Division, Department of Metallurgy, M.I.T., "Parameters for Controlled Solidification in Aluminum Casting," Final Report, Department of the Army Ordnance Corps Contract No. DA-19-020-ORD-3674, July 1, 1957.
5. Taylor, H. F., Walther, W. D. and Adams, C. M., "Techniques for Improving Strength and Ductility of Aluminum Alloy Castings," AFS Transactions, *vol. 62*, 1954.
6. Flemings, M. C., Norton, P. J. and Taylor, H. F., "Performance of Chills on High Strength-High Ductility Sand Mold Castings of Various Section Thicknesses," AFS Transactions, *vol. 65*, 1957.
7. Passmore, E. M., Flemings, M. C. and Taylor, H. F., "Fundamental Studies on Effects of Solution Treatment, Iron Content, and Chilling of Sand Cast Aluminum-Copper Alloy," AFS Transactions, *vol. 66*, 1958.
8. Uram, S. Z., Flemings, M. C. and Taylor, H. F., "Effect of Pressure During Solidification on Microporosity in Aluminum Alloys," AFS Transactions, *vol 66*, 1958.

Authors' Reply: In their research, Professors Flemings and Taylor desired to obtain perfectly sound light alloy castings. Consequently, they employed metallographic and microradiographic techniques to evaluate soundness. Our objective was to determine parameters and from these parameters deduce principles and rules that could be used to control and minimize the development of shrinkage voids or gross porosity in castings of commercial utility, specifically armor castings. Conventional radiography was used as the tool for evaluating the soundness of our experimental castings. It was on this basis that we evolved the principles and rules for producing sound castings of the Al-7Mg alloy.

At this time our conclusions are pertinent only to the Al-7Mg alloy. The mode of solidification can differ markedly among the light metal alloys. This could result in drastic modification of our present specific rules that apply to the Al-7Mg alloy. Drawing from the field of copper-base alloys for an example, we would hazard a guess that the 85-5-5-5 alloy follows the principle found for the Al-7Mg alloy, i.e., the feeding range of a riser is shorter as the section thickness increases. However, the rule for soundness as a function of thermal gradient is quite different for these two alloys.

In the Al-7Mg alloy, we found it was necessary to have a gradient at least as large as 5 F/in. Flinn and Mielke* found it necessary to achieve a thermal gradient in excess of 60 F/in. to obtain soundness in the 85-5-5-5 alloy. Further research will be required to catalog the various non-ferrous alloys with respect to their mode of solidification and then establish the universal principles and specific rules that govern how they should be cast to make them sound.

We related soundness in our experimental Al-7Mg alloy castings to thermal gradients at the equilibrium solidus (1030 F). This technique was similar to the one used by other investigators to determine the risering requirements for steel. Flinn and Mielke* also used the equilibrium solidus as the reference point in measuring thermal gradients in their copper-base alloy. A quick review of our data reveals that none of our conclusions would be changed significantly if we had abritrarily selected a reference point of 1000 F instead of 1030 F for determining the thermal gradients in the solidifying Al-7Mg alloy castings. From a practical viewpoint, it does not appear to make any difference in our conclusions that, at 1030 F, as much as 16 per cent liquid metal still existed during solidification of our castings.

As a starting point in our research, we specifically decided to test the rules used for risering steel to determine to what degree they applied, or to what degree they would have to be modified to be applicable to the Al-7Mg alloy. The need for a new set of principles and rules for rigging the light alloy became apparent from our work. The most outstanding principle found was that the feeding distance of the riser decreases with increase in section thickness. Also, contrary to the rules for steel castings, end chills do not increase the feeding range of the riser in such a plate. None of this information had been published elsewhere prior to our investigation.

No single factor, such as a critical thermal gradient, can be assumed to be the only criterion determining the soundness obtained in casting of any alloy. Of course, macrosegregation to the degree cited by Professors Flemings and Taylor in an unconventional alloy such as Al-20Cu could have an over-riding influence on the resulting soundness of a casting. Fortunately, the conventional alloys do not appear to have segregation to the degree they cited.

Therefore, so long as we are not concerned with microporosity that cannot be revealed by ordinary radiographic techniques, but rather with visible shrinkage voids and macroporosity, the criterion of a critical thermal gradient controlling whether soundness is achieved is acceptable and useful. On the basis of this definition of objective and method of correlation and evaluation, it was possible to consistently differentiate between an unsound and a sound area in Al-7Mg alloy castings and correlate this with the critical value of 5 F/in. for the thermal gradients.

Professors Flemings and Taylor prefer to believe that their work with chills constitutes an abundance of data. All data are valuable, including theirs. However, as pointed out specifically in an earlier paragraph, additional information on the effect chills have on the feeding range of risers was one of the

*Flinn, R. A., and Mielke, C. R., "Pressure Tightness of 85-5-5-5 Bronze," progress report, AFS Brass and Bronze Div. Research Committee, Transactions, vol. 67, 1959, pp. 385 ff.

small new contributions by our work. This information had not been revealed by the M.I.T. programs. The general statement in our paper with regard to the scarcity of information on the thermal gradients in non-ferrous castings is still true, even in spite of our own small contribution to the knowledge in this specific area.

Nor did we intend to imply that steel chills are standard for use in the non-ferrous foundry. The important fact is that, for horizontal plates, end chills of any conventional material are inadequate for increasing the feeding range of the riser.

In any research program, invariably many points are not covered completely, or are not touched upon at all because of lack of time. One such important point that was not touched upon in our work because of lack of time was the study of a necked-down top riser on horizontal plates. Instead, the work at M.I.T. on this type of riser was relied upon to fill the gap temporarily. Therefore, our paper made no mention of it since we had not investigated it. Therefore, we were in no position to draw specific conclusions on the effect of necking on the feeding efficiency of such a riser. Logic would lead one to believe that it should be advantageous in eliminating under-riser shrinkage.

We agree with the discussors' analysis of the improvement in feeding of the 195 alloy plate when the size of the end riser was increased, as shown in their Fig. 3. It is obvious that the 8-in. long plate with the small riser was underfed because it was beyond the feeding range of the riser. When the size of the riser was increased (left radiograph, Fig. 3), the plate was sound for a distance of 5T from the edge of the plate opposite the riser. Shrinkage porosity was still present for a distance of 3T in front of the riser. This is a remarkable confirmation of the work we reported, especially when it is realized that we worked with an Al-7Mg alloy whereas Professors Flemings and Taylor worked with Al-4Cu-2.5Si alloy.

Although their larger riser improved the overall soundness of the plate, the larger riser did not increase the feeding range. It appears that a larger riser was required on the 195 alloy than on the Al-7Mg alloy to reveal clearly that the feeding range is 5T in a 1-in. thick plate. However, at no time did we infer that the risers which provided optimum feeding in Al-7Mg alloy would necessarily also hold for any other alloy.

With regard to the discussors' radiographs of their Al-7Mg alloy plates in their Fig. 4, again precise confirmation was obtained of the work that had been done in our laboratory. The information is presented below now, but had been deemed as unnecessary detail for inclusion in the paper. We had found that an increase from 2½ to 5 in. in the diameter of the riser minimized porosity in a 1-in. thick plate, 8 in. long. The radiograph shown at the left in Fig. 4 matched ours. By our standards of evaluation, both radiographs revealed that the feeding range of the riser is 5T. The 8-in. long plate is beyond the feeding range of either a 5 or 2½-in. diameter riser.

A series of experiments with 8, 5, and 3-in. long plates more clearly defined that the feeding range of a 2½-in. diameter riser is 5T in the 1-in. plate. These experiments were described in the paper. Since it was concluded that a 5-in. diameter riser did not have a greater feeding range than 2½-in. diameter riser, the smaller riser would be preferred for economical reasons.

The discussors stated that, "compared with steel castings, effective distances in non-ferrous castings are amazingly long when adequate risering is employed." It is difficult to support such a broad statement either from the literature or from experience. It may be true for isolated or specific cases. However, since so many of the important non-ferrous alloys have a relatively broad freezing range, freezing progresses through a mushy zone. This simple fact mitigates against risers having a feeding range greater in non-ferrous alloy castings as compared with steel castings.

The authors wish to voice concurrence with the discussors' conclusion. Fundamentals constitute the ultimate basis of our knowledge. Until fundamental research can progress to the stage where the numerous parameters of casting can be brought under strict and precise control, we must resort to empirical relationships based on broad principles. Such relationships aid the foundryman to do a better job and can assist the research into fundamentals. Therefore, more of both types of research should be conducted.

H. E. ELLIOTT:[2] Have you tried a chill with uniform thickness rather than a tapered?

W. H. JOHNSON: I have not examined this type of chills. However, there is a good possibility that this could be used successfully.

Presiding — C. A. KOERNER, Central Foundry Div., General Motors Corp., Saginaw, Mich.
D. L. LAVELLE, American Smelting & Refining Co., South Plainfield, N.J.

Secretary — J. G. KURA, Battelle Memorial Institute, Columbus, Ohio.

Radiography, Microstructure and Mechanical Properties of Cast Magnesium-Thorium-Zirconium Alloy HK 31A, a correlation, T. R. BERGSTROM and R. G. BASSETT, Boeing Airplane Co., Seattle, Wash.

B. LAGOWSKI and W. A. FOLLARD[4] *(written discussion):* The "spherical" inclusions described by the authors appear to be similar to those first observed by Emley* who termed them "ring nebulae" from their characteristic radiographic appearance. Emley believed that they may have resulted from reaction of the molten alloy with entrained sand grains. He also described the eutectic-like constituent shown in Fig. 8 of the present paper.

We have examined similar inclusions in various magnesium-base alloys (both with and without thorium), and have confirmed that they are caused by entrained sand grains which react with the magnesium to form alternate lamellae of magnesium silicide and magnesium oxide in the characteristic eutectic-like structure.

The appearance of the inclusions in radiograph is due to a shell of compounds of zirconium and thorium with silicon which is more opaque to x-rays than the matrix.

If the authors will re-examine their samples they will find that a transparent silica particle is present at the center of most inclusions. These particles appear similar to voids when observed under vertical illumination.

The results of tensile tests reported in the paper are interesting, and show that the inclusions can have a serious effect on the strength of sections in which they occur. It is, therefore, especially important that their nature and source be established so that they can be eliminated by appropriate precautions in the foundry.

H. E. ELLIOTT[2] *(written discussion):* This paper is a valuable contribution to our knowledge of how the several types of radiographically detected segregation affect the mechanical properties of HK31A alloy castings. On the basis of this careful study, designers can proceed with confidence in HK31A castings containing these radiographic conditions, knowing that the effects on properties are moderate rather than extensively damaging.

In selecting an alloy for a given application, certainly the designer should consider the effects of the conditions that were investigated in this paper. However, a point not thoroughly covered in this paper should also be considered; it should be understood that no evidence has been brought forth, nor has it been suggested, that HK31A is more prone to segregation than are other commercial casting alloys such as the conventional magnesium-aluminum-zinc family alloys that have been successfully used for decades.

Since there is greater contrast in the radiographic densities of the constituent elements of HK31A, macrosegregation is more readily detected radiographically; this does not mean that macrosegregation in HK31A alloy is necessarily more severe. Realistically, this characteristic of HK31A should be regarded as an advantage, since it permits more critical inspection of this material than is possible by nondestructive methods with such alloys as the magnesium-aluminum-zinc alloys.

NO DESIGN HAZARD

Thus, designers should certainly not infer that it is hazardous to design with HK31A alloy because of segregation problems; it would be a mistake to specify other materials such as the magnesium-aluminum-zinc family alloys in the attempt to avoid such problems. It is generally found, and this fact is reflected in specification property minima, that a higher percentage of separately cast bar properties are attained in castings of HK31A alloy than is the case in the Mg-Al-Zn family alloys.

Another point worthy of note is the fact that HK31A alloy is not unique among the thorium-containing magnesium-base alloys in its susceptibility to these segregation conditions. At

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*E. F. Emley, "Non-Metallic Inclusions in Magnesium-Base Alloys Containing Zirconium," *J. Inst. Metals, 75,* pp. 481-512, 1949.

the author's company's foundry, we have had numerous opportunities to cast the same part alternately both in HK31A and in its sister alloy HZ32A. We have consistently encountered the same segregation conditions in both alloys, and to about the same degree. Others* have reported similar segregation effects in the magnesium-rare earth metals-zinc-zirconium alloys.

SUMMARY

To summarize my remarks, HK31A alloy shares with many other cast metals of engineering importance a tendency toward macrosegregation effects. These conditions are more easily detected radiographically in HK31A alloy than they are in many other casting alloys. However, HK31A castings containing these conditions still retain highly interesting engineering properties. It would be regrettable if these conditions created such "perturbation to personnel in the x-ray laboratory" that other materials that require destructive examination to disclose similar conditions would be specified in preference to HK31A.

We are indebted to the authors for this careful study that will enable design engineers to assess quantitatively the effects of the segregation conditions that were investigated.

AUTHORS' CLOSURE: Messrs. Lagowski and Pollard's suggested cause of the spherical type inclusions was not considered by the authors. However, sand grain inclusions are a possibility and should be investigated. Reexamination of our specimens under oblique illumination did not reveal any sand grains. A fair percentage of the spherical inclusions do not have a void center, and if the proposed reaction is the cause of these inclusions we must conclude that the reaction goes to completion.

Data could not be found to determine the rate of this reaction, therefore, the feasibility of its proceeding to completion in the limited time that the casting is molten could not be determined. We have started a microautoradiographic investigation to determine the distribution of thorium in the particles. This investigation should resolve the question of whether the lamellar eutectic-type structure contains Mg_5Th as one of its phases.

Mr. Elliott's comments concerning macrosegregation in magnesium alloys are certainly true. The writers have not made an attempt to evaluate segregation in magnesium as such, but merely to determine effects on structural capabilities of alloys being used. Other alloys have been investigated when the need arose.

The purpose of an investigation of the type reported is to determine the effect of anomalies that are present and to decide upon limits in amount, type and distribution of the anomalies that will insure structural integrity. After an investigation of this nature the confidence we have in any cast part is greater than it would be for a part where only normal sampling and tensile testing are performed. The selection of materials and quality level is the responsibility of the Engrg. Dept., x-ray standards are then issued to prevent unjust rejection of parts with seemingly peculiar x-ray behavior.

Presiding — H. E. ELLIOTT, The Dow Chemical Co., Bay City, Mich.
K. B. BLY, Fabricast Div., General Motors Corp., Bedford, Ind.

Secretary — R. W. BRUNER, JR., American Smelting & Refining Co., Alton, Ill.

Hot Cracking Test for Light Metal Casting Alloys, E. J. GAMBER, Aluminum Co. of America, Cleveland.

MEMBER: Concerning Table 4 of your paper. Should the first alloy listed under "U Casting Tests" be designated EK41A instead of AK41A?

E. J. GAMBER: Yes, the alloy should have been designated EK41A instead of AK41A. Also, I would like to correct two other designations appearing in Table 4. Under "Modified Ring Tests," alloys should be designated HK31A instead of HK31XA and HZ32A instead of HZ33A.

Sample for Rapid Measurement of Gas in Aluminum, H. V. SULINSKI and S. LIPSON.

S. URAM,[5] M. C. FLEMINGS, H. F. TAYLOR[3] *(written discussion):* The authors have demonstrated a thorough understanding of the factors which are important in the measurement of gases in aluminum alloys. Such precise measurements are generally reserved for a laboratory where analyses can be made under careful control. That the authors have been able to make measurements on the foundry floor with excellent reproducibility is proof enough of their accomplishment. They have added an important quality control tool to those available to the foundry engineer.

The occurrence of porosity and its relationship to dissolved gases has been illustrated in many papers, and it is indeed shown dramatically by Fig. 9 of this paper. The detrimental effect of large amounts of microporosity on tensile strength and ductility of cast metals is well known. The effect of smaller amounts of microporosity is perhaps not so well understood, but even small amounts of porosity can be deleterious to such mechanical properties as fatigue strength, impact strength and ductility. A major reason why we do not understand better the effects of small amounts of gas is the difficulty of measuring gas contents in practice.

We believe that the authors' test methods have made it possible to make accurate comparisons between gas contents and mechanical properties of many alloys. This method represents an important improvement over the reduced pressure test commonly used in foundries. It should be adopted as a quality control tool in foundries endeavoring to make high quality aluminum castings.

Presiding — F. C. BENNETT, The Dow Chemical Co., Midland, Mich.
D. S. MILLS, Process Development Staff, General Motors Corp., Warren, Mich.

Secretary — J. G. MEZOFF, The Dow Chemical Co., Midland, Mich.

How Aircraft Designers look at Light Metal Castings, A. R. MEAD, Grumman Aircraft Corp., Bethpage, Long Island, N. Y.

W. E. SIGHA:[6] Mr. Mead's paper has presented information on factors that are important to aircraft designers in using light metal alloy castings.

Comments of this type are instructive and helpful to casting producers. Unfortunately, there are two items in the paper which could leave readers with erroneous impressions.

It is indicated that small additions of Mn and Cr have been used in 220 alloy to improve the resistance to stress corrosion. There is evidently some misunderstanding involved because neither of these elements, singly or in combination, has any significant influence on stress-corrosion characteristics, at least as far as 220 alloy is concerned.

The statement is made also that 220 alloy castings should be quenched rapidly after solution heat treatment to provide high resistance to stress-corrosion cracking. Actually, this view is diametrically in contrast to the facts established by extensive laboratory testing and almost 30 years of commercial experience by the developers of 220 alloy.

Two patented quenching practices developed to improve the resistance to stress corrosion of alloy 220 entail quenching in oil at 250 to 300 F or an interrupted quench in boiling water. Both are relatively slow quenching treatments.

Specimens of 220-T4 alloy exhibit excellent resistance to stress corrosion regardless of the quenching rate when exposed to an accelerated test soon after quenching. However, after aging at room temperature for six months or longer, the susceptibility to stress-corrosion cracking increases with the severity of the quench.

Alloy X250-T4 is a new development designed to replace alloy 220-T4. It provides practically the same tensile properties as 220-T4, has the same excellent resistance to general corrosion and has superior resistance to stress-corrosion cracking and room temperature aging despite the fact that it receives the more drastic, yet simpler, full boiling water quench.

AUTHOR'S CLOSURE: It was well known that alloys of the 220 type are susceptible to stress corrosion cracking and this point was emphasized by many authors. Messrs. Dix and Brown have stated on page 230 of the 1948 A.S.M. *Metals Handbook:*

> Binary alloys containing more than about 4½% Mg. are susceptible to stress corrosion cracking particularly when they are cold worked. By the addition of inhibitors such as manganese and chromium satisfactory freedom from stress corrosion cracking can be obtained in aluminum-magnesium wrought products containing as much as 5½% magnesium. Castings that contain about 10% Mg., although susceptible to stress corrosion cracking,

*Skelly, H. M. and Sunnucks, D. C., "Segregation in Magnesium-Rare Earths-Zinc-Zirconium Alloys," AFS TRANSACTIONS, *vol. 62,* 481-491, 1954.

have been used successfully with no reported failure from this cause.

Dix and Brown also state on page 792 of the A.S.M. Handbook:

> It will be observed that manganese and aluminum form a compound that has almost the same electrode potential as aluminum. This is true also of the constituent formed when magnesium and silicon are added in the ratio of the compound Mg_2Si. These alloying elements produce alloys that are comparable with commercially pure aluminum in their resistance to corrosion. Chromium and manganese behave similarly with respect to corrosion, but chromium is employed only as a minor alloying element, particularly valuable in improving the resistance to stress corrosion of several types of aluminum alloys.

After being convinced that 220 was known to be prone to stress corrosion cracking and after reading an article by Dix in *Corrosion of Metals*, Feb. 1946 which states:

> Commercially pure aluminum used in wrought form or as a basis for the manufacture of alloys contains an average aluminum content of about 99.3%. The remainder is made up principally of iron, generally less than 0.5% silicon, generally less than 0.25% and copper, generally less than 0.1%.

I felt that Alcoa must add other elements to suppress stress corrosion in 220 alloy. Analysis of actual 220 castings from 19 heats indicated the average chromium content was 0.01 per cent and the average manganese content was 0.03 per cent. Data compiled by Alcoa state that high iron, silicon or manganese decrease mechanical properties. This information is under the heading of Consequences of Exceeding Impurity Limits.

Since the commercially pure aluminum used as a basis for the manufacture of alloys does not have chromium or manganese mentioned as constituents, and 0.10 per cent maximum manganese is specified in 220 alloy and chromium was found in 19 heats, these findings tend to indicate that the addition of chromium and manganese are of a proprietary nature.

Many proprietary additions are made in both the ferrous and non-ferrous field. These additions are usually a small percentage of the total but perform valuable functions. Specifications do not always cover these minute additions and when they are found they might be considered as impurities.

Beryllium, sodium, boron, zirconium, calcium, barium, strontium, titanium, columbium and tantalum are some of the elements added to light metal alloys.

The metallurgy of minute additions is a very interesting subject. Certainly the composition of Silicon Aluminum Casting Alloys, produced by several metal houses, makes no specific mention of a residual sodium percentage which might range from 0.05 per cent to 0.014 per cent, yet this small amount of sodium can shift the eutectic composition from 11.7 per cent to 14 per cent silicon. The eutectic temperature can also be lowered from 577 C to 564 C. The microstructure of the 13 per cent Si Aluminum alloy is altered considerably by the addition of 0.014 per cent sodium. This addition also improves the tensile strength and elongation of the 13 per cent silicon aluminum alloy. For the aluminum magnesium alloys, a very small addition of beryllium can do a great deal of good in preventing oxidation of the surface of the melt.

Boron* also has benefited the heat treatable alloys of 10 per cent magnesium types, by bubbling through the melt boron trichloride for a sufficient time to leave a residual of 0.010 per cent boron, improvements of tensile strength and elongation can be realized.

As for the statement concerning rate of quench for the 220 alloy, my comments were based on tests performed at The National Bureau of Standards and at Grumman. Discussions with W. E. Sicha of Alcoa after presentation of my paper indicate that the length of time between quenching and testing could lead to erroneous results. This is being checked.

Brass and Bronze Division

Presiding — W. H. Baer, Fort Belvoir, Va.
J. E. Gotheridge, Foundry Services, Inc., Columbus, Ohio.

*H. W. Gillett Memorial Lecture, 1953.

Secretary — W. M. Spear, Worthington Corp., Harrison, N. J.

Electrical Conductivity of Sand Cast Copper-Base Alloys, D. G. Schmidt and F. L. Riddell, H. Kramer & Co., Chicago.

M. V. Davis[7] *(written discussion):* The authors are to be congratulated for their excellent paper which is a most welcome addition to the copper-base casting alloy literature. The compilation of electrical conductivity data on these alloys is long overdue as anyone searching the literature will attest to. The authors are to be commended, also, on the wide scope of their investigation.

As regards the selection of the test specimen, it is regrettable the test specimens did not conform to the requirements of A.S.T.M. Designation B193-57, Standard Method of Test for Resistivity of Electrical Conductor Materials. This specification requires the test length to be at least 1 ft or 30 cm (obviously the cast test bars do not permit a test length of 1 ft). On test specimens of sub-standard length, if the weight resistivity rather than the volume resistivity of the test specimen is determined, a higher order of accuracy can be obtained. A simple conversion factor is used to obtain volume conductivity from weight conductivity. In our laboratory, we believe that the measurement of resistivity or conductivity on a weight basis for sand castings is the more reliable method and has a higher order of accuracy than values obtained on a volume basis.

The authors could have made more persuasive arguments for the selection of a 2 in. test or gage length. The results shown in Table 1 of the paper were obtained with specimens obtained from a wrought bar, while the data on other alloys were obtained from sand-cast test bars. Perhaps a better procedure would have been to machine the entire test bar length to 0.505 in. diameter. The resistivity of a 6 in. test length could be compared to a 2 in. test length (essentially the comparison of the resistivities using wrought alloys). If this procedure had been used with several types of alloys, a more reliable indication of the accuracy of the 2 in. test length would have been obtained and, incidentally, a measure of the soundness of the test bars.

It would be interesting to learn of the authors' experience with test bars cast in the form and dimensions, as shown in Fig. 4 of the paper, when casting high-shrinkage alloys such as aluminum bronze (A.S.T.M. Designation B208-54 does not recommend the use of this type test bar with high-shrinkage alloys). Although, at the author's company we have occasionally used this type test bar for convenience, it has been our experience that erratic mechanical property values have resulted and, in some cases, the test bars are unsound.

With the growing acceptance of the eddy current method for rapid measurement of electrical conductivity, it would be of interest if the authors were able to show the relation of their values obtained from Kelvin bridge measurements and those obtained by eddy current measurements on the same test specimens.

The electrical conductivity values obtained in our laboratory for aluminum bronze (80-5-10-5), commercial red brass (83-4-6-7,) and silicon bronze (92-4-4) are in close agreement with the values presented in this paper.

My comments should in no way be construed as detracting from the overall excellence of this paper. All of us concerned in producing copper-base castings having electrical conductivity requirements owe the authors a vote of thanks for a job which, at best, must certainly have been tedious and laborious.

W. B. Scott[8] *(written discussion):* In the past, electrical conductivity figures have been thrown about with abandonment. Messrs. Schmidt and Riddell are to be complimented for their efforts to bring some degree of order into the picture.

I would like to ask for elaboration of the report paragraph as to the selection of test specimens. Is there any explanation as to why the wedge-type test bar has failed to give reproduceable results? Perhaps the normal run of sand castings will not give reproduceable results.

In regard to the manganese and aluminum bronzes, the decrease in conductivity with increasing copper content does not seem consistent with explanations. Is there any explanation? What is the role of:

Grain size?
Phase changes or changes in microstructure?
Density?

Furthermore, do the authors have any report as to the practical effect of size of section and surface finish?

AUTHORS' CLOSURE: Since the majority of conductivity measurements were less than 33 per cent I.A.C.S., the accuracy obtained by volume measurement was suitable for our purposes, when all other sources of error are considered. A threaded test bar does not easily lend itself to a weight resistivity measurement.

The ⅝ Web Webbert test bar (ASTM Designation B208-54) is suitable for high shrinkage alloys, with excellent reproduceability, as noted by Mr. A. J. Smith, in his paper, "Some Foundry Problems In The Development Of A New Marine Propellor Alloy." Occasional erratic results are usually caused by small oxide spots, which the careful observer can readily discern. We have never had any bars that were unsound, due to shrinkage, as the casting is adequately gated and risered. Any unsoundness usually is the result of gassy metal or excessive moisture in the mold.

No comparison was made between the Kelvin bridge and the eddy current measurements for electrical conductivity. The wedge type test bar for conductivity specimens was not satisfactory, because of the difficulty in running this casting, drosses caused by pouring created a lack of uniformity, shrinks often occurred and the bars were expensive and difficult to machine, being bent quite often in the process. The normal run of sand castings could give non-reproduceable results, particularly if gating and risering were inadequate.

The role of grain size, microstructure and density were not assayed for this report. It is felt that the phase changes have more effect on conductivity than grain size or density. The decrease in conductivity with increasing copper content, was probably due to phase change from a single phase to a two phase system.

Section size did not noticeably alter conductivity values. Surface finish was held constant, with 3/0 emery paper finish. A rough uneven finish certainly would be a deterrent to reproduceable results.

Some foundry problems in the development of a new marine propeller alloy, A. J. SMITH, Bethlehem Steel Co., Shipbuilding Div., Staten Island, N. Y.

W. G. YOUNG[9] *(written discussion):* The author is to be complimented on his complete presentation of the history of the development of this new alloy in his plant. While the aluminum bronzes are not new their application to large marine propellers is new and requires the solution of many problems before commercial production can be commenced. The author's company has been a leader in this field in the United States, and his paper is a valuable contribution to literature on the subject.

In his discussion of early experiments with heats in crucible furnaces, the author mentions that gating for A.S.T.M. test coupons was reduced in size under that for manganese bronze, in order to obtain maximum properties. We were unaware of this need, and would be interested to know how much the gates were reduced, and what was the effect that caused improved physical properties. Was it only a matter of keeping the dross to a minimum?

It is not clear whether the reverberatory furnace burners were altered to impinge on the metal (as mentioned in the text) or simply moved closer to the bath as stated in the summary. If the former is the case, we assume the burners are adjusted to take the flame off the bath when manganese bronze heats are required. Experience has shown that a flame impinging on the bath of a manganese bronze heat can cause a disastrous amount of oxides in the metal.

In the melting of the aluminum bronzes, as mentioned by the author, quick melt-down is the key to minimizing gas absorption. In converting reverberatory furnaces to this service, it is not uncommon to raise the flame temperature (and if necessary increase fuel consumption) by replacing some of the burner air with oxygen. This is of particular value in raising the heat the last few hundred degrees to the pouring temperature.

While controlling furnace atmosphere is difficult, as stated in the paper, we always attempt to maintain an oxidizing atmosphere. As much as possible, 2-3 per cent oxygen is maintained in the center of the flame length.

The author's illustrations of gassy heats are most interesting and a familiar sight to all who have worked with this alloy. We are aware of some foundries using a vacuum gas test routinely for aluminum bronze heats, even though, as the author states, this is not a quantitative test. Does the author feel there is any value in such a test on production work?

Degassing is commonly achieved by the use of dry nitrogen, and we have used a volume of 0.2-0.3 cu ft/100 lb of metal, delivered at a rate of 1½ cu ft/min. We have information of carbon dioxide being used for this purpose, but would fear the possibility of carbon pickup. Mention is made by the author to the use of an inert gas. Are we right in assuming this gas to be nitrogen even though it is not strictly speaking inert?

DEOXIDATION

The author makes no mention of the practice of deoxidizing the heat with magnesium, and we presume this is not the practice in his foundry. The precaution is not considered necessary by all foundries, although it is not harmful. Approximately 2 oz/100 lb of magnesium is held in a steel cage, and plunged into the ladle prior to degassing. What gage does the author have of the need for this precaution?

Mention is made of the need for the use of virgin metals in the manufacture of aluminum bronze ingot. This is generally acknowledged to be essential in the control of quality. However, means must be devised to handle such scrap, as may be marketed in the future, together with the manufacturer's normal production of skulls, flashing and turnings, which must now be disposed of outside the foundry. We would be interested to know the author's opinion of what may develop in this field. While the point is admittedly minor, it has a direct bearing on the manufacturers costs.

We completely agree with the author's statement that gating to minimize turbulence is the key to producing dross-free castings. We have done a moderate amount of experimenting with inert atmospheres in complicated molds, and, like the author, find them to be ineffective.

While our experience with aluminum bronze is limited, manganese bronze has a similar tendency to dross formation in the mold. It is not uncommon on heavily raked manganese bronze propellers, to run auxilliary gates from the base of the hub to the tip of each blade. This avoids tumbling metal down the steep blade face from the hub, and forming dross which will be trapped at the blade tip. Since dross is a serious problem on aluminum bronze, similar gating should be effective.

All dross which is delivered to the hub can be kept out of the blades by paddles while pouring, and kept in the head. Here, as the author suggests, it can act as a nucleus for solidification, and should be removed. However, it is common practice to insulate the head with material such as asbestos shorts, and this in itself may have a nuclei effect, making the removal of dross almost redundant.

The author's report on the effect of chilling the base of the hub is most interesting. Does he consider the results warrant continuing the practice for all large propellers? While our experience in such large propellers does not extend to aluminum bronze, we are aware of no difficulties which indicate a strong need for chills.

There seems to be no question but that the nickel aluminum bronze alloys are the best practical materials developed for large propellers in seawater service. While high strength is an advantage, there are practical limitations to its use, and the greatest value of the alloys is their high resistance to corrosion, cavitation-erosion and erosion. This is best exemplified by the first illustration of replacement of manganese bronze propellers in the author's paper.

LIMITING FACTORS

There are, however, practical limitations to the use of these alloys. Propellers in these alloys cost about 25 per cent more than in bronze, and in lower power installations the material cannot be justified economically on increased service life and higher efficiency. The second difficulty is that of repair. As indicated in the paper, aluminum bronzes can now be welded satisfactorily (provided composition is carefully controlled). Straightening of bent propeller blades is, however, a most difficult job, and many propellers damaged only moderately by standards for manganese bronze propellers prove impossible to heat and straighten in aluminum bronze. This factor must be considered when selecting propellers for many types of service.

A relatively new development in England is a high manganese nickel aluminum bronze. This material is claimed to have slightly improved physical properties, and to be much more easily handled in the foundry. Does the author have any experience with this alloy?

This excellent paper will be a valuable addition to the AFS records.

AUTHOR'S REPLY: Better Test bar properties were obtained when the gate was reduced from 1 in. as used for manganese bronze to ¾ in. This change resulted in a reduction of oxide entrapment. Other methods such as gating design changes would accomplish the same thing.

Burners were made adjustable so that the flame would impinge on the nickel aluminum bronze. They could then be changed so that manganese bronze would not receive this impingement. Burners which do not impinge but which fire close to the metal will work equally well.

Although we have discussed the practice of introducing oxygen into the oil-air mix we have not tried it up to the present time. I would expect that oxides would be increased and that the furnace refractory would suffer from the higher heat, unless drastic adjustments were made in the burners. We do, however, plan to check our opinion by using oxygen at an early date.

The principal value of the vacuum test for nickel aluminum bronze is in evaluating the results of changes in practice which might influence gas absorption of degassing effectiveness. After using it for several months as a routine control test, we found the results completely repetitive. Now it is used only for special tests.

Mr. Young is correct in assuming that the inert gas used is nitrogen. The volume used is approximately the same as that mentioned by Mr. Young.

Our practice involves the use of one lb of magnesium per 1000 lb of metal, plunged into the ladle immediately after tap-off. Theoretically the magnesium is supposed to dissolve the aluminum oxide suspended in the metal bath and allows the nitrogen flush to perform its function more effectively.

We have no concern over the use of scrap nickel aluminum bronze propellers when they become available for remelting purposes provided the material meets our specifications. Skulls and flashings can be remelted if they are thick enough to resist heavy oxidation. Turnings are reprocessed into ingot. Remelting any thin sections in a large heat generally results in heavy oxidation which in turn blankets the heat and slows up the melting operation. Metal losses are also increased.

Although tip gating has been successful in minimizing dross in manganese bronze propellers such gates have been troublesome when used with the nickel aluminum bronze alloy. Because of the smaller amount of superheat the metal tends to freeze in the small tip gate. When higher temperatures are used, shrinkage occurs from the blade feeding the gate. Strangely enough, dross entrapment in the blades of nickel aluminum bronze propellers is less than that found in manganese bronze propellers without tip gates.

Hot, dry and light insulating materials such as perlite do not have a nuclei effect on the riser top as the material floats. There is some danger, however, from materials such as plaster which readily absorb moisture from the atmosphere.

WATER-COOLED CHILLS

While the author feels that water cooled chills can greatly ease the feeding problems for nickel aluminum bronze propellers, they are not an absolute necessity and because of the potential danger involved, their use has been discontinued. However, the feeding of nickel aluminum bronze propeller castings is considerably more difficult than for manganese bronze propellers of similar dimensions.

The higher strength of nickel aluminum bronze does allow for a reduction in the scantlings of a given propeller with its resultant engineering advantages.

It is agreed that for what we term "normal service" the additional cost of nickel aluminum bronze can hardly be justified. However, there appears to be a misconception regarding the weldability of this material. Our experience, along with that of other manufacturers, indicates that the material is quite weldable. Straightening is not difficult but requires relatively high temperatures.

At present we have had little experience with the high manganese nickel aluminum bronze developed in England.

G. BRADSHAW[10] *(written discussion):* This paper offers a great deal of information on molding, melting and pouring of Ni-Al-Br propellers. The Philadelphia Naval Shipyard Foundry has also performed much work both in developing the alloy and its application to the manufacture of large propellers. The following comments are offered for your review and discussion.

Table 1 shows the chemical composition of the alloy and the corresponding mechanical properties. The Al range of 9.0-10.3 per cent is shown with an elongation range of 15-20 per cent. We are unable to obtain the above elongation values with a 10.3 per cent Al (Ni-Al-Br). We have found and use an Al range of 8.50-9.20 per cent which results in an elongation value of 20.34 per cent and equally as good tensile and yield properties as shown in Table 1.

MELTING PRACTICE

We, too, can corroborate the author's findings in that no difficulty was encountered in melting and pouring small heats from the oil-fired crucible furnaces. However, when Ni-Al-Br is melted in the reverberatory furnace in quantities of 40,000 to 45,000 lb, combustion problems are encountered to maintain the desired atmosphere. It has been established that a neutral atmosphere (which is almost impossible to maintain) or a slightly oxidizing atmosphere is desirable in melting Ni-Al-Br to prevent hydrogen absorption by the metal.

Because the combustion conditions are constantly changing due to fluctuations in the air and oil pressures at the burners, the shipyard foundry utilizes oxygen lances which are placed beside the burners to insure complete combustion and an oxidizing atmosphere. It may be argued that the melt loss is increased, however, the quality of the end product and the time saved in melt down warrants this increase in melt loss.

TREATMENT FOR ABSORBED GAS

As stated by the author, all heats are gassed to a certain extent. How does the foundryman know or what can he use as a guide to determine what will be detrimental and what can be considered tolerable in order to pour a sound casting? At present the vacuum test, and the solidification of a sample under atmospheric pressure, are the only rapid tests available. I think further work in devising a rapid method of gas content would be most useful in Ni-Al-Br casting manufacture.

The shipyard foundry never pours a propeller casting before the metal in the ladle has been thoroughly flushed with dry nitrogen and many test samples observed. In the early stages of experimenting with this alloy a small propeller casting was made without flushing the ladle with dry nitrogen and the results were poor. The hub sections showed large cavities caused by gas. Actually these were shrinkage cavities resulting from the excess gas not being able to escape through the riser.

In addition to the degassing operation with dry nitrogen, all large heats melted in the reverberatory furnace are also treated with 0.06 per cent to 0.125 per cent metallic magnesium (1 oz to 2 oz/100 lb of melt). This deoxidant is used based upon the theory that the Mg will react with and reduce the aluminum oxide and thus be skimmed off in the form of dross. It is also used to purify the metal from sulfur contamination which may occur from the oil. We have found that sounder castings result, more especially in the heavy hub sections, when metallic magnesium is used.

At this point your attention is called to the fact that the Mg actually alloys with the Ni-Al-Br and becomes an impurity. This fact must be carefully noted when remelting risers and gates. From a series of keel block specimens, and those cast as an integral part of the casting, no detrimental effects upon the mechanical properties have been noted. What effect the retained magnesium has upon the corrosion properties, if any, can not be answered at this time.

Another deoxidant called out in "The Development of High Tensile Aluminum Bronze Alloys For Marine Propellers" (published by the Mond Nickel Co., Ltd.) is listed for your information:

Element	Per Cent
Aluminum	37.0
Calcium	1.0
Manganese	6.0
Titanium	0.6
Magnesium	2.0
Copper	remainder

This alloy is added, in amounts of 0.5 to 1 per cent, to the molten metal about 5 min before pouring. To date this deoxidant has not been investigated by the shipyard foundry. Your comments are requested.

MOLDING MATERIALS

The author gives the advantages and disadvantages of cement bonded sand and natural bond baked gravel. He claims that he was unable to substantiate the fact that propellers can be cast in cement molds closer to the finished size with savings in the final machining operations. We are at present investigating the casting of propellers in cement molds to close finished dimensions.

To date two Mn-c wheels and one Ni-Al-Br wheel have been made. The results obtained indicate this procedure to be worth further investigation. However, the molding procedure employed should be left to the shop manufacturing the propeller.

EXOTHERMIC SLEEVES

The findings in the paper state the use of exothermic sleeves are only partially effective and costly. We feel that the use of an exothermic sleeve hot top compound, and the hot head pots are all necessary to keep the riser open and obtain the desired directional solidification, evolution of gas and proper feed. However, again this decision must be left to the shop manufacturing the casting.

DROSS OR OXIDE ENTRAPMENT

In my opinion, the tendency toward dross (aluminum oxide) entrapment in the casting is still one of the problems to be overcome in casting Ni-Al-Br. The high skin tenacity which is characteristic of this alloy, results in dross entrapment which is found beneath the as-cast surface and is noted only after the casting is machined.

Your comments on this condition are invited.

Engineering study is required on this alloy, in order that it may be recommended for use on pressure castings such as valves and pumps.

AUTHOR'S REPLY: The aluminum range has been changed from 9.0 - 10.3 per cent to 8.5 to 11.0 per cent (Ref. MII-B 21230). Most heats cast by our foundry are in the range 8.5 - 9.5 per cent. It is agreed that the higher aluminum content poses a difficulty in meeting the minimum elongation requirement.

We cannot comment with any degree of authority on the advantages of oxygen in the combustion atmosphere as we have not tried it to date.*

It is felt that until such time as a positive rapid quantitative test for tolerable gas content is developed, flushing of all heats with nitrogen is advisable.

The use of magnesium at the Philadelphia Naval Shipyard Foundry is similar to that used in our foundry. We have had no experience with the deoxidant referred to in the Mond publication.

We appreciate the feelings of the writer with regard to the use of exothermic sleeves. However, as stated in the paper, such sleeves are only partially effective for large castings because of the long time required for solidification. When the heat from the exothermic material is needed, it is not available. Even manufacturers of exothermic compounds have agreed with us on this reasoning.

There is no question about dross entrapment being a major problem in producing nickel aluminum bronze castings. Further refinements in pouring and gating practices are required to overcome this condition. Most of the drossing referred to, however, has no appreciable depth. Service followup has shown no ill effects from the dross, at least for our particular application.

Studies of nickel aluminum bronze castings in other types of applications are currently under way.

W. B. SCOTT[8] *(written discussion):* Mr. Smith deserves a great deal of credit for attempting to marshal all the factors involved in propeller casting into one single paper. No one can be truly sympathetic with Mr. Smith's effort unless they themselves have been through the casting process of these large specialized castings.

There is in the paper, however, ample room for criticism of the foundry trade in general. The casting of large marine propellers is a fairly limited activity confined to perhaps five or six highly specialized shops. These shops have cast propellers in all sizes for many years.

It is surprising therefore that such a small group of producers could not have gotten together long before this to answer some of the fundamental problems involved in adopting a higher strength bronze to the application. I believe the impetus for the use of aluminum bronze propellers came from Europe, and provided the first challenge to the tradition of our own marine practice.

NEW APPLICATION

It should be further noted in regard to the title of Mr. Smith's paper that nickel bearing aluminium bronze is not a new alloy, only the application to the large propeller castings is new. The old Air Corps specification 11076, in use at the time of Pearl Harbor, was essentially this same alloy, and the A.S.T.M. specifications go back at least 15 years or more. There was not any appreciable development of 50,000 lb castings that many years ago, but there were aluminium bronze castings of 5, 10 and 20,000 lb size used in steel mills and chemical plants.

The nickel content of these castings varied from ½ per cent to the full 5 per cent of the alloy under consideration. Melting practice was fairly well established by foundries supplying small and medium sized castings to the machine tool trade in the early 1930's. This commentator had a hand in making small racing propellers at that time, and the small variable pitch propellers on the landing barges of World War II were made of aluminium bronze.

It is true, however, that none of the foundries producing these castings were using reverberatory furnaces. However, such large melts were made by smelters, there being at least three in the Chicago area that I know used reverberatory-type furnaces. All of this reflects on the foundry industry that failed to make know-how available to the application Mr. Smith had to deal with. Perhaps, however, nobody was asked for this information.

CRUCIBLE HEATS

The author's account of his preliminary crucible heats leads one to wonder why so much attention was paid to this endeavor. Certainly aluminium bronze has been crucible melted for years with good results. Moreover, reverberatory melting develops factors of atmosphere and heating rates that have no counterpart in crucible practice. Any experimentation on crucible melting could not be readily translated into terms of reverberatory problems.

To embark on the reverberatory melting of aluminium bronze with only the simple concept of reducing-neutral-oxidizing conditions as a guide is to court disaster. Slow, inefficient heating rates may be satisfactory for low tensile manganese bronze. With aluminium bronze it is necessary to heat rapidly and efficiently. Aluminium will oxidize to some extent under practically any fossil fuel combustion atmosphere, and as such the definition of oxidizing or reducing atmosphere is somewhat meaningless.

Excess air or excess fuel lowers the rate of heating, extends the heating time and contributes to faulty metal. Furthermore, the construction of a reverberatory furnace would make the definition of equilibrium conditions of atmosphere within the furnace entirely arbitrary.

GOOD FURNACE PRACTICE

It is this commentators experience that what was good furnace practice for manganese bronze was good practice for aluminium bronze. The metal charge should be as carefully stacked for the one as the other. Generally it was found that stacking at the flue end of the furnace and working back toward the burner was the proper procedure. This allowed the metal to be in the path of the highest temperature combustion gasses.

Large slabs of metal are preferred to ingot for easy, safe, furnace positioning. The slabs are best placed on end rather than flat. When ingot is used it is necessary to hand stack on wood blocks to protect the furnace side walls and allow heat venting to the floor. Stacking of banded pallets has some advantage, but the burning wood may cause load shifting and damage to the furnace walls. It should be understood that in loading a reverberatory furnace the arch sections of the roof are placed in position after the furnace loading is complete.

Seldom is the material loaded through the doors of the furnace. It is essential to maintain good roof section joints, even to the point of sealing, in order to maintain good heating conditions within the furnace. Burners and fuel lines should be kept clean and at full capacity. The angle of the burners may be the same for both manganese and aluminium bronze with satisfactory results. Flue tunnels should be kept clean and

*Reference is made to answers on comments submitted by W. G. Young, on this subject.

at optimum capacity. Heating rates are a better criterion of furnace conditions than gas sampling.

GAS SAMPLES AND TEMPERATURE

I would be interested in knowing how Mr. Smith took gas samples and temperatures in attempting to analyze the conditions in the reverberatory furnace. The conventional pyrometer protection tubes are cumbersome and difficult to handle in taking the temperature of the metal bath. Maintenance is high on such pyrometers. As a result, an air-cooled light-weight protection tube was found to be much more satisfactory.

Mr. Smith's account of his efforts to use flux coatings in the reverberatory gives this commentator a bit of moral satisfaction. At the time Mr. Smith apparently was doing his work, we were besieged and beleaguered by proponents of the flux treatment to cover our bath with the proprietary fluxes as was claimed was the European practice copied by Bethlehem Ship with a high degree of success.

Since the metal bath in a reverberatory is heated by radiation and convection it seemed foolish to cover the bath with an insulation such as the flux provided. We never tried it. We did find that a small charge of fluoride-base flux fed to the metal being tapped from the furnace kept the accumulated dross in the ladle to a manageable metal cover.

We knew from prior experience on small melts that an excess of fluoride-chloride flux could cause trouble from mold surface reactions were it to evade skimming and reach the mold cavity. This was particularly true of those flux mixtures that used borax as extenders.

GAS ABSORPTION

I would agree with the author as to the susceptibility of aluminium bronze to gas absorption. The nickel bearing aluminium bronzes are the most difficult to handle of all the aluminium bronzes. The nickel seems to lower the temperature of hydrogen evolution below the solidification point. This means that chilled castings or small castings may appear quite sound and free of gas. Heat treatment or long solidification time of large castings appears to permit the hydrogen to migrate to internal centers, causing rupture or feeding failures.

Gassed metals will also prevent satisfactory welding of these alloys. What may not be recognized in many cases is that the aluminium oxide film that ordinarily covers aluminium bronze could protect the metal from gas adsorption providing that film is not broken by mechanical agitation of the metal. This may account for the successful welding of aluminium bronze in large stationary reverberatories. It is also true that a properly operated reverberatory operates under a slight vacuum, and by control of furnace leakage the metal may be protected by an envelope of air.

GAS CONTENT

The author stresses the quantitative factor of gas content with which one must agree. Still no one knows how much gas content is a safe content. The foundry with which this commentator was associated tried vacuum bell testing but found such tests of little value, and we were unable to prepare any kind of a standard of comparison. We did use a shrink test which was inherited from a predecessor along with considerable mumbo-jumbo.

No one actually knew what a good shrink was although there was evidence that a uniform shrink rate was indication of a low gas content. Whenever the shrink accelerated to produce tangent root effects, we generally were faced with a marginal quality metal. I would like to ask if the samples Mr. Smith shows of gassed metal were taken from production heats or test heats? If these were test heats how was the metal gassed? Did Mr. Smith ever use lithium or magnesium to degas his metal?

I would like to ask Mr. Smith if he thinks the vacuum stream degassing technique so successfully used by his parent company for steel forging ingots could not be adapted to propeller casting?

I would also like to ask Mr. Smith to elaborate on the inert gas flushing which he suggested was the final solution of their problem. I assume the gas was nitrogen. What pressures and what time cycles were used, and at what temperatures?

MOLDING PROCESS

In regard to molding of propeller castings, the plant with which I was associated used the cement molding technique for large propellers. Gravel molds were used for smaller pieces. The chief advantage was that the large mold sections in cement did not require bake ovens. There was also an indication that the cement molds had a higher heat dissipation factor that contributed to the favorable thermal gradients.

Only prejudice could decide that better propellers were made by cement or gravel in that satisfactory wheels have been made by both methods of molding. It was true that an improperly air dried cement mold resulted in gassed metal castings even when satisfactory metal was delivered to the mold. A common source of trouble was the use of oil heaters from which the products of combustion were blown down the cold mold.

This was almost a certain way to produce gassed castings. Inadequate venting of the cement molds during drying could also contribute to gassed metal.

In Fig. 6 of Mr. Smith's paper, he shows a series of time temperature curves. Can he make any explanation of the cyclic thermal gradients that appear to have existed? Moreover, the curves indicate thermocouple readings of 10 degrees increment or less. This indicates a pyrometer technology of high order and may further suggest the reason for the cyclic effect noted.

INSULATING MATERIALS

The use of plaster, calcium sulphate, might have been anticipated as a failure. There are other insulating materials that are effective with the aluminium bronzes. Mr. Smith has done a good job of analyzing the action of exothermic materials and discounting their effectiveness on large castings. Whoever promoted the use of the water-cooled bottom chill showed considerable courage and some indifference to the safety of the shop personnel. Is the hazard worth the accomplishment, and could the same results have been accomplished with less hazardous equipment?

Mr. Smith passes lightly over the welding characteristics of the nickel bearing aluminium bronzes. I do not believe that the Bureau of Ships has yet approved a welding technique for propellers of this material. Welding has been successfully accomplished, but the standards of the test procedure as originally proposed and specified was somewhat unrealistic. The straightening of damaged propellers is more difficult not necessarily impossible, but definitely more difficult than with the lower strength manganese bronzes.

In closing these comments, I would like to express regret that Mr. Smith saw fit to crowd so much of a story into one paper. Many of the phases of his paper could have been more fully explained in individual papers to the education of other foundrymen and enlargement of the program scope.

AUTHOR'S REPLY: Mr. Scott's lengthy discussion is much appreciated. He has brought up several controversial points which are stimulating for a discussion on this type of paper.

First of all it must be appreciated that there is little if any free exchange of information between manufacturers in highly competitive industry. Despite the pleas of technical societies, management is certainly justified in guarding the results of productive research and development even to the disadvantage of the general trade. It is believed that most of us are cognizant of this fact.

ALLOY PROBLEMS

The significance of the title refers to the development of a practice for our foundry of an alloy which was not only new to us but to the entire propeller field. Many problems had to be solved before good marine propeller castings could be produced, and to this day many foundries are unable to manufacture good castings in this material. It should be pointed out that although smelters were producing this metal in reverberatory furnaces, they were making ingots, not castings, and the problems involved in making castings, especially large castings, are, to say the least, considerably more complex.

All possible information was obtained from smelters regarding melt quality-control and melting practice. Even here however, little of this information applied to foundry operations.

We feel that it is poor policy to start an experimental and development program using production equipment. Much valuable information and experience was obtained by beginning our program with crucible furnace melting. Certainly it would have been folly to experiment with large tonnages of metal in reverberatory furnaces before cutting our teeth on the smaller operation.

Although we agree with some of Mr. Scott's points on reverberatory melting practice and feel that they have already been mentioned in the report, we disagree with some of his practices — at least from the standpoint of what works best in our furnaces. For example. we find that stacking at the burner end of the furnace gives us the fastest possible melting. This difference may be due to dissimilarity of furnace and burner design. Slabs can be satisfactorily melted in the flat position provided they are spaced and criss-crossed.

As pointed out in the subject paper the importance of burner position was evaluated very carefully on both the experimental reverberatory furnaces and the three production reverberatory furnaces and has since been substantiated by other shops. Burners firing parallel to the furnace bottom but placed relatively close to the metal bath can be used for both manganese bronze and nickel aluminum bronze. Burners comparatively remote from the metal bath, such as those designed to cut the zinc losses in manganese bronze melting, must be tilted or directed at the metal if fast melting is to be obtained, with nickel aluminum bronze.

A long-handled spoon is used to remove the metal sample from the furnace. This is poured into a bone-ash coated Armco iron crucible which is in turn placed under the bell jar on the vacuum tester.

Temperatures are taken with long-handled chomel-alumel thermocouples inserted through the furnace mixing door. The handles are refractory coated to prevent sagging. Temperatures are read on a recording electronic potentiometer attached to the furnace. We find this system very economical as the thermocouple tips can be replaced and generally last for several months.

It is agreed that certain foundry sales personnel carry stories between competitors regarding the alleged successful use of their product. However, we are interested in running the experiment before the conclusions are drawn. That is the reason that tests were run to determine the influence of proprietary fluxes. In this way our conclusions have some foundation. They are not based upon opinion. Development of our practices was accomplished without the aid of copying European practices. Mr. Scott's statement that the metal bath in a reverberatory furnace is heated by radiation and convection is only partially factual. As our experimental furnace runs show, direct flame impingement (or proximity to the metal bath) is necessary for rapid melting. This was further substantiated in our production furnace experience and the experience of some of our smelting and refining contacts.

Fluxes for this service are not intended to provide insulation but rather to dissolve the oxides formed during melting and thereby *reduce* metal bath insulation and metal losses. The fact that the intended service was not performed could only be determined by actual test. The use of fluoride flux additions to the metal during tap-out can cause an unpleasant and dangerous pollution of the foundry atmosphere unless adequate ventilation has been provided. Few foundries are so equipped.

HYDROGEN INFLUENCE

Mr. Scott's discourse on the influence and manifestations of hydrogen differ from the author's viewpoint. I would leave this to further laboratory or pilot plant studies. However the author is not aware of the influence of dissolved hydrogen upon the weldability of the alloy.

It is agreed and quite readily accepted that gas absorption can be prevented or reduced by leaving the skin of oxide on the molten metal undisturbed. However the author doubts the existence of large production reverberatory furnaces which can be controlled so that no air leaks exist. Certainly he has not seen any in his visits to many foundries and smelting shops.

The vacuum test samples illustrated in the subject paper were taken from production heats. The additional slides shown during the presentation of the paper were supplemental and were taken from experimental reverberatory furnace heats. Gassing was unintentional except where noted. As stated in the paper all heats are gassed during melting. Although no longer use as a production control test much valuable data were obtained with the vacuum tester, and it is anticipated that it will provide us with a means for determining the effectiveness of further changes in degassing methods. The vacuum tester is a far more reliable method of testing for gassed metal than the atmospheric shrink test and we have tried to clarify this in the subject paper.

We have not used lithium in nickel aluminum bronze. Magnesium is used but not for degassing purposes. Lithium and magnesium are strong deoxidizers and have no influence upon hydrogen. Magnesium is used to reduced Al_2O_3 and allows the nitrogen flush to perform its function of hydrogen removal more effectively.

The stream degassing technique which has been successfully used on steel by our company could conceivably be used for nickel aluminum bronze degassing. However, there are several problems involved. The steel degassing is done for steel billets or ingots so that the metal is allowed to solidify in the vacuum chamber. To pour a casting from metal treated in this manner would require rapid removal of the ladle from the vacuum chamber, a feat which at the present state of development would be difficult, if not impossible. Also the cost of such equipment could hardly be justified in a brass foundry. Equipment size could also act as a limiting factor. I believe that this is one for the future of our industry.

The procedure for inert gas flushing is a simple one. Nitrogen gas is allowed to pass through a steel, refractory coated tube under very low pressure. Pressure is regulated just low enough to keep the metal from splashing out of the ladle. The nitrogen flow is begun when there is approximately twelve inches of metal in the ladle and is continued until the ladle is full. After the magnesium treatment the flushing is continued until approximately 500 cu ft of nitrogen has been used. The temperature of the metal during this period should be between 2150 F and 2250 F.

We concluded that the disadvantages of the cement sand outweighed its advantages for use in our shop and under our conditions of operation. Each shop of course would have its own reasons for preference of molding materials.

Unfortunately an error was made in plotting two points on the graphs shown in Fig. 6. This does not change the fundamental direction of the curve. The corrections are noted below.

3rd curve—15 min after cast—change point 2 from 1660 to 1670 F.
4th curve—24 min after cast—change point 1 from 1645 to 1675 F.

We feel that Mr. Scott is second guessing when he states that the use of plaster might have been anticipated a failure.

In the writer's opinion the use of water cooled chills in the foundry, if properly supervised, is no more a hazard than some other so called "normal" foundry operations. It is further his opinion at the moment that there is no substitute offering the advantages of water cooled chills. It must be admitted, however, that these opinions are not equally shared by all personnel in the author's company.

We are not quite sure of Mr. Scott's implication regarding the lack of welding approval by the Bureau of Ships. Our company has been given approval of its recommended welding practice for nickel aluminum bronze castings and it is understood that other manufacturers have received similar approval. From our experience there are no more problems involved in welding nickel aluminum bronze than exist for manganese bronze. It is true that the straightening operation is more difficult for nickel aluminum bronze as higher temperatures are required.

M. H. DAVISON[11] *(written discussion):* After reading your article "Some Foundry Problems in the Development of a New Marine Propeller Alloy" I feel that you should be complimented on the thoroughness of your undertaking in this development. I was especially interested in your use of a cast iron chill having a 1-in. diameter spiral copper coil and the presented data that you obtained.

At present, we are evaluating the use of water-cooled chills for some of our foundry casting applications. We have cast several small arc furnace heats of low alloy steel against water-cooled chills and have established working feasibility for our applications. We are now in our final development phase, aimed at optimizing and controlling the temperature distribution in the molten metal.

I am interested in learning more about your work with water-cooled chill; i.e., chill sizes and designs, more detailed information of the time temperature distributions in the chills and castings, the extent of your present use of water-cooled chills, etc. If you are still using water-cooled chills, any information you have gathered on them would be appreciated.

M. A. FUREY[12] *(written discussion):* I would like to congratulate you for your article on propeller development.

Your treatment of the subject was of paramount interest to

us, since our efforts are along the same lines, and it confirms a number of our own ideas.

Since this Navy Yard Foundry is now engaged mainly in the production of Ni-Al-Br propellers, your mention of degassing with inert gasses is of particular interest to us.

We are now using dry helium, but our method of application could be improved. At present, the helium is being sent through a calcium chloride tower and dispersed through the melt by flowing through a graphite rod and out of a porous graphite head.

If you could supply us with a description or drawing of your system for degassing a 15 ton charge, we would be grateful.

AUTHOR'S REPLY: Thank you Mr. Furey for your interest and comments on my AFS paper.

Your method of degassing is probably more elaborate than that which we use at Staten Island. Our practice is to use water pumped nitrogen. The specifications for this product are: 20 ppm or less of water, a dew point of −100 or better and 99.99 per cent pure. The manufacturer assures us that it would not benefit us to run this gas through a drying train.

Our method of degassing is as follows: a 12 ft length of refractory coated, ¼-in. diameter pipe it attached to the pressure gage of a full bottle of nitrogen by means of a 6 ft length of heavy wall rubber hose. About 1 min after the metal has started into the ladle (no degassing is done in the furnace) the bubbling is begun. About ⅓ bottle of nitrogen is used for each ladle (30,000 to 50,000 lb of metal). The pressure must be kept low so that metal is not splashed onto the operator. A protective cover is laid over part of the ladle to guard against excessive splashing.

With the vacuum tester we have found that all heats of nickel aluminum bronze are gassed during melting, and so as a matter of course all heats are degassed. Although the degassing treatment described does not by any means completely degas a melt. it is sufficiently effective to allow us to produce good castings.

A more desirable practice might be worked up if we were able to make a quantitative determination of the hydrogen present in the melt and the amount of degassing necessary for effective treatment. However, up to the present time such a system has not been worked out.

Pattern Division

Presiding — W. E. MASON, Westinghouse Air Brake Co., Wilmerding, Pa.
N. C. JONES, New York Air Brake Co., Watertown, N. Y.

Secretary — J. W. TIERNEY, Houghton Laboratories, Inc., Olean, N. Y.

Pattern Engineering Materials, new horizons, J. E. OLSON, Dike-O-Seal, Inc., Chicago.

MEMBER: Will the wear strips adhere to steel?

J. E. OLSON: The wear strips will bond adequately provided the proper adhesive is used.

J. KREINER:[13] Will the material bond to old squeeze plates?

J. E. OLSON: The bond is good provided the proper adhesive is used and the plate has been cleaned.

MEMBER: What is its resistance to hot sand?

J. E. OLSON: The resistance to hot sand is good up to 250 F.

MEMBER: What is the shrinkage?

J. E. OLSON: The shrinkage is good and also a possibility of some expansion. It may become slightly harder over long periods of time but will not change dimensionally.

Sand Division

Presiding — R. H. JACOBY, St. Louis Coke and Foundry Co., St. Louis.
W. A. WRIGHT, Woodruff and Edwards, Inc., Elgin, Ill.

Secretary — W. O. MCFATRIDGE, International Harvester Co., Chicago.

Experimental Determination of Specific Surface and Grain Shape of Foundry Sands, FRANZ HOFMANN, George Fischer, Ltd., Schaffhausen, Switzerland.

Presented by — V. ROWELL, Harry W. Dietert Co., Detroit.

G. DI SYLVESTRO:[14] Have you found that some of these sands are peculiar in the way of depending upon how coarse or at which end of the screens they are? That they will go from round to angular, or vice-versa?

V. ROWELL: Actually, I am not quite sure I understand, but I will try to answer. It is generally believed and quite true that the finer grains of a given deposit, simply because they have less mass and are not subject to as much stress or strain in weathering, do not get the corners knocked off, and wind up being more coarse than the coarser grain from the same deposit. In such a case, we still have the specific surface angularity which will make it comparable to another sand of similar grain distribution.

Grain distribution is a study in itself, and I do not think that it is quite fair to compare sands of widely differing distributions strictly on the basis of specific surface or angularity. As a matter of fact, I personally feel that of the most significant properties we can measure, distribution is more important than shape.

G. DI SYLVESTRO: One more point. As you showed, you had some means of measuring any one of those grains, and you could tell whether you were looking at large or small grains. If you had a round grain, and you measured it as 20 mesh, and you had some small grains that were angular, you can see that in the sand it can be round or angular or subangular depending on how many large or small grains you have.

V. ROWELL: That is right. In other words, when we take our individual sieve fractions to compute the theoretical specific surface of equivalent size spheres on coarser screens, the true angularity is rather low, and on the finer screens the angularity becomes progressively higher. That may or may not fall into a predictable proportion. I do not know.

MEMBER: What do you mean by angularity?

V. ROWELL: To me, if I may attempt to define it rather crudely, it is the degree which a specific particle digresses from a sphere. In other words, the further away from spherical it is, the more angular it is.

MEMBER: Cubes and elongates are more angular than round sand, is that correct?

V. ROWELL: I would say so.

MEMBER: Why is it that in the method described the cubes show up rounder, or elongates even rounder?

V. ROWELL: They do not. This is a possibility that has come up in informal discussions before. If we use this permeability method to measure specific surface of a hypothetical sand, consisting of uniform size cubes, it is possible that flat surfaces would be in close enough contact after thorough rapping that the per cent solid would be abnormally high and it might not be possible to measure the specific surface of these hypothetical particles.

MEMBER: You spoke about a piece of equipment in your paper. What does it replace, or what would it combine with to give us the results for our evaluation of foundry sands?

V. ROWELL: It appears to me that the test offers distinct possibilities. It replaces the individual evaluation of what is seen through the microscope. It has been established that commercial sands, when ranked according to visual angularity, seem to invariably fall into the same order as classified by this method.

The ranks are forward in the mind, so there must be something to it as it gives us something better than opinion. If we write to a friend a couple of hundred miles away describing a surface of a sand we call subangular, what we feel in our mind as subangular is exactly the same thing he feels is subangular.

I have had the situation that when I have looked at a sand sample one time and called it rounded; a week later I have looked at the same sample and called it subangular. Which is right? A more uniform procedure would be to have a measurement.

Getting back to the question asked previously about defining angularity — it is an increase in specific surface or surface area of a solid of a given mass as it departs from the absolute minimum surface area to mass of a sphere. Your smallest surface for a pound of any particular material will be in the form of a sphere. You take that pound and arrange it as anything but a sphere, and you have more surface. Flatten it out and you have still more surface.

MEMBER: In a 1947 investigation of a society, they found there were two things that develop inaccuracy to reporting this type

of thing, 1) The method of compaction – the material must be compacted to its maximum or there will be erroneous data, and 2) surface area of a distribution of particles cannot actually be determined because accuracy is lost as the particles have broader distribution. However, it is reasonably accurate if you have broader distribution. However, it is reasonably accurate if you have single particles, but the instrument becomes more inaccurate as the distribution broadens. However, perhaps we are working in the right direction, trying to determine the surface area of the foundry sand distribution.

V. ROWELL: That is a good point. One of the difficulties is simply the fact that it is hard to take a grain of sand and take a yard stick or micrometer and measure it and find out what the surface is because of the grain irregularity. Any approach has to be empirical, and we have to start with some assumptions. I do not think there is any way to get away from this. Quite a number of things are considered here.

MEMBER: I want to mention that the Basic Concepts Committee has gone on record with figures for determining the surface area of specific particle sizes after a thorough investigation. The figures decided upon coincide closely with the figures which Dr. Hofmann has reported in his paper. But, several researches have not been cross-checked. The figures Dr. Hofmann reports here are quite satisfactory.

V. ROWELL: I think, basically, it is intended more as a means of comparing sands than it is to be absolutely unquestionably accurate. I do not think Dr. Hofmann would insist that this is the final answer. However, it does work with normal sands. I believe it does advance the art of grain shape classification a little bit beyond what we have had hitherto. I have a great deal of respect for Dr. Hofmann, and knowing the man I think he would be the first to admit that it is not perfect. There are some assumptions that are empirical; but it is better than nothing. I think it is a lot better.

Optimum CO_2 Molding, critical formulations of sodium silicate and sand, R. J. COWLES, Walworth Co., Braintree, Mass.

MEMBER: What kind of pattern equipment did you use during the experiment, and what kind of variations can be expected on the basis of different kinds of pattern equipment?

R. J. COWLES: This was a laboratory study, and although we did go to the plant to prove out the properties of mix flowability, collapsibility and surface condition of the castings as predicted through laboratory studies, the plant studies were kept at a minimum. We used wooden equipment for cores of various sizes. The cores used in the laboratory studies were all made in a metal cylinder with a metal ram. This equipment has a 4 sq in. cross-sectional area, and the surface properties determined in the furnace gave us some idea of the resulting surface we would get in foundry operations. This was borne out in some plant trials. I would rather not say that we used this study directly in foundry operations.

The study is prepared for a variety of metals and a variety of sands, and if we attempted to study each one separately on a cut and try basis, we would be working for years. Rather than do that, we studied it from this particular approach where we could get more of a universal-type answer that would be applicable to the variety of sands and metals, and possibly to the variety of equipment in use. We are using considerable epoxy patterns which are favorably applicable for the sodium silicate process. The problems of mold or core release from pattern equipment is almost completely obviated by using the proper percentage of sodium silicate for each sand as determined by these laboratory methods.

G. DI SYLVESTRO:[14] You mentioned the use of iron oxide in some of your formulations. What advantages do you get from the use of iron oxide, and what was the chemistry of the iron oxide which was used?

R. J. COWLES: An iron oxide-soda-silica eutectic composition appears to form at the metal surface. This is also caused by ferrous metal on the mold surface when pouring iron containing metal; but we used iron oxide to prepare this surface which at the high temperatures of the molten metal form the eutectic compositions.

For me to present the complicated chemistry and physical properties of these various compositions is impossible. We do not know just exactly what is formed, but taking published ceramic data presented as phase diagrams of these ceramic materials, we can see that there are these possibilities (see Fig. 6). The powder, iron oxide, did not increase the binder requirements considered as per cent sodium silicate. It became a part of the binder adhesive and is quite compatible with the sodium silicate. This is not true with clays.

Clays have to be selectively chosen, and they can interfere with the sodium silicate because many clays will agglomerate considerably at high pH's (see reference 30). Sodium silicate is a high pH material. So we have to be careful when we use clay. Iron oxide gives us the high heat properties, the cushioning effect evidently, for increasing collapsibility and getting better surface.

MEMBER: Is that a normal silica powder that was used – between 90 to 140?

R. J. COWLES: I would say that the greatest percentage of a 90 silica flour is through 90 mesh. In other words you have what is + 90 per cent through a 90 mesh screen, or through a 140 mesh screen in the case of a 140 silica flour. If you are talking about a 200 mesh silica flour or 325 silica flour, this all has to go through, that is the greatest percentage, through the designated screen, and has to meet your AFS requirements. The AFS SAND HANDBOOK gives a description of what these silica flours are, and they are available from many of the sand producers.

This 90 to 140 is not meant as a range, it is meant that we chose standard silica flours, designated at 90 or 140, or somewhere in that range. Individual silica flours which were in this designation of 90 to 140 proved advantageous, but when we used 325, for instance, the effect of our adhesive was impaired. So we had to stay in the coarser range of flours. This supports the hypothesis of point to point contact adhesion between sand grains.

New Foundry Resins and Application Techniques for Shell Molds and Shell Cores, W. C. CAPEHART, Monsanto Chemical Co., Springfield, Mass.

MEMBER: Should we settle at a certain hexamethylenetetramine content? In most powdered resins today we have what we might say is 12 to 13 per cent of the resin binder is hexa.

W. C. CAPEHART: The resin will cure with a minimum of hexa, i.e., around 7 per cent. However, your time to cure is slower than with the 12 to 14 per cent hexa. The 14 per cent hexa was selected because shell and core production rates demanded it. We know this, that with higher hexa, we cross-link the resin more (and we do build in extra hot strength by greater cross-linking with greater amounts of hexa although we do sacrifice some of our cold tensile), and lower the thermoplasticity of the bond.

MEMBER: Do you feel that perhaps there is some minimum requirement for the hexa?

W. C. CAPEHART: This I do not know. That is something that as you use resins, primarily with the liquid type where you are adding your hexa to tailor it to the desired properties, you experiment to arrive at a point you find best suited to your operation whether it be 8, 10, 12, 14, and in some cases, 16 per cent hexa. This 16 per cent hexa is used because some people want a fast curing resin so they can eject the shell rapidly and secure high production rates. Sometimes a 18 per cent hexa is used. But these users are sacrificing some of their cold tensile strength and they are getting a more brittle shell with the greater cross-linking.

MEMBER: Is it possible to make a water-borne system with no volatiles? You mentioned no volatiles, but all of them have volatiles in them.

W. C. CAPEHART: I did not mean to leave that idea at all. Actually, your water-borne systems in one-stage resin systems the carriers are primarily water; in the two-stage system you have some juggling where you have the majority water, and a lower concentration of volatile solvents, assuming you mean by volatiles alcohol or some similar types of solvents. We have some resins that at 215 F the solvent system boils away with no flash. Until the dried resins reach the burning point there is no flash. So the flash point we can say is well above 215 F.

MEMBER: You mentioned the addition of some thermoplastic material to cut down on thermal cracks. What type of material is this, and what per cent was added?

W. C. CAPEHART: A proprietary material. In this case, a range of about 7 per cent of your resin solid would be a fairly good range. If we go too high, than we get into dimensional distortion. So it is something you have to determine by your hot

deflection tests or castings, and the type of resin you use. Some resins and castings can tolerate more than 7 per cent, and some less. However, 7 per cent is an average number.

Presiding — T. W. SEATON, American Silica Co., Ottawa, Ill.
R. H. OLMSTED, Whitehead Bros., New York.
Secretary — J. G. SMILLIE, John Deere & Co., Moline, Ill.

How to Avoid Sand Segregation, by W. D. CHADWICK, Manley Sand Co., Rockton, Ill.

G. DI SYLVESTRO:[14] Has any work been done on hanging chains or belting in the bins to break up the nonuniformity in filling?

W. D. CHADWICK: Years ago some people tried various types of dangling belting or chains to change the segregation patterns. From what we know of dry sand, and these experiments only apply to dry sand (any moisture, bond, clay, cereal, etc., was not taken into consideration), we think that if you will let sand segregate normally or naturally in the bin which is what it will do if you do not try to disturb it, and then remove the sand in the same proportion of coarse, medium and fine and blend it back together, that you will come out closer to the base sand originally introduced.

You can do this by some type of discharge where you will get a cross-section of the storage bin. In other words, if you can lower the sand in the bin as a unit, then when it comes down and a cross-section is taken, you are getting the same proportion of coarse, medium and fine. It is easier to do it this way and less expensive.

G. DI SYLVESTRO: In other words, the loading of the hopper and unloading it in trucks is the important part?

W. D. CHADWICK: That is right. If you can lower the bin as a unit and do not change the segregation characteristics in the bin, then you can discharge ¼, ½ or ¾ of the bin and recharge the bin and you do not have an extreme segregation problem at the point where you started to recharge. If you discharge through one or two holes so you have a funneling effect, then that funnel must be filled, and the pyramid of sand is built back up and has extreme segregation in that area of two cones set end to end together at that point.

MEMBER: In a boxcar, if the coarse sand has tumbled to the sides, and you go in with a scoop truck or a shovel and get a concentration of that, would it upset your core mix?

W. D. CHADWICK: You must know how your sand producer loaded the boxcar. Then you will know the segregation characteristics of the sand in the car. If it is thrown in, sand will be piled up one way with the fines in the middle and the coarse along the sides. On the committee recommendation, you unload the car so that you are unloading across the car, then you will get a proportionate amount of coarse, medium and fine with a little more fine in some places due to the hump of the sand. You will get a better cross-section of the material into the bin this way. Taking a sample of the sand from near the door, even though the sand has been loaded uniformly, will probably show that the sand is way off from the specifications.

MEMBER: Since there are ripples at the top of the silo, should you try to use these to distribute the sand?

W. D. CHADWICK: No experimental work was done on this.

Ramming, Superheat and Alloys (Type of Metal) Effect on Metal Penetration, G. J. VINGAS, Magnet Cove Barium Corp., Des Plaines, Ill.

MEMBER: At the same pouring temperature, with a soft ram and a hard ram, would not the one with the soft ram penetrate worse?

G. J. VINGAS: One casting of the experiment, rammed to four different degrees, was penetrated almost identically.

Presiding — B. H. BOOTH, Carpenter Bros., Inc., Milwaukee.
W. R. MOGGRIDGE, Ford Motor Co., Windsor, Ont., Canada.
Secretary — G. DISYLVESTRO, American Colloid Co., Skokie, Ill.

Green Tensile and Shear Strengths of Molding Sands, by R. W. HEINE, University of Wisconsin, Madison; and E. H. KING and J. S. SCHUMACHER, The Hill and Griffith Co., Cincinnati.

D. WILLIAMS:[15] I would like to present to you an examination of the data in this excellent paper that will show that we can get more mileage than indicated in the paper.

First of all, there is Coulomb's equation for the failure of granular materials. If this well known equation is valid, perhaps we ought to take another look at the data, and, perhaps, we ought to take another look at our testing procedure. Because, if this standard equation, which the soil mechanics people use to evaluate the soil on which large buildings stand is valid, perhaps another look would be justified.

I would like to present a piece of evidence to show that I believe they know what they are talking about. On our campus they built a 10 story hospital. They knew enough about the properties of this soil to lay the first floor of the western half of this building ½-in. lower than the eastern half. They knew from using the equation that when they got the building up the eastern half would settle ½-in. more than the western half.

STRENGTHS RELATIONSHIP

In using the equation, then, there is a definite relationship between tensile, shear and green compression strengths. I have plotted each one using the given information, using the degree of ramming for every mix which was presented in this paper. Let me assure you that it has been a long time since we had this kind of excellent information in a paper.

If Coulomb's equation is right, and assuming that the soil mechanics people are correct, then if you will take mix no. 3, I will give you a sample of some of the data that will actually come out.

I am assuming here that the tensile values are satisfactory, and also the green compression strength values. We have three values here, and perhaps one of them (or all of them) is open to question. This is a matter of the test equipment, and the way we make this equipment. It has nothing to do with the kind of information the authors' presented.

Under the above cases, for 3 rams, the shear strength should be 6.6; for 7 rams, 10.6; for 10 rams, 10.5; for 12 rams, 12.2. This must be because the above equation gives us the information at the time of failure.

On the basis of this kind of information, if you have reliable green compression strength and green tensile test, you can immediately get what we know as the angle of internal friction.

Plotting that information we find this value varies slightly as the ramming goes along. They are much larger than that which is found ordinarily in soils compacted to the highest degree under tremendous weights. This means we ought to take another look at how our tests are made.

STRENGTHS RATIO

If, however, you plot these data on the basis of the equation we find this: if you take the ratio between shear and green compression strengths, the data from the paper show that the green compression strength varies on the average, and first of all for those mixtures containing 8 per cent western bentonite.

In the paper, the ratio is 3.5. If you take the same ratio using the equation it is 2.95. With a 12 per cent western bentonite, from the paper it is 3.06; with the equation it is 2.97. With the 10 per cent western bentonite it is 3.10, against 2.91. With the 6 per cent western bentonite mixtures data in the paper show the ratio 3.2, against 2.85 with the standard equation.

However, for the 3 rams on which we have most of the data, you will find the data in the paper have the average of 3.51, whereas if you use the standard equation and plot the results, and taking the value of shear from the equation, the ratio is 2.91. This is shown—that on the basis of the standard equation you will get the same ratio between shear and green compression strengths regardless of ramming and regardless of the type of bentonite or fireclay used.

The next thing I would like to point out is that the angle of internal friction went up to 60 degrees. I would venture to say that this is impossible on the basis of what the soil mechanics people know today. You do not get that high an angle of internal friction. It was found, however, according to these data, that this particular angle of internal friction did go on an average of 48 degrees on 3 rams, to well over 60 at 15, 17 and 19 rams.

There is a relationship, however, between the angle of internal friction and the moisture content of the sand mixture. It does not make any difference what you have got in it. At 15 rams we find it is a straight line relationship in this case—but only at 15 rams. With the other degrees of compaction, it has gone all over the map.

It is possible to use these data and the angle of internal friction in the foundry. You set a mold on the floor and take off the jacket, and it cracks. This can be calculated.

We can determine if this mold is going to crack or not, whether there are cores in it or not. You can use these data, this simple method which the soil mechanics people do, and find out why shell molds crack. The same kinds of data we use here can be simply applied to those kinds of things. It is just a matter of us taking over lock, stock and barrel well known information derived from the standard equation mentioned, which is a simple one.

R. W. HEINE: I wish to thank you for your discussion of the application and implications of our data in the direction of soil mechanics. I think that one must recognize that papers are written for different purposes. This particular paper was written for the purposes of presentation to the American Foundrymen's Society.

I really do not feel so strongly about the testing procedures as you do. I am quite interested in the symposium on clay minerals put out by the Bureau of Mines and Standards about a year and one-half ago. They, in one series of experiments, dug up samples of clay from the western deposits of bentonite, and preserved these clays in their as-mined state so they would not change in moisture content.

From these chunks, they cut standard samples for testing, and obtained from these tests the same identical strength range, as far as compression strength is concerned, that we obtained when we fully rammed high clay content sands using the compression testing equipment we use in the foundry. I feel that this indicates a good correlation between our work on the compression properties and the strength properties we may expect to obtain in clay bonded material.

G. DI SYLVESTRO:[14] In Table 2 you showed two mixtures which were similar with the exception of the bentonite. One was called low gel, 21 sec. Can you tell us why, and what was the difference?

R. W. HEINE: Viscosity.

G. DISYLVESTRO: How did you measure green tensile and shear strengths?

R. W. HEINE: There were two testing devices used which were mentioned in the paper. First, a modified piece of equipment from that mentioned on page 92 of the AFS SAND HANDBOOK, and recalibrated; and second, the latest model testing machine. They gave values within 10 per cent of each other.

Presiding — R. L. DOELMAN, Miller and Co., Chicago.
J. A. GITZEN, Delta Oil Products Co., Milwaukee.
Secretary — L. D. MARINELLI, Kensington Steel Co., Chicago.

Tempering Molding Sand, H. W. DIETERT, V. ROWELL, and A. L. GRAHAM, Harry W. Dietert Co., Detroit.

MEMBER: In what molds do you normally run creep deformation?

H. W. DIETERT: At the present time creep deformation finds its greatest use in sands for gray iron, both large and small castings.

For sands used for small castings a load of 3 lb is suitable while for medium to large castings, a load of 5 lb is desirable.

E. ZIRZOW:[16] What do you consider actual and what do you consider critical?

H. W. DIETERT: I assume you refer to moisture in the sand. Actual moisture variations that are frequently found in sand by manual tempering is in the order of ± 0.4 points. For high quality casting production, the moisture variation should not exceed ± 0.2 points.

Many sand laboratories have an error of ± 0.2 points due to the fact that it takes extreme care to determine the moisture percentage accurately. Many errors may creep in as detailed in our paper.

It would be well that AFS Green Properties Committee prepare detailed test procedures covering both a standard moisture test and a compensated moisture test.

Presiding — T. E. BARLOW, International Minerals and Chemical Corp., Skokie, Ill.
E. C. ZIRZOW, Werner G. Smith, Inc., Cleveland.
Secretary — R. F. DALTON, American Colloid Co., Skokie, Ill.

Timing of Expansion Scab Formation, J. E. HALLER, James B. Clow & Sons. Coshocton, Ohio.

R. DALTON:[17] You showed these expansion scabs. Do you have any experimental results to show how to get rid of them?

J. E. HALLER: No. Only that in the article it is stated that for this particular process there was too much moisture in the sand. There would not be too much moisture in that sand if we were going to skin dry it or dry it, but for green sand and that particular process, there is too much moisture.

R. DALTON: Can you get rid of the scab by reducing the moisture?

J. E. HALLER: Yes. We do.

MEMBER: Why do you get rid of the scab by reducing the moisture?

J. E. HALLER: I believe there has been a lot of work done that shows that the greater the amount of moisture, the greater amount of dried and hot strength, and the higher the hot strength. In other words, you can get enough strength from the sand by increasing moisture that you can make a mold and pour a casting, but you do not have sufficient amount of clay to combine with the water so you eliminate the expansion caused by making the mold strong enough for putting the water in. Perhaps that may not be clear, but that is how we feel about it.

MEMBER: Do you have any shear strength data on the sand?

J. E. HALLER: 3 rams 2 x 2 sample, 2.8.

MEMBER: What would the compression strength be?

J. E. HALLER: 10 lb.

J. SCHUMACHER:[18] What is the mold hardness at the 3 ram level?

J. E. HALLER: 89.

J. SCHUMACHER: The high mold hardness was 90?

J. E. HALLER: Yes. You, of course, know that the 90 is an average. In some places it was less, and in some places it was more.

J. SCHUMACHER: What was your clay content in the sand?

J. E. HALLER: 10 per cent.

J. SCHUMACHER: 10 per cent fireclay?

J. E. HALLER: No. Bentonite. A combination of western and southern bentonite.

R. W. HEINE:[19] That 7 per cent moisture content—was that gummy and stiff and rather hard to ram?

J. E. HALLER: No, normally the moisture for jolts is held around 5 per cent. Five per cent $\pm$ 0.2 or 0.3 per cent.

R. W. HEINE: Would this casting scab under that moisture?

J. E. HALLER: No. The difference is ramming without the rammer.

D. WILLIAMS:[15] I would like to defend the ideas put forth in this paper with, perhaps, a couple of examples. I learned of one of them last night. On the basis of the contention in the article that the author referred to, the foundries that I know of are fighting to get rid of one-half of one per cent scrap. The ideas presented here are sufficient enough that the head of the Chair of Foundry Production at the Leningrad Polytechnic Institute has asked me to submit another paper giving perhaps some different ideas than what is commonly accepted by the foundry industry. That paper has just arrived in Moscow. I know I am at a definite disadvantage in a group that is absolutely sold on expansion of silica, but all I would like to do is say that perhaps we can take another look at expansion-type defects.

R. DALTON:[17] If at normal moisture content it does not scab, under normal conditions, what would you suggest to foundrymen who do get scabs and they feel their moisture content is under control? In other words, let us not take what we call the abnormal condition that produce a common foundry defect.

J. E. HALLER: The question is that if these scabs are caused by higher moisture, what will the author suggest when the scab occurs when the moisture is at the proper level. My opinion is, not fortified by any tests, that if the sand is in the right moisture range, or is considered to be in the right moisture range, and there are a number of scabs, you should add clay. I would.

J. SCHUMACHER: Did not you also point out that in many cases the scabs you feel are due to slow pouring rate? I think that was brought out early in your paper. That would also by my opinion—that if the sand were under normal conditions of moisture, a slow pour could still cause a scab.

J. E. HALLER: That is right.

J. SCHUMACHER: Something that might be interesting is that in our laboratory one non-ferrous foundryman brought in a sand that he said would definitely scab. So a rammed specimen was heated up in a gas-fired furnace. What happens in 1 min in this type of furnace will happen in 30 sec in the mold. We had the foundryman change his feed system to choke, cut the pouring time, and the scabs disappeared.

R. W. HEINE: From looking at the pattern did you use a riser and a sprue?

J. E. HALLER: Just a sprue. On the photograph of the pattern you saw a button where the sprue would be cut to. When the thing is turned over and all made in the drag, the sprue goes down on top of that. That makes it the basin for the iron.

R. W. HEINE: Are you keeping pressure on the metal?

J. E. HALLER: All you have is the cope pipes, which is the 5 in. that was considered to be effective sprue height. We made the casting in the cope and we did not have enough feed metal so it solidified too quickly and we had those shrinks on the top step in the 2-in. section.

MEMBER: In the skin dried mold is it your opinion that some scabs could be caused by the water in the mold wash?

J. E. HALLER: No. I assume the mold is washed and dried. Now if you have scabs the only thing I would think of is that the mold was not dried.

MEMBER: Say it was dried, and the sand tempered and the mold rammed. Could you produce scabs and buckles with the mold wash, due to the excessive water from the mold wash?

J. E. HALLER: I do not think so. I would not understand it if you did. Are you referring to a blacking scab?

MEMBER: Not necessarily.

MEMBER: Do I understand that scabs are only caused by water? I do not think this should be the inference—that if you have high water you get scabs, and if you control the water level you get rid of them. The question I have is why the emphasis on water as the primary cause, when there are other possible causes.

V. ROWELL:[20] Just as a practical observation on high moisture and scabbing, almost invariably they go together—not as the direct cause, but if you put water in to make the sand plastic, green deformation becomes higher. Sometimes there seems to be a little confusion on the effect of the water on the hot deformation. It continuously has quite a bit to do with the ability of the sand to absorb its own expansion within original dimensions without mold wall rupture. The green deformation increases inversely proportional to the moisture. One of the first things I always like to do is to get the water content down; that will not go by itself usually, but usually in proportion to that.

T. BARLOW:[21] The gentleman has some basis for remarks in reference to core wash. There are differences in core washes. Some of them contain high bonds, and some do not. Some are soft and some are hard. I can conceive of putting a hard bonded core wash on a mold and a core, and the water would soak up into it ⅛-in., to crack out dried and leave quite a hard crush in the ⅛-in. the water penetrated. However, it is going to depend on the core wash used.

MEMBER: That would be something similar to what people who make large castings refer to as sweatback scabs. Where they skin dry a mold and then it is left long enough so that the moisture starts to move back into the face so there is a sweatback condition. This is quite common in heavy castings. I think the same situation occurs, or can even be emphasized by, possibly, an incorrect choice of blacking or of bonding and blacking.

Sand Movement and Compaction in Green Sand Molding, by T. J. BOSWORTH, R. W. HEINE, University of Wisconsin, Madison; J. J. PARKER, SPO, INC., Cleveland; and E. H. KING, J. S. SCHUMACHER, The Hill and Griffith Co., Cleveland.

T. SEATON:[22] Is it not true that the method of introduction of the sand into the flask has some influence. as far as the molding operations are concerned?

R. W. HEINE: Introducing the sand in the flask, as far as the molding operation is concerned, is the starting point of all successful molding. If you do not do that right, you will not get anywhere from that point on. If we get sand out of the hopper, and it falls 6 or 8 ft and it hits, and we feed back this to 75 lb/cu ft, and the molder shoves the mold over so it is filled up and you have 55 lb/cu ft, there is absolutely nothing that we do in conventional molding to rectify the situation. We apply a flat squeeze board and the inevitable things that will happen are that this will reach hardness and density before the other part will. The squeeze board will tilt slightly and you will have difficulty drawing, and of course you will not have as much hardness from the one place to the other. You may have a little ram-off because of this, so that proper filling of the flask will have to be the essential operation in order to make good molds.

T. SEATON: Would you care to make a suggestion as to the proper way to introduce the sand into the flask to rectify the situation.

R. W. HEINE: We are trying to find this out now.

T. BARLOW: As I remember the third dimension is 1 in., looking at a face that is in contact with the wall during molding. That particular dimension may be the smallest. In the illustrations where you have a pattern and a wall they are 6 in. apart. It is quite obvious that the pattern and the wall are restricting that wall of sand. Now you have two walls that are only 1 in. apart. I would think that the restriction would be even greater.

R. W. HEINE: You must realize that in using the squeeze plate, for example, that you have a squeeze plate that fits between two plates so you are in effect doing the same thing you do in using a stripping post on a specimen tube. If you get sand between that squeeze plate and the sidewalls, then you are in trouble. When you section that open there is a little curvature that is slight and it does not change the results at all. We would not want to base that just on observing this section; the point would be that this was followed by three-dimensional application of the principles. This is not included in this paper, but it in no ways changes the conclusions drawn.

MEMBER: Would you go into it a little further and tell us how you prepared your samples of sand? How were they put into the mold? I do not think it is quite clear just what was done.

R. W. HEINE: Tom Bosworth, would you like to answer that? Tom is one of the co-authors of the paper.

T. BOSWORTH: We weighed out the sand to the desired density and filled an ice cube tray arrangement, so to speak, to get the squares. We could weigh out these squares within ± 1 or 2 grams.

MEMBER: Do you ram the cubes? How did you ram them?

T. BOSWORTH: We dumped the sand in, and pressed it in with our fingers.

MEMBER: And this gives a uniformly rammed specimen? Those you were just talking about would not be so uniform in density.

R. W. HEINE: You weigh the sand. If it were not weighed you would not know how much sand was in there.

MEMBER: The squeeze combination and the squeeze jolt combination on the cope—I wonder what you have relative to the cope and drag—did the cope squeeze the jolt or the cope-drag? What I am refering to is—When you make a matchplate mold, you jolt the drag, put on your bottom board, and turn it over and then some people jolt the cope, whereas some people say never jolt the cope. They just squeeze the cope, whereas the drag is jolted and squeezed. Would there be a practical difference?

R. W. HEINE: Would there be a difference relative to a shift?

MEMBER: If your bottom board is not inside the flash would not the position of the drag remain relatively the same? With the conventional bottom board (fitting inside), would not the sand still be against the pattern?

R. W. HEINE: If there was a ram-off, it caused the mold to move sideways, and it might appear as a shift. As to jolting the drag, I think the amount of sand in the flask would have quite an effect.

MEMBER: You explained the differences in mold hardness when you squeeze or jolt and squeeze, so actually what you are doing is to hand squeeze the drag and jolt and squeeze the cope. You have two different mold hardnesses in the cope and drag, and if as in most cases the molder shovels his sand in or has it dropped in from a hopper, you get this distribution of the sand, and when he squeezes the cope that also adds a difference in mold hardness from one side of the mold to the other so in effect what he is getting is a compacted mold in the drag, and a swell on one side of the cope that goes up as a shift.

R. W. HEINE: That is right. In other words, due to mold wall movement the one side is softer than the other. If the mold is filled properly and we use the proper squeeze board, we would eliminate this.

MEMBER: What causes a crack in a mold after squeezing?

R. W. HEINE: A crack might not be associated with molding operations. You have to disassemble the flask used in testing to get at it, and when you do this once in a while it is unavoidable to flex it and you can start cracking it. Cracking in a mold would be most commonly due in ordinary molding conditions to 1) you frequently spread the pattern and flask when you apply the squeeze, and when you release the squeeze they spring back and you get either a change in the mold cavity formation or a crack in the mold, 2) An inadequately bonded sand that was overrammed could cause a crack, so that you have got spring back from the sand itself. And if the squeeze is not uniform, you get a lot of flexure and it is hard to predict what will happen.

MEMBER: What will a diaphragm type squeeze do?

R. W. HEINE: We have done quite a bit of work with this type of squeezing. There are different ways of squeezing with a diaphragm. You get different results from the different ways depending on the way you do it.

MEMBER: How do we utilize this information to the best advantage?

R. W. HEINE: This is another phase of this total study. The first part of the study is the principles of sand movement and compaction, as presented in the present paper. There are at least three different machines or mechanisms that I can think of that are under study now. And I do not believe I should comment on this any further than this. I hope you will accept this answer.

Steel Division

Presiding — A. J. KIESLER, General Electric Co., Schenectady, N.Y.
W. A. KOPPI, International Nickel Co., New York.
Secretary — R. L. LORD, Wehr Steel Co., Milwaukee.

Aid for the Design Engineer, foundrymen! "accentuate the positive," J. J. HENRY, Missouri Steel Castings Co., Joplin.

J. WARING:[23] How is it practical to use stress analysis techniques as a sales tool in view of the complexity of castings and the cost involved?

J. J. HENRY: Foundries must make information on the strength of various cast shapes available to the design engineer. We can start by obtaining data on stresses in simple shapes such as ells and tees and then progress to the complex shapes.

Brittle lacquer can be used to study castings to obtain basic information on the entire structure. This method has the advantage of providing qualitative data on a design at low cost. If more precise data are required, the more costly methods such as strain gages can be used.

W. K. BOCK:[24] We should give more consideration to the effect of defects such as shrinkage or tears on stress concentration. We cannot ignore test bar properties in our comparison with wrought products.

J. J. HENRY: Test bar properties should be recognized as required to qualify the metal, but not as being representative of the ability of the structures to carry loads. Test bars will not evaluate the properties of a structure in fatigue or torsion. There are no tensile stress concentrations in test bars. Position of the defects will affect the stress concentrations. Surface defects in highly stressed areas are, of course, serious. Internal defects or defects in unstressed areas can be relatively unimportant.

R. J. ERISMAN:[25] What is the effect of contraction stresses during solidification and cooling on the serviceability of the casting?

J. CAINE:[26] The heat treatment performed on all steel castings relieves these stresses completely.

L. W. SONGER:[27] Due to the complexity of many cast structures, stress concentrations cannot be calculated accurately. It is necessary to use strain gages in these cases in order to get the desired analysis. Generally, it is only necessary to prove new designs by this method, which is rather costly and time consuming. Test bars are not representative of the casting due to mass effects and variations in cooling rates.

J. WARING: When a complete, thorough and expensive stress analysis has been carried out on a particular casting design, there remains the problem of possible stress concentrations at internal defects if such exist in critical sections. Stress concentration factors of more than 100 to 1 have been reported at the tips of sharp notches such as might be present from hot tears, interdendritic shrinkage voids and the like. How are such difficulties overcome?

J. HENRY: Some defects such as cracks can lead to very high stress concentrations. However, internal defects are often harmless. We should obtain our customer requirements and take extra precautions in areas of high stress, then by using special inspection techniques, such as magnetic particle inspection and x-ray, we can assure the absence of harmful defects in critical areas.

Method of Obtaining Castability and Maximum Field Service from Cast Products, J. W. BECKHAM, Texas Foundries, Inc., Lufkin.

J. WARING:[23] In view of the difficulties and costs of carrying out stress analysis, the limited amount of data available from photoelastic and other methods, and the inability to carry out mathematical analysis for all but the simplest shapes and load types, to what extent does a stress-analysed casting retain an economic or cost advantage over an equivalent wrought product, especially for the short-run production or jobbing type of casting? Since Mr. Beckham has given us several case histories, can he perhaps give us rough figures of savings obtained versus the cost of the analyses carried out on the castings described?

J. BECKHAM: Wrought products are frequently superior to cast products when the foundry is working with a bad design, either for the foundry or the customer. Stress analysis can often point the ways to make a casting superior to the wrought product.

J. WARING: Can you quote figures showing savings obtained versus cost of analysis?

J. BECKHAM: The cost of stress analysis can be quite low, if brittle lacquers (stress coats) are used. Sometimes a customer will request and pay for stress analysis. Often it provides the means of satisfying a customer in trouble or of aiding a production run in the shop.

J. WARING: The cost of the analysis and of stimulating the service load must be quite high in many cases.

MEMBER: The economic advantage obtained by stress analysis is often considerable. However, even greater advantages are obtained by gaining customer confidence and good will. Future business depends on this. Also, a foundry which can demonstrate competence in casting design is more likely to be consulted early in the design stages of new castings.

J. HENRY: To sum up, stress analysis techniques are further tools for the foundry to use to the advantage of both itself and the customer. How much these techniques are used will depend on the economics of each individual study. Obviously, it is not possible to do a thorough stress analysis on a short run casting. For a few pieces, a simple brittle lacquer study can be all that is required and will be rather inexpensive. The more thorough studies using strain gages are generally used for high production runs where the expenditure of a considerable amount of money is economically justified.

Presiding — D. N. ROSENBLATT, American Foundry and Machine Co., Salt Lake City, Utah.
S. L. GERTSMAN, Dept. of Mines & Technical Surveys, Ottawa, Ont., Canada.
Secretary — E. W. O'BRIEN, Oklahoma Steel Castings Div., American Steel & Pump Corp., Tulsa, Okla.

Mn-V-Mo Age Hardening Austenitic Steel, foundry characteristics, N. C. HOWELLS and E. A. LANGE, U.S. Naval Rsch. Laboratory, Washington, D.C.

J. VARGA[28] *(written discussion):* The authors' comment on the feeding characteristics of the Mn-V-Mo austenitic steel indicates the necessity for specific feeding data for the various cast metals. During the progress of research at Battelle on the feeding characteristics of stainless and heat resistant steels, the feeding characteristics of side risers were studied.

A series of 1-in. thick plates were cast with top risers of sufficient size to produce completely sound castings. A second series of plates, of the same dimensions, were fed with side risers having the same dimensions. The shrinkage patterns of the castings are shown schematically in Fig. 1. The end zone was 2½T, as for class B steel, but the riser zone was reduced to approximately 1½T.

The loss in yield, caused by changing a riser from a top riser to a side riser, is shown in Fig. 2. The results of our research showed that a horizontal plate, 4 in. wide, was the

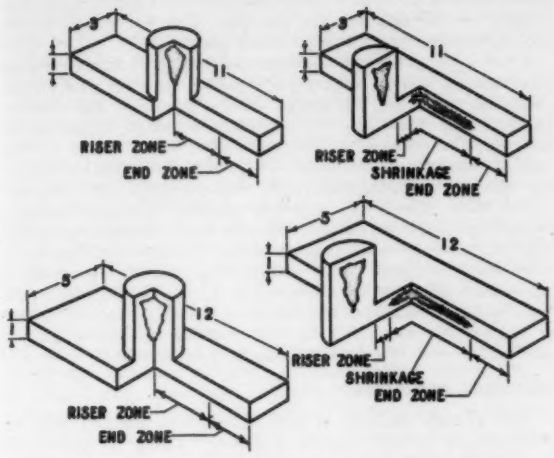


Fig. 1 — Comparison of feeding characteristics between top and side risers.

widest plate that could be made completely sound with a side riser, in austenitic steel. This coincides with the results reported by the authors for the Mn-V-Mo austenitic steel.

I would like to thank the alloy casting institute for permission to present this information which was developed during research sponsored by them.

AUTHORS' CLOSURE: The authors agree that specific riser size and feeding distance data is necessary for the calculation of riser dimensions and positions for various cast metals. Unfortunately this research did not permit an extensive investigation of the feeding characteristic of these new alloys, but the results of the limited tests conducted did show that the alloys are more difficult to feed than are the Class B steels.

The data presented in Fig. 1 and Fig. 2 of this discussion do point out that effective feeding distance is lost when a top riser is moved to a side position. Judging from the schematic sketch of the shrinkage cavities in Fig. 1 it would appear that the feed metal from the side risers was chocked off prematurely by the movement of the freezing front from the base of the riser. It would be interesting to determine the increase in effectiveness that could be obtained by adding at least a hemisphere to the bottom of the side risers.

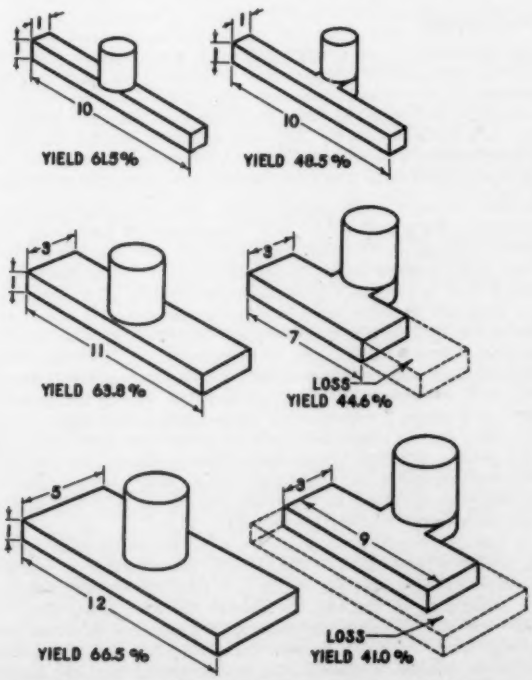


Fig. 2 — Yield of simple shaped castings as affected by the reduced feeding of side risers.

Presiding — R. L. LORD, Wehr Steel Co., Milwaukee.
C. A. FAIST, Burnside Steel Foundry Co., Chicago.
Secretary — C. K. DONOHO, American Cast Iron Pipe Co., Birmingham, Ala.

Application of Press Forged Castings, P. GOUWENS, T. WATMOUGH and J. BERRY, Armour Rsch. Foundation of Illinois Institute of Technology, Chicago.

R. G. BASSETT:[29] Has any work been done with rolled forgings?

AUTHORS' REPLY: None at Armour. Results similar to those reported for castings have been reported in the literature for wrought steel.

MR. LYDE:[30] Is shape important in affecting the pressures required?

AUTHORS' REPLY: Shape is important. An advantage of ausforming in relation to marforming is that the pressures are not as critical. Cold working a martensitic structure, although imparting higher mechanical properties, requires higher loads.

MR. KNOWLAND:[31] What tolerances can be expected?

AUTHORS' REPLY: The question of tolerances is complicated by springback. There is no springback when austenitic structures are pressed. Significant springback occurs when cold working martensitic structures. In general 0.010 in. tolerances can be held with the simple shapes studied.

Reclaim Sodium Silicate Bonded Sand, G. C. WARNECKE, National Malleable & Steel Castings Co., Melrose Park, Ill.

MR. BELUSKO:[32] What sand mixes were used for the dimensional tests? What stripping procedures were used for the CO_2 molds?

AUTHOR'S REPLY: A 4 per cent sodium silicate-kaolin clay sand for the CO_2 molds, a standard Western bentonite-cereal sand for the green sand molds. It was necessary to vibrate the patterns before the CO_2 molds could be drawn.

MR. EGGERT:[33] What were the pouring temperatures? What type of steel was used? What type of crushing and molding equipment was used? Would higher green sand mold hardnesses have decreased the dimensional variations of the green sand castings?

AUTHOR'S REPLY: All molds were poured at 2840-2845 F with grade B steel. An old single wheel muller was used to crush the sand. Standard jolt-squeeze machines were used to ram the green sand molds. Perhaps higher mold hardnesses would be effective in decreasing dimensional variations, but this point was not checked.

MR. KEPFER:[34] The use of a sugar addition to the sodium silicate binder should improve shakeout and also crushing for reclamation.

Rapid Hydrogen Determination for Steel Foundry Control, C. C. CARSON and B. J. ALPERIN, General Electric Co., Schenectady, N. Y.

MR. EGGERT:[33] What is the cost of the apparatus?

AUTHORS' REPLY: About $1000.

MR. BORRIS:[35] What pressure can be expected from the mechanical pump? How much CO is evolved? Can high alloys, cast iron, or slags be analysed?

AUTHORS' REPLY: A pressure of 0.1 micron is attained after some minutes of pumping using no diffusion pump. No crucibles, or boats are used. Over 80 per cent of the evolved gas is H_2, the rest is CO and N. Cast iron and slags will require special precautions due to the possibility of adsorbed water, if nothing else. Only carbon and low alloy constructional steels have been analysed to date.

J. BERRY:[36] The British Cast Iron Rsch. Assn. recommend the use of chilled, white iron as samples for extraction analyses. This overcomes the problem of graphite. High sulfur may complicate matters. Were any special sealing precautions taken? What is the blank rate?

AUTHORS' REPLY: All joints were sealed with grease and wax. The blank rate is negligible when things are working right. After the system has been pumped out for a day the blank rate is less than a micron of the 1500 ml system volume.

MEMBER: Was the time of extraction varied depending on chemical composition?

AUTHORS' REPLY: No, the same time is used for carbon and low alloy steels. No work has been done on high alloy steels.

MR. SIMS:[37] What is the order of magnitude of the correction for H diffusion through the quartz tube and how is the correction made?

AUTHORS' REPLY: The diffusion correction is in the order of magnitude of 1 micron to 32 microns and is proportional to the pressure. It is measured by introducing H at various pressures into the system and noting the drop in pressure for various temperatures and times.

Presiding — C. K. DONOHO, American Cast Iron Pipe Co., Birmingham, Ala.
E. W. O'BRIEN, Oklahoma Steel Castings Div., American Steel & Pump Corp., Tulsa, Okla.

Secretary — W. A. KOPPI, International Nickel Co., New York.

Steel Castings for the Aircraft Industry, quality requirements, Y. J. ELIZONDO, Chance-Vought Aircraft Co., Dallas, Texas.

J. RASSENFOSS[93] *(written discussion):* The manufacture of steel castings for aircraft is an important subject and it is gratifying that Mr. Elizondo has brought the subject to the attention of the foundry industry by writing this particular paper and presenting it before our Society. It is apparent that steel castings have presented both possibilities and problems to the aircraft industry. Apparently the problems now greatly outrank all other factors with the result that the promise of steel castings is not being fulfilled. Like most problems, it seems that there are two sides to this one. The following comments are offered to demonstrate this statement.

The author states that the foundry industry has not provided a set of "uniform and rigid specifications to govern the production of aircraft castings." It is true that the steel foundry industry has not produced a specific specification for aircraft castings. It has long been the desire of the industry to reduce the number of specification writing bodies and specifications to a minimum. To that end increasing emphasis has been given to the placement of this activity within the framework of the American Society for Testing Materials. Although there is no specification now in the A.S.T.M. that will cover the future requirements of the aircraft industry, the present specifications offer the opportunity for that to be done rather simply.

With this possibility available, the logical course of action would seem to be that the aircraft producers and the actual and potential aircraft parts suppliers could get together within the A.S.T.M. organization and develop materials purchase specifications which are suitable to both parties. This would enable full usage of present information and the addition to it of the new material which the aircraft industry desires to bring in.

MECHANICAL PROPERTIES CONTROL

Mention is made of the inability of foundries with whom the author is dealing to supply test bars which give consistent results. We regret as much as the author that these foundries were unable to do a satisfactory job. However, we cannot accept the statement that this is a proof that all units of the industry are incapable of controlling mechanical properties of steel. In the steel castings industry, as in all others, there is a wide diversity of ability when all of the members of industry are considered. Therefore, it behooves those who seek particular qualities in purchased material to find those business concerns which are most capable of doing this particular job best.

It is apparent that if the aircraft industry seeks a satisfactory experience with castings and desires to obtain the many benefits which they offer, they must seek out those foundries which have the necessary qualifications for the difficult work which they require.

It is possible to obtain satisfactory strength control of cast steel, and this can be done at high strength levels by some steel foundries. For instance, the author's company has marketed for several years now, a high strength cast steel at the level of 225,000 - 250,000 psi. We consistently manufacture this material in quantity, controlling the hardness and impact characteristics within the tolerances which the author apparently desires. We are confident that there are other foundries which can do as well.

Apparently the author has had considerable difficulty in obtaining castings according to the deliveries promised by the foundries involved. The ability to deliver on time varies considerably from one company to another whether it is the foundry industry or some other. It is not known from the author's presentation what type of work was promised for delivery, and whether or not this was the first run of a new design or was the production from an established part.

It has been our general experience that the aircraft industry is constantly seeking higher performance, and for that reason consistently changes the design of the part and the material requirements. This progressive attitude is indeed commendable; however, the constantly changing requirements makes it difficult for any materials supplier to advance his practices to keep up with the changing goals while maintaining a tight schedule.

It is most difficult for a person quoting on a part having a new design and a brand new materials requirement to accurately state the delivery which he can make. Perhaps both parties in this case have been guilty of assuming that this is strictly a production problem when, in reality, it is a development problem. Meeting a deadline for a solution of a development problem is considerably more difficult than meeting a schedule for producing an item of established design and material.

FOUNDRY REQUISITES

The aircraft industry is seeking the best product that the industry can produce and is constantly raising the aim so far as strength-weight ratio is concerned. The aircraft industry seeks not only high quality of production, but also a continuous development program. To be able to perform this, the qualified foundries must have the following attributes:

1) A technical staff of high quality and in good depth.
2) Excellent laboratory and pilot plant level equipment to carry out the necessary experimentation and analysis to solve problems such as are created by the new demands.
3) A trained and disciplined manufacturing organization to execute faithfully and in detail processes which the experimental and pilot level groups develop.
4) Adequate financial resources and a keen management group to be able to maintain the above facilities and personnel, and be able to cope with the difficulties of maintaining a sustained effort.

From the author's experience, and some of our observations, it would seem that although the aircraft industry has had a good experience with some steel castings suppliers it has had poor results from others. We feel that in those instances in which poor results were obtained, the organizations involved did not have all of the above attributes. It would seem that for the work which the aircraft industry desires, they should seek out only those firms which are so equipped.

CONCLUSION

In conclusion, it is granted that the steel castings industry may not have satisfied the wants of the aircraft industry to date. However, if the aircraft industry will carefully define its goals in realistic and tangible terms and seek out research organizations and foundries which have the qualifications for the type of development and production work necessary, we feel that it will be demonstrated that steel castings can meet the challenge.

AUTHOR'S CLOSURE: The author agrees that there are two sides to every problem and it was with the purpose of presenting the aircraft side to the foundry industry that the paper was prepared. With respect to the specifications, it is the author's contention that so long as a part produced under a given specification does not function satisfactorily for the design purpose intended the specification is inadequate regardless of whether the problem is low strength, lack of uniformity, etc. This has been the case for a large number of the aircraft castings produced to date. The suggestion that the specification writing be consolidated to a single group or activity is a step in the right direction. But, equally important will be recognition that there are many widely different design requirements, design philosophies and environmental requirements which will require various specifications in order to meet the casting users needs in the most economical manner in each case.

It was not intended to imply that the foundries with whom the author's company has dealt have been unable to control the mechanical properties of the test bars. Quite the contrary, the uniformity of the test bar results has been quite good; it is the inability to reproduce these same properties in production castings with the same or even similar uniformity that is the source of trouble. It should also be mentioned that the course of seeking out qualified foundries has been carried out to the maximum extent possible by the author's company with no

significant change in performance noted. If, as stated by Mr. Rassenfoss, steel castings with guaranteed properties in the castings of the levels indicated were produced, I am sure they would find a receptive market in the aircraft industry.

As for the schedule, or rather lack of schedule, the data presented were carefully selected to include only those castings for which shape, complexity and material were not new to the foundries involved. In fact, only those sources claiming prior production experience were utilized. Certainly where any development, new design or first run is involved consideration is given to this in establishing schedules. It must be recognized that the problems discussed by Mr. Rassenfoss are not unique to the foundry industry but are faced by all suppliers whether they deal with castings, forgings, rolled products, extruded shapes or any other. It is when the performance of the foundries is compared to that of the other suppliers that criticism might be leveled at the foundries.

The author agrees with Mr. Rassenfoss on the attributes necessary for a foundry in order to produce high quality castings. At the same time the author cannot help but observe that these attributes are the prerequisites for any well organized progressive manufactuing concern regardless of the product involved. These attributes are not anything new nor are they unique to a foundry that is to produce aircraft castings. Certainly the degree to which these attributes are possessed by any individual concern may have to change somewhat but the basic attribute should be present in all.

As for establishing realistic and tangible goals, this has been done as far as is possible when it is considered that most of the goals are dictated by our national defense needs. These in turn are not static, they are continually changing, therefore, all industries associated with support of this effort must adjust and learn to operate on these ground rules. It can be done and has been done by many companies and industries today. The author feels that steel castings can fill a need and perform a service for the aircraft and missile industry that no other fabrication process can accomplish economically, and it is with this goal in mind that the information and data in the author's paper were prepared.

Gray Iron Division

Presiding — S. G. JOHNSON, International Harvester Co., Indianapolis, Ind.
K. G. PRESSER, Buckeye Foundry Co., Cincinnati.
Secretary — M. H. HORTON, Deere & Co., Moline, Ill.

Continuous Carbon Injection, J. E. WILSON and R. C. SCHNAY, Canada Iron Foundaries, Ltd., Montreal, Que., Canada.

MR. MORRISON:[38] Was there a problem in getting rid of the carbon not in solution?

AUTHORS' REPLY: The authors did not have this problem, although it could happen if the feed rate was too high.

MR. HASTINGS:[39] Could this application of carbon injection be applied to irons of 3.30-3.50 per cent final total carbon level if the metal came from the cupola at 3.00 per cent? This would apply to reducing pig iron consumption.

AUTHORS' REPLY: A 3.00 per cent total carbon metal when injected should give a 100 per cent efficiency. High carbon iron is more difficult to obtain if total carbon level is 3.30-3.40 per cent at the start.

MEMBER: Has the sulfur content been a factor in regard to total carbon pickup?

AUTHORS' REPLY: It is felt sulfur is a factor, but no data was submitted.

MEMBER: What effect has impurities in the graphite used for injection?

AUTHORS' REPLY: Particle size and distribution has been found to be more important than impurities, although high sulfur content will prevent high total carbon pickup. A 90 per cent carbon content level is desired in the graphite.

Gray Cast Iron Machinability, quantitative measurements of pearlite and graphite effects, W. W. MOORE, Battelle Memorial Institute and J. O. LORD, Ohio State University.

E. A. LORIA[40] *(written discussion):* We were pleased to review another study of cast iron machinability utilizing the constant-pressure lathe test since we, in conjunction with Mr. Moore's predecessors, Messrs, F. W. Boulger and H. L. Shaw, originated this way of measuring the machinability of cast iron.[1,2,3] The present paper can be considered as a sequel to our paper on the multiple correlation of the constituents affecting machinability of cast iron.[2]

Since the test was introduced there have been some who have objected to it as a means of analyzing machinability. They maintain that the equipment would be helpful in providing direction in experimentation but should not be used for estimating tool life where abrasive wear is an important factor. We would appreciate the authors' comments on this opinion, realizing that the constant-pressure lathe reveals the frictional characteristics of the metal in a cutting operation.

As we all know, machining is a complex operation and consequently the writer, while at Mellon Institute and Carborundum Co. from 1948 to 1954, conducted a number of machinability testing procedures which led to the conclusion that a combination of such test methods was needed to provide a complete evaluation of the machinability of a particular composition. These included lathe turning tests on rings and bars measuring tool wear,[4,5] constant-pressure lathe turning tests on bars,[1,2,3,6] drill penetration tests under constant load and speed on bars and castings[7] and lathe turning tests on rings measuring cutting temperature and force.[8]

Although some of these are considerably quicker and cheaper than tool wear measurements, the net effect of all the complex factors which determine the rate of wear on the tool is obtained in actual tool life tests conducted under practical conditions. Tool life curves thus obtained furnish data from which commercial speeds and feeds can be selected.

CAST IRON MACHINING

In the machining of cast iron the writer has shown that the cutting characteristics are related to the micro-structure of the iron rather than to the criteria of hardness and strength,[9] and that the casting skin and subsurface are potent factors in producing tool wear.[10] Consequently, we have favored the tool wear test on the as-cast surface of rings.[3] We would ask if Battelle, by this time, has not been successful in casting geometric rounds so that the constant-pressure test could be made on the as-cast surface rather than on a pre-machined surface?

The importance of rims and substructure produced by various casting factors cannot be denied. In fact, our last contribution[11] was specifically concerned with the casting factors affecting machinability of cast iron, as determined by our ring test. We divided them into two broad classifications: chemical factors which are related to melt composition, and physical factors which include thermal and "heredity" effects.

In general, the influence of chemical factors on machinability has been explained chiefly on the basis of the tendency of each chemical element in the melt to acclerate or inhibit the formation of graphite during solidification. The influence of most physical factors is still relatively unexplained. Some of these which we considered in our paper[11] include pouring temperature, rate of cooling, mold conditions and inoculation practice.

The constant-pressure lathe test for measuring the machinability of free-cutting steels was introduced in 1949. Have the authors been successful in speeding up the testing procedure for cast iron in view of today's trend towards higher machining speeds? Generally, tool wear breaks down into at least two distinctive types 1) that is predominantly temperature dependent, and 2) in a region of lower temperatures where the wear rate appears to be influenced primarily by unit pressures and distinctive properties of the work microstructures.

In other words, we would like to see the test adapted so as to provide information at cutting rates generating higher temperatures on either the as-cast or pre-machined bar surface.

SCATTER

A wide degree of scatter is evident in the variation of machinability with relative size of graphite flakes in the irons studied. It is realized that the relative volume and size of the graphite flakes are potent factors in controlling the machinability of the gray iron, but we wonder if the relationship with graphitic carbon and combined carbon would not have reduced the width of the scatter band shown in Figs. 2 and 3? In noting the relationship of machinability to carbon equivalent in Fig. 5, we would inquire if a multiple correlation analysis of the individual effects of total carbon and then silicon has been made, as was done in our paper?[2]

In Fig. 9 it can be seen that inoculation apparently did not improve the machinability index in the 165-175 range for the

test bar section utilized. If tests could have been made on concentric as-cast bars a beneficial effect might have been detected. Also, there is a possibility that it could be revealed in casting smaller test bars, say 0.75 in. diameter, and making the test in the usual manner.

The authors' results on the phosphorus content effects parallels our own work.[1,7] Would they comment on the amount of steadite in their 0.30 per cent phosphorus irons? We have noted that steadite in amounts of 3 to 5 per cent has no appreciable effect on tool life. No doubt greater amounts of steadite do have an effect and the result is tied in directly with the overall metal chemistry and casting section size.

For the usual constant-pressure test conditions, 1.2 in. diameter bar and pre-machined surface, we obtained results on irons averaging 0.50 per cent phosphorus (5 per cent steadite by volume) that were equivalent to low phosphorus irons. One must draw the broad conclusion from the work done over the years on phosphorus that the investigators are about equally divided in their opinions so the chances are that it is not particularly detrimental to machinability, per se, at least as far as carbide tools are concerned. One would have to reserve judgment for the broad area where high speed tools are employed.

REFERENCES

1. E. A. Loria, "Machinability of Gray Iron by Constant Pressure Lathe Tests," *Foundry, vol. 80,* pp. 194-200, Aug. 1952.
2. E. A. Loria, F. W. Boulger and H. L. Shaw, "What Constituents Affect Machinability of Gray Irons?" *American Machinist, vol. 96,* pp. 122-124, Oct. 1952.
3. E. A. Loria, "How Machinable are the Cast Irons?" *Iron Age, vol. 168,* pp. 158-161, Nov. 1953.
4. E. A. Loria, "New Methods for Testing Machinability of Gray Iron," *Foundry, vol. 80,* pp. 174-178, March 1952.
5. E. A. Loria, "Machinability of Gray Iron by Constant Pressure Lathe Tests," *Foundry, vol. 80,* pp. 194-200, Aug. 1952.
6. E. A. Loria, "How Melting Practice Affects Machinability of Malleable Iron," *Iron Age, vol. 169,* pp. 168-170. Sept. 1954.
7. E. A. Loria, "Machinability Testing of High Phosphorus Gray Iron," *Western Machinery and Steel World, vol. 46,* pp. 76-81, June 1955.
8. E. A. Loria and D. R. Walker, "Machinability Testing – Relation Between Cutting Temperature and Tool Life for Gray Iron," AFS TRANSACTIONS, *vol. 63,* pp. 261-266, 1955.
9. E. A. Loria, "Machinability and Microstructure of Cast Irons." AFS TRANSACTIONS, *vol. 62,* pp. 163-175, 1954.
10. E. A. Loria and R. L. Keller, "A Machinability Test for Gray Iron Castings," *Iron Age, vol. 165,* pp. 73-76, June 1950.
11. E. A. Loria, "Casting Factors Affecting Machinability of Gray Iron," *Western Machinery and Steel World, vol. 47,* pp. 96-100, Oct. 1956.

MR. ZUPPAN:[41] It was noted the chemistry was of secondary importance and, if more work had been done on using chemistry results as a measure, the results of the machining tests may have shown different results.

AUTHORS' REPLY: Agreed. There have been many papers written on machinability and all articles should be brought together. We were primarily interested in the microstructure as related to machinability.

Significance of Reported Chemical Analysis of Cast Iron, D. E. KRAUSE, Gray Iron Research Institute, Columbus, Ohio.

MEMBER: Are these samples that have been checked available?

D. E. KRAUSE: We may have some available, but we usually do not have enough left. Prepared samples definitely would not be available.

Presiding – J. S. VANICK, International Nickel Co., New York.
B. F. SHEPHERD, Ingersoll-Rand Co., Phillipsburg, N. J.

Secretary – D. KUHN, Chapman Valve & Mfg. Co., Indian Orchard, Mass.

Foundries can Produce their own Cast Iron directly from Ore, H. W. LOWNIE, JR., and A. J. STONE, Battelle Memorial Institute, Columbus, Ohio.

A. H. DIERKER:[42] In listening to dry statistical presentations of papers at technical meetings, I have often wished the authors would develop a little enthusiasm for their subject. In this respect the present paper is an exception. It certainly is a readable, enthusiastic presentation. I am afraid, however, that in this case the enthusiasm may have gotten a bit out of hand.

The mining of iron ore in the upper lake regions, its transportation to lower lake ports, its reduction to pig iron in high capacity blast furnaces is one of the outstandingly efficient industrial operations of the world today. The commercially successful conversion of low grade, abundant taconite into a high grade blast furnace feed that further improves the efficiency of these units and insures a continuing supply of domestic ore is one of the top technical and industrial achievements of our times.

To picture all this, as the authors do in Fig. 1, as a combination Toonerville Trolley and Rube Goldberg creation appears unfair, to say the least, to the industries who have invested hundreds of millions of dollars in these facilities. Actually, if any of the proposed direct reduction processes are carried to a successful commercial basis in areas of large foundry operations, it will probably be because of access to low cost raw materials available only because overhead and development expenses are largely born by the blast furnace industry.

Projecting commercial cost figures for processes still in development stage is difficult at best, and I will not attempt to evaluate the data given. The authors base their figures on continuous operations, but wisely point out the difficulty of integrating the continuous production of the reduction unit into the intermittent demands of the foundry. However, the cost of the steps that must be taken to handle this situation is not reflected in the comparative cost figures shown in Table 4.

Because of the nature of power contracts and other factors the foundryman would probably find, as operators of electric smelting furnaces have found, that it is impractical to operate except on a continuous basis. He then will be faced with the cost of pigging, stockpiling and disposing of cold pigs. In other words, he will find himself in the pig iron business. Before getting this far he would be wise to sit down and have a heart to heart talk with someone already in that business.

As the authors show clearly, production costs go down as the size of the reduction unit increases. It is entirely possible that a large, centrally located reduction unit, distributing cold pigs to foundries equipped with simple melting units may prove more economical, all factors considered, than for each foundry to add relatively complex reduction operations to its existing problems. The early foundry industry was integrated; producing castings direct from blast furnace metal, but experience taught the economic advisability of separating the two operations.

With the rapid increase in the cost of metallurgical coke relative to the cost of electric power, there is little doubt but that the latter will play an increasingly important role in iron smelting. The authors have described the more important of the direct reduction processes now in various stages of development. It will be interesting to see which of these, if any, will be first to challenge the blast furnace on its home ground.

E. C. MATHIS:[43] It is interesting to note the similarity between the picture that Mr. Lownie and Mr. Stone forecast in their paper and the iron foundry situation in the early American colonies. At that time, an iron founder established his operation on top of an iron ore deposit in the middle of virgin timber land whence he obtained and made his own charcoal fuel. He erected a stone, vertical-shaft, open-top blast furnace and poured castings directly from the iron produced in it. That was also a "direct reduction" process, as every blast furnace today employs, and which has been employed ever since prehistoric man first created metallic iron from iron ore. The fundamental reactions are old. The novelty in the processes described, lies in the equipment in which reduction of iron ore is carried out.

I have no first-hand knowledge of the efficiency, economics or product quality of those American pilot plants now in operation, but based on general knowledge of the factors involved, I believe that false hopes have been aroused in the minds of most foundrymen who have read the paper.

The processes fall into two general types based on the reducing agent, namely, hydrogen under pressure for one type, and the other type which makes use of carbon and/or carbon monoxide. Many of the pilot plants now operating use the former method, while it is the second type which is described in the paper. Perhaps a foundry melting 500 to 2000 tons of iron per day around the clock and around the calendar might find the

processes adaptable and desirable. The variety of problems which can arise become apparent upon careful study of the factors involved, some of which are pointed out in the paper. Not the least of them, is the ability to control the metal composition and quality to meet the exacting requirements of today's castings and also to produce, from the same system, the several classes of iron which most foundries are called on to pour during a day's operation.

Those same conditions of abnormally high production periods which Mr. Lownie and Mr. Stone point out lead to shortages of pig iron, foundry coke and scrap also lead to shortages of iron ore, coal and electric power, which would put the foundryman, using the new process, right back in the same dilemma as before. In fact, his situation would be worse because as everyone knows, our high grade domestic ores are becoming exhausted. We are becoming more dependent on high cost beneficiation of the low grade domestic ores which are left, such as the taconite pellet process and imported ore from Canada and from across the open seas. Steel making operations are continually growing and require more and more ore tonnage.

ADJUSTMENTS TO MAKE

It is also difficult to conceive of making alloy and carbon adjustments in an electric furnace through a thick layer of slag on top of the iron, equivalent to 800 lb of slag ton of iron produced. Remember, this is a continuous process, not a batch-type. Iron temperature is not as high from the submerged arc electric furnace used in these processes as from an ordinary direct-arc melting operation. Control of carbon would be difficult since there must be present surplus carbon to reduce the iron oxide present. It appears that a second electric furnace will prove to be essential for the purpose of adjustment of composition, for raising iron temperature suitable for pouring and to remelt gates, sprue, pigged iron. A second furnace would materially increase capital costs and operating costs.

If, in a foundry, molds are poured only 8 or 16 hr per day, and the reduction plant is operated 24 hr, I would not like to have to pour the iron held even in a gas-heated mixer ladle for 8 or 16 hr. A large electric holding furnace would be needed to maintain suitable temperature and prevent oxidation of the metal. The process is an intricate, involved metallurgical engineering system with large quantities of material in process at one time. Not only is a huge capital investment required, but also unique talents in the supervisory and operating staff.

Admittedly, the current costs of erecting a new blast furnace, including coke ovens, recovery plants for the coal chemicals and gas-consuming facilities are so high (higher than the 40 million dollars indicated) that new furnaces can scarcely be justified at current product prices. For this reason, pig iron and steel producers are carefully studying the new processes for possible incorporation in their own expansion programs. So far, the experience has not brought forth appropriations for full scale plants.

When these new adaptations of iron ore reduction processes are fully evaluated, should they prove more economical than a blast furnace, I foresee that they will be utilized at central producing points by independent operators not integrated with the casting producer, the same as all primary metal production, non-ferrous as well, is presently set up. Because of lower capital investment in such plants than in blast furnace plants, they could be located at certain strategic locations and perhaps could be arranged to deliver hot metal to the foundry.

Although I question the costs quoted for the new processes, I am in no position to contest them, except to say that in existing blast furnace plants, conversion from ore to metallics is carried out by the most efficient and lowest cost methods now available and proved.

Whether or not these new processes will become adaptable for foundry uses will depend on the data obtained from the pilot plants now in existence or contemplated, concerning the economics and technical feasibility. Time will tell.

AUTHORS' CLOSURE: Messers. Dierker and Mathis each show an awareness of some of the problems that must be solved before direct reduction of iron ore can be integrated into an iron foundry. The authors agree in general with most of the specific points raised by Messers. Dierker and Mathis. These problems (and many others) were familiar to the authors when the paper was planned and written. If foundrymen have been made aware of some of the potentials of direct reduction, then the purpose of the paper has been realized.

The authors disagree, however, with implications that direct-reduction processes have no chance for successful integration into some foundries.

Mr. Dierker and Mr. Mathis each comment on the problem of matching a 24-hr-per-day smelting unit to a foundry's 8-hr. or 16-hr demand for metal. This matching will indeed require some major adjustment of foundry thinking and scheduling. However, if such adjustment will result in a lowered cost of metal for the foundry, the authors believe that those foundries which will survive the present competitive squeeze best will be those which have the ability to make such major adjustments.

In rebuttal to several points in the fifth paragraph of Mr. Mathis' discussion, the authors wish to state 1) the metal can readily be tapped from an electric smelting furnace at temperatures higher than from a cupola, 2) an "excess" of carbon is not needed, and carbon contents of the product have been controlled to at least as low as 1 per cent, and 3) some foundry returns are expected to make a reasonable addition to the charge in the smelting furnace.

The authors do not see the direct-reduction processes as competitors for the blast furnace, nor do they expect direct-reduction processes to replace blast furnaces. The authors contend only that for some situations direct-reduction processes may make fine supplements to the blast-furnace production in this country so far as foundries are concerned. In the six months since the paper was written, continuing evidence has been forthcoming to show a swing by foundries away from the use of pig iron wherever possible. Pig iron has become expensive relative to other metallics used by foundries.

The authors are grateful to the acknowledgment in the closing sentences by each Mr. Mathis and Mr. Dierker that more time and more development may bring about the change described by the authors, even though the change does at this time seem quite shocking when it is first considered. As Mr. Mathis so succinctly puts it "time will tell."

Presiding — J. VARGA, JR., Battelle Memorial Institute, Columbus, Ohio.
W. J. SOMMER, The Plainville Casting Co., Plainville, Conn.

Secretary — R. W. FREEMAN, Union Carbide Metals Co., Niagara Falls, N. Y.

Factors Influencing the Soundness of Gray Iron Castings, I. C. H. HUGHES, K. E. L. NICHOLAS, A. G. FULLER and T. J. SZAJDA, British Cast Iron Rsch. Assn., Birmingham, England.

C. M. ADAMS, JR.,[44] M. C. FLEMINGS, JR. and H. F. TAYLOR[8] *(written discussion):* The authors have made a concise, comprehensive, and interpretive review of the extensive research findings which have been reported in England since 1950 on the relationships among solidification patterns, dimensional changes and shrinkage porosity in graphitic cast iron. The oversigned are in complete agreement with the general practical conclusions of the report:

1) That mold plasticity contributes to expansion and unsoundness.
2) That directional solidification as obtains in the freezing of steel, is substantially nonexistent in cast iron and that risering (and indeed reasoning) based thereon can do more harm than good.

However, the essential problem of improving the soundness of gray iron castings dependably in production, requires close scrutiny of underlying causes and effects pertaining to the many observations reported by the authors.

For example, quoting from the abstract, "casting expansion is promoted by mold cavity volume changes . . . ," and on page 154 "because of the relatively long freezing range . . . the casting is likely to be sufficiently plastic to follow any variation in mold cavity volume . . . ," the implication is that yielding of the mold causes the casting to expand whereas, in fact, a rigid mold is found necessary to completely inhibit casting expansion; the mechanics of the process suggest it is the "plastic" outer layers of the solidified casting which exert the "push" required for the consequent expansion. It is not sufficient to observe that dilated castings exhibit more internal porosity; this is the necessary result of a materials balance, and does not explain the relationships between casting geometry and shrinkage in

green sand or the unusual functions of risers and gates on green sand castings.

GRAY IRON RESEARCH

In the course of an extensive research program on solidification of gray irons in the M.I.T. Foundry Laboratory,[1] it was demonstrated the expansion of a gray iron casting in a plastic mold is the result of a substantial outward push, exerted late in the solidification process. One of the means taken to demonstrate this fact was to cast identical, large, solid cylinders of manganese bronze, lead, steel and gray and ductile iron in the same molding material. The lead was chosen for its density, the steel for its high pouring temperature and the manganese bronze because its temperature of pouring and mush mode of solidification resembled those of cast iron.

Only the castings of gray and ductile irons showed mold wall movement. This means that considerable positive pressure prevails in the metal near the mold surface. The authors, as well as the oversigned, have taken the position that dilation resulting from graphite expansion causes a void to form in the iron castings, but questions which the authors leave unanswered are: How can the expanding pressure of the casting be so localized and anisotropic? Why does this expanding pressure deform the mold instead of forcing liquid metal in the direction of the low pressure void?

It is the opinion of the oversigned that dilation transpires at a stage late enough in the solidification process that the casting cannot function as a hydraulic system. In other words, it appears that much of the graphite expansion which causes mold dilation is expansion near the surface of the casting, and is exerted upon an essentially solid austenite matrix.

MOLD VARIABLES

Depending on casting geometry, the action of mold variables and of atmospheric pressure, the expanding walls of a casting may either buckle outwards and accentuate shrinkage or inwards and minimize shrinkage. Most casting geometries favor outward buckling, but it has been conclusively shown that if certain types of gray iron castings are risered judiciously they can be made to buckle inwards (slightly) and hence demand little or no feed metal. In these cases, it is essential to use risers which are small enough to freeze before dilation begins.

The casting then becomes a closed system and atmospheric pressure acting uniformly over the entire casting surface is an important factor helping to prevent outward dilation.[1] Although not understood in full detail, it now appears that, viewing the geometry of gray iron castings in light of this buckling mechanism, gates and small risers can be located judiciously to produce sound hypoeutectic gray iron castings, even in green sand molds.

With respect to effects of nucleation, inoculation, eutectic cell size and structure and the detailed pattern of solidification, it is felt these bear on the problem only as they influence the relative degree with which graphite expansion is exerted upon liquid versus solid surroundings. Treatments, such as magnesium inoculation, which momentarily suppress precipitation of graphite at the eutectic temperature (i.e., promote some supercooling), result in greater dilation and greater shrinkage because the graphite expansion occurs too late to make its influence felt on the interdendritic flow of liquid metal. Additions which inhibit supercooling, bring the graphite expansion to bear when more liquid is present, thereby promoting self-feeding.

An extreme example of a metal in which the graphite expansion is exerted against a nearly solid matrix is ductile cast iron; hence the extreme dilation and internal porosity frequently encountered in green sand castings of this material. The oversigned quite agree that the true shrinkage of ductile iron differs little from that of a flake graphite iron of similar composition and, in completely rigid molds using the same gating and risering practice, feeding requirements and soundness will be substantially the same.

PHOSPHORUS EFFECT

Finally, it should be observed phosphorus is troublesome on two counts. Not only does the low melting phosphide eutectic promote the interdendritic unsoundness clearly emphasized by the authors, but phosphorus also has an unexplained effect on dilation characteristics. Other factors being equal, low sulfur, low oxygen irons exhibit more dilation the higher the phosphorus concentration. This effect can be spectacular in the case of high phosphorus ductile iron, in which a casting made from metal containing about 1 per cent phosphorus will cause the upper surface of sand in the mold to heave upwards 1 in. or more and buckle markedly.[2]

In conclusion, the authors are to be commended for a paper which reflects clearly the magnitude and high quality of research being conducted through the agency of the British Cast Iron Rsch. Assn.; future contributions are awaited with much interest.

REFERENCES

1. Adams, C. M., Jr., Flemings, M. C., Jr. and Taylor, H. F., "Solidification and Risering of Grey Iron Castings," *The British Foundryman, vol. 51,* Dec., 1958, pp. 625-636; Also AFS TRANSACTIONS, 1958.
2. Results of Unpublished Research at Massachusetts Institute of Technology Foundry.

AUTHORS' CLOSURE: We feel encouraged by the interest which has been shown by the contributors to the discussion, and we are particularly grateful for the sympathetic and constructive written remarks of the M.I.T. investigators.

We agree that the expansive force exerted by the casting on the mold plays a major part in the mechanism of porosity formation, but we still believe that the size of the mold cavity may change as a result of heating and that this change plays some part in the process. Hendry and Wrightson[A] found that a variety of nonferrous alloys expanded during solidification in green sand molds (less so in dry sand molds). Although various explanations have been postulated we are inclined to the view that mold cavity expansion played some part in this work. The fact that at M.I.T. a similar effect was not found in other nonferrous materials and in steel is interesting and suggests that further investigation of the phenomena is merited.

Vaschenko and Sofroni[B] have also argued that in nodular irons casting expansion occurs at a late stage when the casting cannot act as a hydraulic system. However, both their results and those of Gittus seem to show that casting expansion begins at a time not very different from that of the beginning of eutectic solidification when much liquid is present throughout the casting. Furthermore, the experiments reported by the authors with rigid molds have shown that provided the casting is sufficiently contained the last liquid does in fact behave as though it were continuous throughout the casting. The only alternative explanation would require plastic flow of solid metal to maintain soundness in rigid molds.

As is rightly pointed out, the presence of quite small amounts of phosphorus, involving a dispersed low melting point liquid at a late stage of solidification, causes a remarkable increase in external swelling and internal porosity in nodular irons cast in green sand molds. Again this can be largely offset by casting into dry sand molds, indicating that the last liquid is in fact a continuous network.

Reiter,[C] in attempting to explain why nonferrous alloys expanded when cast into a green sand mold, suggested that when the last liquid wets the solid being formed the solid crystals will tend to be forced apart by the continuous inflow of liquid until this liquid is exhausted. The authors find this proposition attractive since it provides the following solutions to some of the most important observations which have been made:

a) The force tending to expand a casting is derived from surface tension at the interface of the solid and last liquid. This force is finite and may be counterbalanced by a sufficiently rigid mold to contain the casting.
b) The introduction of a low melting point wetting constituent is likely to increase casting expansion. Phosphorus in gray iron has this effect.
c) Increasing the surface area of the solid formed will increase the total force derived from surface tension. Thus inoculation, which increases eutectic cell number, increases casting expansion and unsoundness.

The effects of risers of different dimensions are also partly explicable in terms of the surface tension theory. A larger riser will provide more feed liquid which will, as shown by Pellini,[D] increase the casting expansion. A small feeder, starving the casting of feed metal, will reduce or even reverse casting wall movement, though it may not always eliminate internal porosity in thicker castings. However, manufacturers of thin-sectioned or small gray iron articles generally make full use of this fact and pursue it to its logical conclusion. Great quantities of

absolutely sound light-sectioned castings are made in gray iron with over 1 per cent of phosphorus using carefully designed gating systems and no risers whatever.

REFERENCES

A. Hendry, W. B. and Wrightson, S. N., Some Expansion Phenomena Associated with the Freezing of Non-Ferrous Alloys, *British Foundryman, vol. 51,* Feb., pp. 82-90, 1958.

B. Vaschenko, K. I. and Sofroni, L., Contribution to the Study of the Graphitisation Process in Nodular Cast Iron, *Revue de Metallurgie* (Roumania) *vol. 2,* pp. 51-81, 1957.

C. Reiter, H., Discussion of Paper by Hendry and Wrightson, *British Foundryman, vol. 51,* Nov., p. 579, 1958.

D. Dunphy, R. P., Ackerlind, C. G. and Pellini, W. S., Solidification feeding characteristics of Gray and Nodular Irons, *Foundry, vol. 82,* June, pp. 108-115 et seq, 1954.

R. A. CLARK:[45] In regard to the effect of inoculants on the soundness of castings, it appears that most of this work was done with high-carbon equivalent cast irons. Has the author any remarks regarding low-carbon equivalent cast irons?

I. C. H. HUGHES: Carbon equivalents of 2.5 to 4.5 were studied, which is rather a large range. The effect of inoculants holds for most irons.

R. A. CLARK: It is felt that inoculants are quite useful in low-carbon equivalent cast irons. Is it not possible that they may help in regard to porosity?

I. C. H. HUGHES: I would agree. The effect of inoculants on low-carbon equivalent cast iron is to remove primary carbides. It is well known that the primary carbides can cause enormous shrinkage; therefore, it may be said that inoculants would cut shrinkage.

L. A. GROTTO:[46] Have the authors noted any co-relation between the number of eutectic cells and the physical strengths of the irons?

I. C. H. HUGHES: Yes, in many cases the more eutectic cells exhibited by the iron, the greater the strength. This may be related to the fact that there are more grain boundaries in those irons containing the greater number of eutectic cells.

MR. MALONEY:[47] Is there a greater volume of graphite in ductile cast iron than in gray cast iron?

I. C. H. HUGHES: No. The actual volume of graphite in either iron, all things being equal, would be the same; the important fact is the distribution of the graphite.

MR. MALONEY: Have the authors noted that an increase in the eutectic cell number shows an increase in the ferrite present in the as-cast condition?

I. C. H. HUGHES: For gray cast iron it has been noted that the finer the graphite the more ferrite will usually be present. An increase in eutectic cell number does not necessarily refine the graphite in gray iron, but it does in ductile cast iron.

F. SEFING:[48] The authors are to be congratulated for an excellent presentation; however, some of the practical aspects have not been spelled out clearly. For example, in the electric furnace melting of cast iron, it has been found that longer holding time in the furnace increases the eutectic cell size. When a heat is quickly poured after melting, from the electric furnace, more shrinkage is exhibited than if the heat is held for a greater length of time in the furnace. In spite of the fact there are fewer eutectic cells or larger eutectic cells exhibited in irons with excessively long holding time, shrinkage still exists. It would be important if an explanation could be found for shrinkage from iron held excessively long in the furnace.

R. W. RUDDLE:[49] The authors have referred several times to the force exerted by metal in the mold causing actual deformation of the mold, and, consequently, porosity. Could this force be explained possibly by the growing of the graphite plates in the liquid between the eutectic cells?

I. C. H. HUGHES: I could not prove or disprove this theory. However, different push is exhibited by metals of different cell number counts. The explanation for the large force is not clear. The view currently held is that the force to expand the casting is partly supplied by the surface tension of the last liquid to solidify.

R. W. RUDDLE: I feel that there is no room for expansion of graphite in this liquid. Aluminum alloys work differently; for example, the finer the grain size, the less stresses or less mold deformation.

J. CROFT:[50] I might suggest that the austenite surrounding the nodules in ductile cast iron inhibits the expansion of the nodules.

J. S. VANICK:[51] The practical aspects of such information might be difficult to anticipate; for example, it would be difficult to anticipate the cell size ahead of time; we therefore must control the cell size by the use of inoculation and take care of any tendency for increased shrinkage by the use and proper locations of risers.

I. C. H. HUGHES: I agree that for full control it will be necessary to be able to decrease eutectic cell number when required as well as to increase it when required, and work is presently being done along these lines.

Copper in Cast Iron, principal considerations, A. DESY, University of Ghent, Ghent, Belgium.

MEMBER: The author used the term "pearlite promoter" several times during the presentation. Would he please elaborate on the meaning of this term?

A. DESY: Direct precipitation of ferrite from the austenite in the eutectoid temperature range is much more important and much more probable than ferrite formation through breakdown of the Fe_3C, into ferrite graphite. Copper is the pearlite promoter, that is, at the eutectoid range. It is an anti-graphitizer at this critical level. It is interesting to note that at the liquidus temperature, copper functions as a graphitizer.

C. LOPER:[52] The author refers to pregrouping of atoms in the liquid metal. Is this experimentally proved?

A. DESY: Yes, experimental proof exists and published work exists on low-melting metals; however, not on iron-base alloys, at least to my knowledge.

Die Casting Division

Presiding — H. E. ERIKSEN, Chrysler Casting Plant, Kokomo, Ind.
R. C. CORNELL, Litemetal Diecast, Inc., Jackson, Mich.
Secretary — A. SUGAR, Castmaster, Inc., Bedford, Ohio

Aluminum Die Casting Metal Cost, its practical and analytical evaluation, G. H. FOUND, Arthur D. Little, Inc., Cambridge, Mass. and J. LAPIN, Central Foundry Div., General Motors Corp., Saginaw, Mich.

R. P. DUNN:[53] This paper should prove to be of considerable interest to die casters and others. The authors are to be congratulated on their significant practical contribution to the casting industry.

By plotting the metal cost (as per cent of total cost) of example shown in Table 2 and Fig. 2 vs. the Trimmed casting weight shown in Table 1 and Fig. 1, the graph in this discussion illustrates their relationship. It will be a curve rather than a straight line since it never can reach 100 per cent, no matter how large the casting may be.

Ingot + Scrap* ➡ Good Trimmed Castings + Scrap* + Metal Lost in Dross.

*Recirculated on current basis, scrap consisting of bad castings, runners, gates, biscuits, flash and overflow pads. Therefore, Ingot ➡ Good trimmed castings + Metal loss in dross.

Using their nomenclature and adding

E = Ratio of good castings to total castings made

then

$I = WE + L_m$ (Revised 1)

(or transposing, $L_m = I - WE$)

Their equation (2) is correct: $L_m = I L_i + S L_s$

Combining (Revised 1) and (2):

$L_m = (WE + L_m) L_i + S L_s$ (Revised 3)

gross cast weight = good trimmed castings + scrap

$C = WE + S$

or $S = C - WE$ (Revised 4)

Their equation (5) is correct: $C = \frac{W}{Y}$

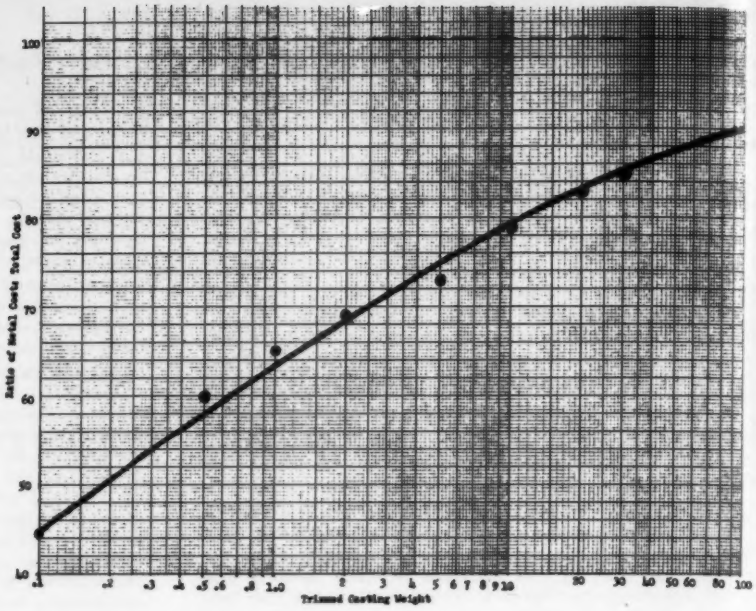
Combining (Revised 4) and (5):

$S = \frac{W}{Y} - WE = (\frac{1}{Y} - E) W$ (Revised 6)

Combining (Revised 6) and (Revised 3):

$L_m = (WE + L_m) L_i + (\frac{1}{Y} - E) W L_s$ (Revised 7)

In their example, they selected an ideal condition where $E = 1$



Graph illustrating relationship of metal cost as per cent of total cost vs. trimmed casting casting weight.

(That is no scrap is produced). Substituting $E = 1$ in (Revised 7), we arrive at their equation (7). On this basis only, then Table 3 and Fig. 3 are valid as shown. If scrap is produced, then E is less than 1 and (Revised 7) would be suitable.

Equation (8) would become

$$L = L_m + B \qquad \text{(Revised 8)}$$

if (Revised 7) is used, since E has already been incorporated in L_m.

Along with reducing casting rejects, flash and die blow, the operator can reduce remelt scrap by keeping biscuit size uniformly at a minimum.

The operator can reduce spillage, oily sweepings, skimmings and sludge. Usually these materials are sold but with proper melting facilities and techniques they can be recovered. Sludge is preventable. However, if sludge forms it is normally best to sell it.

Metallics can be recovered from dross and remelted if proper equipment and techniques are used. In small operations, it is impractical to recover any except the large visable pieces of metal.

Trimmings and borings are usually sold. In small operations this is best but in larger operations it is usually practical to remelt them if proper melting equipment is available and special techniques are followed.

Remelting of all or most of these dirty and/or oxidized (and possibly contaminated) materials can result in considerable savings in money. Due to their inherent lower metallic recovery rates, the overall metal loss in dross will increase and the need for suitable dross metallic recovery methods is obvious. If such equipment is not economically justified, it is doubtful that such lower grade material should be reprocessed.

Normally, the amount of these materials produced under good operational methods is low enough to be omitted from all of the above calculations and formulas without adversely effecting their usefulness and accuracy. However, if these materials are reprocessed, they should be included in the formulas and this would necessitate the development of new and more complicated formulas.

AUTHOR'S CLOSURE: Mr. Dunn's remarks are correct and represent another way of treating the subject. It was the author's intent to simplify the expressions as much as possible and to treat the inevitable scrap generated by expression (8),

$$L = \frac{L_m}{E} + B$$

It should be noted that *C*, as defined by the authors, is gross cast weight per casting and not total weight cast as suggested by Mr. Dunn. It is important that this difference be kept in mind. Either definition is possible, but one requires allowance for casting efficiency at a different stage of the development.

Gray Iron Permanent Molding, H. U. MCCLELLAND, Eaton Mfg. Co., Vassar, Mich.

MR. BLACKMUN:[54] The carbon coating on the mold is important as a parting medium. Does it also insulate the mold? If it does insulate, how is control effected?

H. U. MCCLELLAND: The carbon coating acts as an insulating agent as well as a parting medium. The coat is renewed with each cycle by burning acetylene, in the absence of air, on the mold face. The amount of acetylene burned controls the coat thickness, and thus, the rate of casting solidification. Carbon does tend to build up in mold recesses and must be dressed out periodically in order to avoid trouble.

MR. GALLATIN:[55] What is metal pouring temperature? What is the approximate mold life and the principal cause of mold failure?

H. U. MCCLELLAND: Pouring temperature is 2450/2550 F. The life of a die depends to a large extent on the intricity of the die. There have been dies which have produced up to one million castings and others which failed after 1,000 pieces, the average mold life for a two to three lb casting is 50 to 60,000 pieces. Most mold failure is due to cracking from continual flexing in the heating and cooling of the mold and to carbon burnout.

Ductile Iron Division

Presiding — C. K. DONOHO, American Cast Iron Pipe Co., Birmingham, Ala.
A. W. ANDERSON, International Harvester Co., Chicago.

Secretary — L. J. VENNE, Electro Metalurgical Co., Div. of Union Carbide Corp., Cleveland.

Ductile Iron As-Cast and Annealed Tensile Properties, A. H. RAUCH, J. B. PECK and E. M. MCCULLOUGH, Deere & Co., Moline, Ill.

C. R. ISLEIB[56] *(written discussion):* Statistical work such as the authors have done in preparing this paper is certainly useful to fully understand the interplay of all the many variables in engineering alloys.

Most high quality as-cast ductile irons at the low range of Bhn show elongations 25 to 50 per cent greater than those which would be predicted by the formulas offered in this paper. Are the authors sure that they are properly weighing the effects of flake graphite in their irons, and perhaps other factors, in equation (3)? The well-known deterioration of graphite structure in those irons with excessive magnesium residuals (0.09 to 0.11 per cent) is probably more important

than magnesium carbides, the presence of which are open to some doubt.

Meeting a minimum yield strength is something that faces almost all ductile iron producers. They will be grateful to the authors for showing nickel as the one alloying element which will increase the yield strength of ductile iron about 2500 psi for each ½ per cent nickel without sacrificing ductility or ultimate tensile strength.

The formulas presented indicate that if a foundry operates within a reasonable range of chemical analysis, the Brinell hardness test, along with microexamination for satisfactory graphite structure, is the most satisfactory practical day-to-day control for ductile iron producers.

AUTHORS' REPLY: We do not agree that control of hardness and graphite structure along represents satisfactory control for the tensile properties of ductile iron. The chemistry must also be controlled.

We are certain that no flake graphite was present in any of the test bars represented by the data. As to the presence or absence of magnesium carbides, we can only speculate. The equations merely show that in otherwise identical as cast ductile irons, the higher magnesium contents are associated with lower elongations and that in otherwise identical annealed ductile irons the higher magnesium contents are associated with higher elongations.

MR. SHNAY:[57] In your use of multiple regression analysis is there not a possible error on confidence limits?

AUTHORS' REPLY: Linear approximations to the true forms of the confidence limits were used as the exact method requires re-calculation for each combination of independent variables. We do not enter into serious error. The maximum error would occur if all of the independent variables were simultaneously at the extremes of their ranges, which is highly improbable in our case.

MR. CREWS:[58] What is the quantitative method for metallography, particularly on graphite?

AUTHORS' REPLY: We did not use an exact quantitative method such as sectioning and weighing photomicrographs. All values were estimates made by the same metallographer. He was able to check his results so that if he was in error it was a consistent error.

T. CURRY:[59] Were most of the data from 1 in. or 3 in. keel blocks?

AUTHORS' REPLY: All of the data on fully annealed iron were obtained from 1 in. Y blocks. Some of the data on as-cast iron were obtained from 3 in. Y blocks in order to achieve the lower hardness values.

MR. EAGAN:[60] Pouring temperatures are not considered. This will make wide variations in results. This is work by Deere & Co. and may not fit results from other plants.

AUTHORS' REPLY: We believe that the effects of variations in pouring temperatures on the regression equations, if such effects exist, are included in the confidence limits. We poured over a wide range of temperatures. We also believe that these equations do apply to other foundries. We have examined Y blocks from other foundries and the results conform to these equations.

H. HENDERSON:[61] Nothing is said about the effect of phosphorus.

AUTHORS' REPLY: Phosphorus undoubtedly has an effect on some of these relationships. However, our phosphorus content was relatively low and invariable so that the effect did not appear in the analysis. We have done other work with higher phosphorus ductile iron which showed an effect of phosphorus on the elongation-hardness relationship.

MR. HAINES:[62] You have some results with yield below 40,000 and tensile below 60,000. Is this not an abnormal type of ductile iron?

AUTHORS' REPLY: The results quoted were obtained from fully annealed Y blocks and are not abnormal for this material. The low results shown for the as-cast iron were obtained from 3 in. Y blocks slowly cooled to deliberately achieve low properties.

AUTHORS' CLOSURE: To more fully answer Mr. Schnay, the authors believe that linear approximations to the exact form of the confidence limits are necessary if the regression equations are to have practical value for production men and that the errors introduced by the approximations are not serious. As an example, for equation (2) the variance of a predicted value of yield strength would be:

$$s_p^2 = s_E^2 \left[1 + 1/n + c_{11}(x_1-\bar{x}_1)^2 + c_{22}(x_2-\bar{x}_2)^2 + c_{33}(x_3-\bar{x}_3)^2 + c_{44}(x_4-\bar{x}_4)^2 + c_{55}(x_5-\bar{x}_5)^2 + 2c_{12}(x_1-\bar{x}_1)(x_2-\bar{x}_2) + 2c_{13}(x_1-\bar{x}_1)(x_3-\bar{x}_3) + 2c_{14}(x_1-\bar{x}_1)(x_4-\bar{x}_4) + 2c_{15}(x_1-\bar{x}_1)(x_5-\bar{x}_5) + 2c_{23}(x_2-\bar{x}_2)(x_3-\bar{x}_3) + 2c_{24}(x_2-\bar{x}_2)(x_4-\bar{x}_4) + 2c_{25}(x_2-\bar{x}_2)(x_5-\bar{x}_5) + 2c_{34}(x_3-\bar{x}_3)(x_4-\bar{x}_4) + 2c_{35}(x_3-\bar{x}_3)(x_5-\bar{x}_5) + 2c_{45}(x_4-\bar{x}_4)(x_5-\bar{x}_5) \right]$$

In order to obtain exact confidence limits, this equation must be re-calculated for each combination of independent variables used. The authors used the following approximation which does not require re-calculation:

$$s_p^2 = s_E^2(1 + 1/n)$$

This approximation is, of course, exact only at the means of the independent variables which are, in the authors' company, near the most probable values. The exact confidence limits at the extremes of the independent variables are ± 3850 psi. This compares to the linear approximation of ± 3570 psi. The maximum error is, therefore, 7.5 per cent of the exact value. The authors could have made a better approximation over the ranges of independent variables by taking an average between the confidence limits at the means and at the extremes.

As was brought out in subsequent discussion of Mr. Eagan's question, the Y blocks from other foundries which the authors checked with the equations all represented basic cupola operations. It is possible that ductile iron from electric furnace or acid cupola operations would not check. However, the writers believe that the reason for any apparent discrepancies could be found in the differences in content of residual elements, particularly those listed in Table 1.

As to Mr. Isleib's 25 to 50 per cent greater elongation for most high quality as-cast ductile iron in the lower hardness range, we doubt that such elongations exist. At a hardness of 197 Bhn and with an average chemistry, we show an average elongation of 12.5 per cent and a minimum of 9.6 per cent. A 50 per cent greater average elongation would be 18.75 per cent. Most minimum elongation specifications for annealed ductile iron are 10 to 15 per cent.

Presiding — K. D. MILLIS, International Nickel Co., New York.
H. W. RUF, Grede Foundries, Inc., Milwaukee.
Secretary — G. KRUMLAUF, Republic Steel Corp., Cleveland.

Heat Treatment of Ductile Iron, W. D. MCMILLAN, International Harvester Co., Chicago.

MR. PARKE:[63] Does the iron referred to in your paper contain excessive carbides?

W. D. MCMILLAN: Excessive carbides are present in thin sections. Carbides are not troublesome in normal sections of our work.

M. NEPTUNE:[64] If carbides are present, is there much dimensional change upon annealing?

W. D. MCMILLAN: Dimensional changes are normal for ferrous materials of this general analysis.

E. WEHLANDER:[65] Do you find definite changes in microstructure and machinability with some critical anneals?

W. D. MCMILLAN: We find some change in microstructure. Also, some improvement on machinability. I have no data as to actual machinability improvement.

A. H. RAUCH:[66] Do you normalize castings not being heat treated?

W. D. MCMILLAN: Yes.

J. BECK:[67] Do you use rare earth additions to prevent carbide formation?

W. D. MCMILLAN: No.

MR. HUTCHINS:[68] Can you obtain desired low temperature properties by oil quench from 1650 F tempered 1300 F?

W. D. MCMILLAN: We obtain good low temperature properties by oil quench in tempering. We have obtained ductile fractures on this material as low as −100 F on unnotched impact tests.

R. BECK:[69] Should material be normalized before quench in tempering heat treatment?

W. D. MCMILLAN: This depends on the amount of carbon in solution. You must have about 0.50 combined carbon for good

response. Jominy hardenability somewhat like 8640 steels.

A. AVEDESIAN:[70] What is mold movement or dimensional change on various types of inoculation?

W. D. MCMILLAN: This depends upon physical structure of the material.

Carbon Flotation in Ductile Iron, A. H. RAUCH, J. B. PECK and G. F. THOMAS, Deere & Co., Moline. Ill.

MR. PATTERSON:[71] Can you establish amount of carbon flotation reliably by metallographic methods?

J. B. PECK: Yes. This method is used for our routine determinations.

J. MERCER:[72] Is there any effect of section size upon carbon flotation?

J. B. PECK: Yes. even in comparatively light sections. Gating and molding techniques also effect flotation. Pure carbon flotation is generally less than 1/8 in. Carbon flotation associated with dross may be as deep as 1 in.

H. WILDER:[73] Does silicon content vary throughout cross-section depending upon carbon values?

J. B. PECK: The silicon content remains relatively constant over casting sections.

J. MERCER: Are you convinced that carbon wafers are dependable for chemical analysis and how do you establish it?

J. B. PECK: Carbon wafers were checked against calculated carbon values and found to be within close limits. Carbon determinations were run both by induction and gravimetric methods.

E. C. MATHIS:[43] Can you completely suppress carbon precipitation in pin samples and what is diameter of pins used?

J. B. PECK: Pins are 0.35 in. diameter. On occasion, they show some small amount of free carbon for high carbon irons.

MR. FERGUSON:[74] Have you checked this carbon flotation defect with relation to pouring temperature?

J. B. PECK: Not as yet. Work is to be started on this phase of our project in the near future.

A. DUBLO:[75] Do you prefer to lower silicon content to control carbon equivalent?

J. B. PECK: No.

K. MILLIS:[76] Have you investigated the effect of various fluxing agents upon dross defects?

J. B. PECK: Yes. Various fluxes do not seem to have any particular effect.

AUTHORS' CLOSURE: Since the time this paper was prepared, another type of cope surface or flotation condition has been encountered and investigated. This condition is illustrated in Fig. A, the photograph of a fracture surface. This condition is generally considerably deeper than that encountered with carbon flotation alone. The authors believe that the upper affected layer consists of a combination of floated or excess graphite along with dross or magnesium sulfides or oxides.

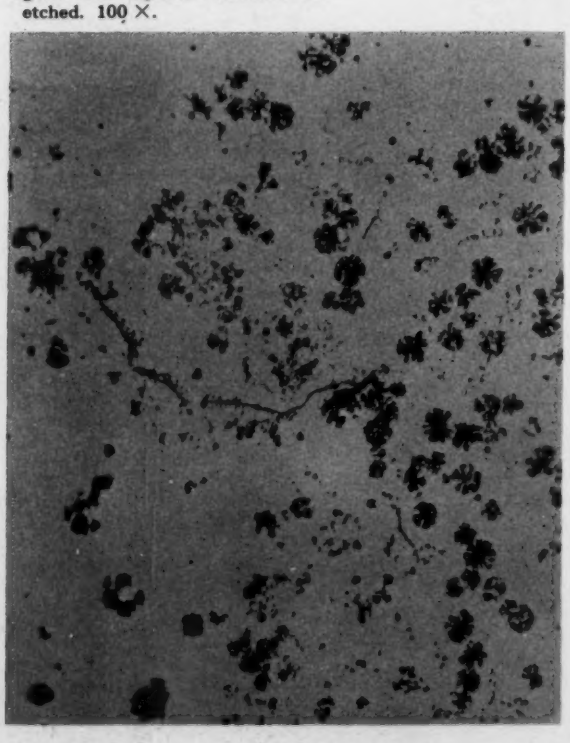
Graphite "films" or membranes, along with "exploded" and other degenerate type of graphite nodules are generally found associated with this condition as shown in Fig. B, a photomicrograph at 100× in the affected layer. Large amounts of "bullseye" ferrite, i.e., ferrite around graphite, is also generally present as shown in Fig. C at 50×.

The carbon equivalent in the unaffected material did not exceed 4.6 and the carbon contents were in the range of 3.55 to 3.65 per cent; however, the carbon equivalent in the affected



Fig. A — Defect as it appears on a fractured face. 1 ×.

Fig. B — Defect area showing the presence of films, graphite in fine crab-like and star-shaped aggregates and exploded nodules. Unetched. 100 ×.



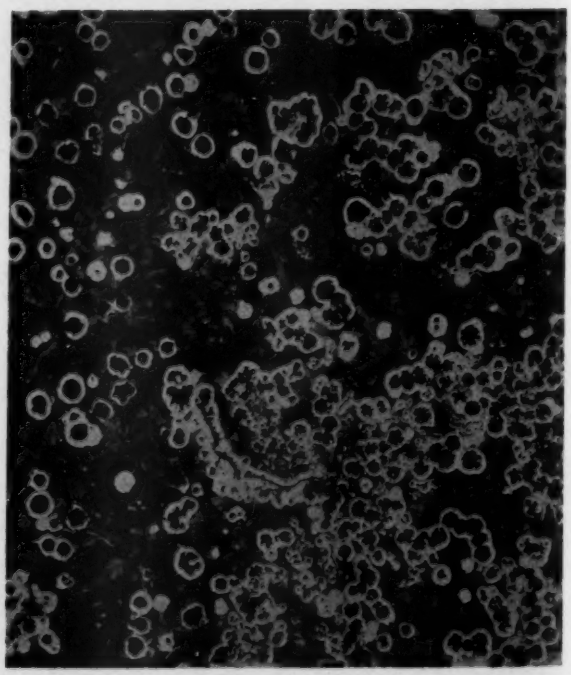


Fig. C — Same field as Fig. B etched to show heavy concentration of ferrite associated with abnormal graphite aggregates. 50 ×.

area varied from 5.3 to 6.0 with carbon contents in the range of 4.3 to 5.1. One explanation that has been put forward is that films of magnesium oxides or sulfides trap excess graphite nodules and prevent these nodules from floating to the surface. Pouring temperature may also have an effect on the depth of the affected layer.

Tin Effect on Structure and Properties of Flake and Nodular Graphite Cast Irons, E. C. ELLWOOD, Tin Rsch. Institute, London, England.

Presented by — J. A. DAVIS, Battelle Memorial Institute, Columbus, Ohio.

MR. CRUSE:[77] When and how was tin added?

J. A. DAVIS: Tin was added both to the furnace and to ladles. The method of addition normally does not make any difference, although ladle recovery is slightly less than that obtained with furnace additions. Ladle recovery is in the order of 90-95 per cent.

W. D. MCMILLAN:[78] Does tin stabilize pearlite?

J. A. DAVIS: We do not definitely know.

K. MILLIS:[79] Does tin affect carbide stability?

J. A. DAVIS: It does not appear to affect carbide stability.

T. HAINES:[62] Does tin carry through upon remelting?

J. A. DAVIS: Yes, tin is not completely oxidized upon remelting.

A. H. RAUCH:[66] What is the effect of tin upon impact values?

J. A. DAVIS: Tin has about the same effect as in gray iron. About 0.10 tin reduces impact.

J. OHLE:[80] Our work confirms data presented in this paper. We would like to suggest that additional work be carried out on the effects of tin.

Fundamental Papers Division

Presiding — J. F. WALLACE, Case Institute of Technology, Cleveland.
C. F. WALTON, Gray Iron Founders' Society, Cleveland.

Secretary — G. K. TURNBULL, Case Institute of Technology, Cleveland.

Solidification Mechanisms of Eutectic and Gray Iron, A. DESY, University of Ghent, Ghent, Belgium.

R. A. FLINN:[81] Could you continue this discussion to include the case of spheroidal graphite? Do sulfur, etc., apply in this case also?

A. DESY: This would be a third transformation mechanism:

1) This type is the case of pearlite coming out of austenite. This is a case of homogeneous nucleation along an advancing front and there is only diffusion in the liquid.
2) This is the anomolous type described in report. It is a race between two growths: γ wins. It is not normal eutectic or proeutectic, but something in between. It can be considered as proeutectic. There is no reason why we can't consider growth through austenite.
3) This is the spheroidal type of transformation. Is graphite nucleated in the melt or the austenite? It is likely that it begins in the melt and is rapidly surrounded by austenite and continues to grow as proeutectic.

A similar effect is anticipated for sulfur in case 3.

C. LOPER:[52] Since silicon ties up the iron and causes precipitation of graphite, would a white iron result if the iron and carbon were kept together (silicon separate)?

A. DESY: Yes, but how could this be done?

J. F. WALLACE:[82] This paper shows that rosettes are a combination of types A and D graphite, and that the center forms rapidly and is followed by normal growth before it is surrounded. Do you attribute reduced expansion in spheroidal irons (compared to normal gray irons with equal graphite) to mold wall movement?

A. DESY: Yes.

Grain Refinement of Solidifying Metals by Vibration, R. G. GARLICK and J. F. WALLACE, Case Institute of Technology, Cleveland.

J. R. HORRIGAN:[83] What effect does amplitude have? What was the amplitude at 60 cycles/sec?

R. G. GARLICK: Both amplitude and frequency have an effect on grain refinement below a certain energy induced into the melt. We were above it. The amplitude was on the order of 0.1 in.

J. VARGA:[28] In the curve of grain refinement vs. solidification contraction was data from literature for comparable frequencies? In some Russian work grain refinement was a function of frequency.

R. G. GARLICK: Yes. The criterion we used was whether the statically cast ingots were coarse grained and whether sufficient energy was introduced.

A. DESY:[84] Please elaborate on the correlation between grain refinement and solidification contraction.

R. G. GARLICK: Pressure wave in vibration drives reaction in proportion to volume change on solidification.

D. V. DOANE:[85] Elaborate on increase of plate lengths and decrease of cell size for eutectics.

R. G. GARLICK: Cell size is determined by the number of effective heterogeneous nuclei. Since vibration increases the number of effective heterogeneous nuclei, the cell size is decreased. Growth of constituents in any cell takes place from these nuclei. In a vibrated ingot the nucleation temperature will be higher and more rapid envelopment of the flake is suppressed since diffusion allows growth of the flakes.

Mass Effect on Castings Tensile Properties, a rationalization, P. J. AHEARN, Watertown Arsenal, Watertown, Mass., G. W. FORM and J. F. WALLACE, Case Institute of Technology, Cleveland.

H. W. LOWNIE:[86] How did you get solidification time data?

G. W. FORM: From cooling curves reported in the literature.

H. W. LOWNIE: Actually, does not one casting represent a range of solidification times?

G. W. FORM: Yes, it was for this reason that the specimens were selected from the thermocouple location which was in the center of each casting.

K. A. BECKIUS:[87] Solidification time is a good parameter, but since you have a range of solidification within one casting, would you indicate where the specimens were taken?

G. W. FORM: As mentioned above, the specimens were taken from the center where also the thermocouples were located. We did not carry out any experiments ourselves, but collected data from the literature and used the reported times of solidification. In some instances we made an educated guess as far as these times were concerned. In order to bring out qualitatively the principle exposed here, it is not so important that the freezing time data are exact. Moreover, the data reported in the literature scatter appreciably, so that an educated guess can not be avoided.

R. W. RUDDLE:[49] I would like to suggest an exception to your findings. Al — 10 per cent Mg has good ductility but is section sensitive.

J. F. WALLACE: This also applies to Cu-Sn, but here the section sensitivity is a result of interdendritic shrinkage cavities due to the large spread between solidus and liquidus. We have reported on data stemming from castings which were sound. In case of an unsound casting, effects from defects are superimposed on the mass effect.

R. W. RUDDLE: This is true, but I believe that shrinkage is the primary and the stress-strain relationship a second order variable.

J. F. WALLACE: Both variables seem to be first order.

Presiding — S. LIPSON, Frankford Arsenal, Philadelphia.
H. R. LARSON, American Brake Shoe Co., Mahwah, N. J.

Secretary — G. A. COLLIGAN, University of Michigan, Ann Arbor.

Dimensioning of Sand Casting Risers, H. D. MERCHANT, Case Institute of Technology, Cleveland.

J. B. CAINE:[26] The author raises some questions as to the factual bases of the theories advanced by this discusser. At the time of publication these data could not be presented because of a restrictive publication policy. After a period of over ten years it would seem that such a policy is only interfering with worthwhile continuing work such as presented in this paper and it is high time these details be released.

The experimental curve in Fig. 4 is the point under riser shrinkage and solidity as determined by hundreds of simple castings produced in a score of steel foundries throughout the United States. The experimental curve of Fig. 4 is the average line, but the scatter on each side of the average is surprisingly small. Practically all the points that did not check the curve

could readily be explained most of all by riser freeze over.

These same experimental castings established the minimum riser section required for steel as 108 per cent of the casting section, in answer to paragraph 3, page 95. This point is quite important. It should be emphasized that the casting section is not necessarily the geometric casting section. The casting section for risering can be quite larger than the geometric section in areas of restricted heat flow.

Item 2, page 95, is probably the most important point for discussion and possibly is the reason for some of the points of disagreement. In the original work, resulting in equation A, page 95, the area of contact between the casting and riser was subtracted from both casting and riser surface area. This seeming detail is quite important. It was impossible to correlate the experimental results with any equation unless this was done.

The point raised by the author in item (1) page 95 warrants discussion. This discusser's evaluation of shrinkage was by visual examination of sections thru the centerline of the test castings. The error when comparing with NRL radiographs of ¼-in. slices at the centerline is, with one exception of not too great consequence for steel.

This metal shows little variation in dispersed shrinkage in areas adjacent to the riser. The evaluation with other metals, especially the non-ferrous metals is not this simple. The one exception for steel is heavy sections. Shrinkage in these heavy sections will vary greatly if evaluated by radiographs of the section as a whole, or a thin slice taken from the section.

This discusser has no brief for the original equation A of page 95. In fact he is surprised it has held up as well as it has. The form was selected only as the simplest equation that would fit the evidently hyperbolic function. If any other form will fit more exact data it should be used.

There is only one point of disagreement, item (7) page 104. Why should H/D be greater than one for light metals? High risers have not worked for the heavier metals, why should they work for the light alloys?

High Strength Steel Castings, mechanical properties and processing techniques, K. D. HOLMES, J. ZOTOS and P. J. AHEARN, Watertown Arsenal, Watertown, Mass.

H. A. FABERT:[88] The authors are to be congratulated for conducting work and preparing a paper of this kind. Such information effectively counters oftentimes preconceived ideas as to levels of phosphorus and sulphur reduction that can be attained through discerning melting techniques. In this regard, I wonder whether the authors would care to comment on the carbon level in the initial furnace charge? Also, about the lower limit of decarburization and the methods used for recarburization. Total heat time would also be of interest.

J. ZOTOS:[89] Using a two slag practice, the carbon level of the steel bath at meltdown was between 0.5 to 0.7 per cent. Addition of iron ore during the oxidizing period, and subsequent exposure to a wash slag prior to commencing the reducing period, decarburized the steel to the 0.05 per cent level. Recarburization was achieved by the addition of "carburite" (Carburite is made up of 70 per cent carbon and 30 per cent steel scrap) to the molten metal.

Starting with a cold scrap steel charge weighing 6½ tons, the time consumed from power-on to tap was 5½ hr. Approximately 1½ hr were required from power-on to meltdown while the additional 4 hr were consumed refining the molten steel.

S. L. GERTSMAN:[90] How did you determine the percentage fibrosity on the fractured test bars? If this was carried out visually, the figures shown would imply greater accuracy than is possible by such a procedure.

P. J. AHEARN:[91] The Physical Test Laboratory at Watertown Arsenal employs a visual comparative test to determine the percentage vibrosity on fractured test bars. Technicians having many years experience are able to measure this per cent vibrosity within two or three per cent accuracy and their results can be checked using a linear point count system if necessary.

Heat Transfer Division

Presiding — R. C. SHNAY, Canada Iron Foundries, Ltd., Toronto, Ont., Canada.

Secretary — W. K. BOCK, National Malleable & Steel Casting Co., Cleveland.

Modern Quenching Oils and Techniques for Foundry Heat Treatment, N. F. SQUIRE, Aldridge Industrial Oils, Inc., Cleveland.

V. PASCHKIS[92] *(written discussion):* It is important that the foundryman confronted with heat treating be familiar with the factors influencing quenching operations. But since many castings which have to be heat treated are rather heavy the importance of the cross-section for the cooling and hardening process should be stressed.

Figures 3 to 5 of the author show separate curves for "surface" and "core" or "center" but do not state shape or size of the piece. The thickness will greatly influence the spread of the two curves. For Fig. 6 information would be helpful regarding location of the temperature observation (surface, center, etc.) and the shape and size of the body.

The writer also wonders regarding the statement, "The difference between shallow hardening and deep hardening of steel is the cooling rate required through a critical temperature zone." The final hardness at any position is determined by the cooling rate at this position. And this rate, in turn, is a function of surface rate and geometry. A given surface rate may result in hardening in depth with one geometry and in shallow hardening with another.

The importance of shape and size may become obvious from the four slides, taken from the writer's paper "Quenching—A Thermal Problem," presented at the A.S.M. Annual Meeting 1958.[1] This paper is based on extensive experimental studies carried out by G. Stolz in our laboratory.[2,3,4]

REFERENCES

1. "Quenching — A Thermal Problem," V. Paschkis, A.S.M. Symposium on Quenching of Steels, Cleveland, Oct. 1958.
2. "How Measurements Lead to Effective Quenching," V. Paschkis & G. Stolz, Jr., *Iron Age,* pp. 95-97, 1956.
3. "Quenching as a Heat Transfer Problem," V. Paschkis and G. Stolz, Jr. *Journal of Metals,* pp. 1074-1075, 1956.
4. "Thermal Considerations in Oil Quenching," to be published, *Brit. Iron & Steel Ins.,* Spring 1959.

AUTHOR'S CLOSURE: Dr. Paschkis' question as to the size and section of the pieces used in establishing the curves in Figs. 3, 4, 5 cannot be answered, for these curves are purely schematic to show the different types of quenching and temper heat treating.

Similar diagrams are shown in the *Atlas of Isothermal Transformation Diagrams,* published by the United States Steel Co.

Figure 6 is also basically schematic, although a number of companies have established similar curves, using high speed potentiometers in the cores of either ½-in. round or 1-in. rounds of stainless steel.

To Dr. Paschkis' question of the statement that "The difference between a shallow hardening steel and a deep hardening steel is the cooling rate required through a critical temperature zone of approximately 1200 F to 800 F (649 C to 427 C), so a complete martensitic structure can be obtained." The writer does not see any particular area of disagreement.

In the third paragraph of the quotation from A.S.M. *Metals Handbook* the variables of geometry, size and section as well as composition are recognized.

Again under the sub-heading "Section Thickness," and in Table 1, is shown a case history of a wide range of sectional sizes for three specific compositions.

The writer appreciates Dr. Paschkis' comments because they serve to again emphasize that there are many variables for the heat treater to contend with and any equipment or quenching media that will minimize these variables should be seriously investigated.

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